TRANSACTIONS

of the

American
Foundrymen's Association

Proceedings of the

FORTY-SECOND ANNUAL MEETING

Cleveland, Ohio

May 14 to 19, 1938

VOLUME XLVI

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of the

American Foundrymen's Association
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Geo. A. Seyler, Lunkenheimer Co., Cincinnati, O.
A. Walcher, American Steel Foundries, Chicago, Ill.
*H. Bornstein, Deere & Co., Moline, Ill.

* Member of Executive Committee.
† Association office, 222 W. Adams St., Chicago.

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Summary of Proceedings
of the
42nd Annual Meeting

The Forty-Second Annual Convention of the AMERICAN FOUNDRYMEN'S ASSOCIATION was held in Cleveland, O., May 16 to 19, 1938. In conjunction with the Convention, the Association held a major Foundry Show, with exhibits of foundry equipment and supplies, these being displayed in the Cleveland Public Auditorium Exhibition Halls from May 14 to 19. For the first time in several years, the exhibit was opened on a Saturday, this Saturday opening being designated as Northeastern Ohio Day, with some 4,000 visitors. The total registered attendance for the convention was 5,700, and the total attendance figures being over 11,000, the greatest in the history of the Association. The number of exhibitors was 209 using exhibit space of 58,835 square feet. While the space used was not the greatest in A. F. A. exhibit history, it compared very favorably with the record exhibits of past years and the exhibit as a whole was judged outstanding in its interesting displays of new equipment.

The program of papers and committee reports was considered by all without doubt as being the most valuable collection gathered together for a convention of the Association.

Supreme confidence in the future possibilities of the industry may be said to have been the keynote expressed by the exhibit and convention sessions, discounting the current expressed condition of business as a whole. Optimism seemed to spread throughout the entire convention.

An outstanding entertainment feature of the week was the vaudeville show and buffet supper sponsored by the Northeastern Ohio Chapter and held in the Little Theatre of the Public Auditorium. Over 2,600 foundrymen and guests gathered for this affair.

OPENING SESSION

Monday, May 16, 9:30 A. M.

The Convention was opened officially Monday morning at a session. with Hyman Bornstein, Deere & Co., Moline, Ill., and the president of the Association presiding. Members and guests were welcomed to Cleveland by B. G. Parker, Youngstown Foundry and Machine Co., Youngstown, O., and general chairman of the Northeastern Ohio Chapter. Mr. Parker outlined the various entertainment features arranged by the Northeastern Ohio Chapter, as host to the Convention guests. President Bornstein next introduced Harold H. Burton, mayor of Cleveland, who extended official greetings of the city. Mayor Burton in his address stressed the importance that machinery must play in the future of the country and discussed briefly the need for government, properly organized and administered, wherever industrial competition exists. President Bornstein then expressed the appreciation of the Association for the sincere welcome extended by the city and the Northeastern Ohio Chapter.

Frank G. Steinebach, chairman of the International Relations Committee, next read a cablegram of greetings from the Institute of British Foundrymen. This read:

"Institute of British Foundrymen send cordial greetings to American Foundrymen's Association and sincere wishes for successful convention. Hope to welcome many American friends to International Congress at London in 1939."

Signed, C. W. BIGG, President

TOM MAKEMSON, Secretary

Mr. Steinebach then presented the following motion:

"That the International Relations Committee be authorized to cable our sincere appreciation and best wishes to the Institute of British Foundrymen."

This motion was duly seconded by Dr. J. T. MacKenzie and approved by a unanimous vote.

Announcement was made by President Bornstein of the appointment of a nominating committee to present to the Annual Business Meeting names of four men to serve with the three last past-presidents as the 1939 Nominating Committee. Those appointed were:

Fred J. Walls, International Nickel Co., Inc., Detroit, Mich., as chairman.

Harold W. Johnson, Greenlee Foundry Co., Chicago, Ill.

R. J. Allen, Worthington Pump & Machinery Corp., Harrison, N. J.

It was requested that this committee report at the Annual Business Meeting.

After several other announcements, the meeting was declared adjourned.

GRAY CAST IRON

Monday, May 16, 10:00 A. M.

Chairman—Dr. Jas. T. MacKenzie, American Cast Iron Pipe Co., Birmingham, Ala.

Vice Chairman—A. W. Weston, Chicago Hardware Co., No. Chicago, Ill.

The following papers and report were presented and discussed:

Hints on Molding and Preparation of Castings for Enameling, by R. B. Schaal, Ferro-Enamel Corp., Cleveland, O.

Relation of Microstructure to Enamelability of Cast Iron, by G. H. Spencer-Strong, Porcelain Enamel Mfg. Co., Baltimore, Md.

Report of Committee on Classification of Graphite Size, presented by committee chairman W. E. Mahin, Westinghouse Electric & Mfg. Co., East Pittsburgh, Pa.

NON-FERROUS FOUNDING

Monday, May 16, 10:00 A. M.

Chairman—Harold J. Roast, Canadian Bronze Co., Montreal, Can.

Vice Chairman—H. M. St. John, Detroit Lubricator Co., Detroit, Mich.

The following papers were presented and discussed:

Heat Treating Zinc Bronze Pressure Castings to Close up Leakage, by H. Fleck and T. C. Bunch, Pearl Harbor Navy Yard, Honolulu, T. H. In the absence of the authors, this paper was abstracted by T. C. Watts, Falcon Bronze Co., Youngstown, O.

Risers and Gates for Non-Ferrous Castings, by A. E. Cartwright and C. C. Brisbois, The Robert Mitchell Co., Ltd., Montreal, Can. This paper was presented by Mr. Cartwright.

Modification of the Saeger Fluidity Test as Applied to Red Brass, by Geo. P. Halliwell, H. Kramer Co., Chicago, Ill.

At this time there was convened the regular annual Non-Ferrous Business Meeting. Division chairman H. M. St. John presided. As one item of business, Chairman St. John announced the election of officers of the Division for the coming year, which were as follows:

> Chairman to serve two years: Harold J. Roast, Canadian Bronze Co., Montreal, Canada.
>
> Vice Chairman to serve two years: Wm. J. Laird, Westinghouse Electric & Mfg. Co., East Pittsburgh, Pa.

As members of the Advisory Committee to serve four years:

> T. C. Watts, Falcon Bronze Co., Youngstown, O.
>
> R. W. Parsons, The Ohio Brass Co., Mansfield, O.
>
> C. V. Nass, The Fairbanks Morse & Co., Beloit, Wis.

Two committee reports presented were the Report of the Committee on the Analysis of Defects by committee chairman H. M. St. John, and the report of the Committee on Recommended Practices by committee chairman H. J. Rowe.

Management — Foreman Training

Monday, May 16, 2:00 P. M.

Chairman—A. D. Lynch, J. I. Case Co., Racine, Wis.

At this meeting A. C. Horrocks, Educational Director, National Association of Foremen, and Goodyear Tire and Rubber Co., Akron, O., addressed the audience on the subject of "Problems of the Executive in Connection with Foreman Training."

Pattern Making

Monday, May 16, 2:00 P. M.

Chairman—Vaughan Reid, City Pattern Works, Detroit, Mich.
Vice Chairman—H. M. Harrold, Caterpillar Tractor Co., Peoria, Ill.
The following papers and report were presented and discussed:
Modern Pattern Production, by E. T. Kindt, Kindt-Collins Co., Cleveland, O.
Pattern Developments and Molding Methods, by Edward Hynan, Saginaw Malleable Iron Division, General Motors Corp., Saginaw, Mich.
Pattern Making Apprenticeship Standards, by committee chairman B. B. Wittfoht, Caterpillar Tractor Co., Peoria, Ill.

Gray Cast Iron

Monday, May 16, 2:00 P. M.

Chairman—Fred J. Wells, International Nickel Co., Detroit, Mich.
Vice Chairman—Garnet P. Phillips, International Harvester Co., Tractor Works Chicago, Ill.
The following papers were presented and discussed:
Physical Properties of Cast Iron in Heavy Sections, by E. R. Young, V. A. Crosby, and A. J. Herzig, Climax Molybdenum Co., Detroit, Mich. Mr. Young presented this paper.
Some Effects of Blast Pressure on Cupola Operations, by C. K. Donoho and Dr. Jas. T. MacKenzie, American Cast Iron Pipe Co., Birmingham, Ala. This paper was presented by Mr. Donoho.
Characteristics of Cokes for Cupolas. A review of a paper originally presented before the Institute of British Foundrymen, by J. G. Pearce and Dr. Hugh O'Neill. This review was presented by J. A. Bowers, American Cast Iron Pipe Co., Birmingham, Ala.
Cupola Practice, by Marcel Dudouet, Ecole Superieure de Fonderie, Paris, France. This was the official exchange paper presented on behalf of the French Foundry Technical Association. In the absence of the author, W. H. Spencer, Sealed Power Corporation, Muskegon, Mich., presented the paper in abstract form.

Cast Iron Shop Practice Course (Session 1)

Monday, May 16, 4:00 P. M.

Chairman—Horace Deane, Deere & Co., Moline, Ill.
Donald J. Reese, International Nickel Co. Inc., New York, N. Y., presented a paper on "Cupola Operation."

ENGINEERING AND SHOP PRACTICE INSTRUCTORS DINNER

Monday, May 16, 6:30 P. M.

Chairman—Frank G. Steinebach, editor, The Foundry, Cleveland, O.

Fred J. Sefing, International Nickel Co. Inc., Bayonne, N. J., presented a discussion on "What a Course in Metallurgy Should Teach."

Carl H. Casberg, University of Illinois, Urbana, Ill., discussed the subject: "How Can a Foundry Course in an Engineering School Meet the Highest Requirements of an Engineering Course?"

FOUNDRY REFRACTORIES

Monday, May 16, 8:00 P. M.

Chairman—J. A. Bowers, American Cast Iron Pipe Co., Birmingham, Ala.

Vice Chairman—Elmer J. Carmody, Ada, Mich.

The following papers were presented and discussed:

Effects of Cupola Practice on Linings, by C. K. Donoho, American Cast Iron Pipe Co., Birmingham, Ala., and Chas. F. Greene, Atlantic Refractories Co., Macon, Ga. Mr. Donoho presented this paper.

Refractories for Foundry Ladles, by C. E. Bales, Ironton Fire Brick Co., Ironton, O.

Characteristics of the Silica Refractories Used in the Cupola, by W. J. Rees, head Refractories Department, University of Sheffield, Sheffield, England. This was the annual Exchange paper and was presented on behalf of the Institute of British Foundrymen. In the absence of the author, E. J. Carmody abstracted the paper.

MANAGEMENT — JOB EVALUATION

Monday, May 16, 8:00 P. M.

Chairman—H. C. Robson, Continental Roll & Steel Foundry Co., East Chicago, Ind.

The following paper was presented and discussed:

Job Evaluation for the Smaller Foundry, by E. L. Roth, Motor Castings Co., Milwaukee, Wis.

A demonstration of the occupational rating plan of the Industrial Management Society was given by Ralph H. Landes, Western Electric Company and Director of Study and Research, Industrial Management Society, Chicago, Ill.

SAND SHOP PRACTICE COURSE (SESSION 1)

Tuesday, May 17, 8:30 A. M.

Chairman—D. Frank O'Connor, Walworth Co., Boston, Mass.

The subject for discussion at this session was "Practical Problems of Non-Ferrous Sand Control," A. C. Arbogast, Northern Indiana Brass Co., Elkhart, Ind., opening the meeting with comments on this topic, which was then discussed.

NON-FERROUS CASTINGS
Tuesday, May 17, 10:00 A. M.

Chairman—H. M. St. John, Detroit Lubricator Co., Detroit, Mich.
Vice Chairman—R. W. Dayton, Battelle Memorial Institute, Columbus, O.

The Role of Silicon in Non-Ferrous Castings, by H. W. Gillett, Battelle Memorial Institute, Columbus, O. In the absence of the author, Dr. C. H. Lorig Battelle Memorial Institute, abstracted this paper.

A Description of the Age Hardening Process as Applied to Castings, by L. W. Kempf, Research Laboratories, Aluminum Co. of America, Cleveland, O.

Production of Castings in Age Hardenable Nickel-Tin Bronzes, by T. E. Kihlgren, International Nickel Co. Inc., Bayonne, N. J.

STEEL CASTINGS
Tuesday, May 17, 10:00 A. M.

Chairman—John Howe Hall, Germantown, Philadelphia, Pa.
Vice Chairman—W. C. Hartman, Bethlehem Steel Co., Bethlehem, Pa.

The following report and paper were presented and discussed:

Report of Committee on Radiography, by committee chairman C. W. Briggs, Naval Research Laboratory, Washington, D. C.

Effects of Aluminum on the Physical Properties of Medium Carbon Cast Steel, by C. E. Sims and F. B. Dahle, Battelle Memorial Institute, Columbus, O. Mr. Sims presented this paper.

GRAY IRON FOUNDING
Tuesday, May 17, 10:00 A. M.

Chairman—John W. Bolton, Lunkenheimer Co., Cincinnati, O.
The following papers were presented and discussed:

Deoxidation and Graphitization, by R. G. McElwee, Vanadium Corp. of America, Detroit, Mich.

Comparative Effects of Late Additions of Silicon and Titanium to Cast Iron, by G. F. Comstock and E. R. Starkweather, Titanium Alloy Mfg. Co., Niagara Falls, N. Y. This paper was presented by Mr. Starkweather.

The Formation of Graphite in Gray Iron, by Alfred Boyles, Battelle Memorial Institute, Columbus, O.

NON-FERROUS DIVISION LUNCHEON AND ROUND TABLE CONFERENCE
Tuesday, May 17, 12:30 P. M.

Chairman—H. J. Rowe, Aluminum Co. of America, Cleveland, O.
The following paper was presented and discussed:

Crucible Melting Furnace Developments, by R. H. Stone, Vesuvius Crucible Co., Swissvale, Pa.

Following this, the discussion was given over to a review of "Analysis of Causes and Remedies for Certain Types of Defective Castings," with C. V. Nass, Fairbanks Morse & Co., Beloit, Wis., acting as discussion leader.

Gray Iron Division Luncheon and Round Table Conference
Tuesday, May 17, 12:30 P. M.

Chairman—H. Kenneth Briggs, Western Foundry Co., Chicago, Ill.
Vice Chairman—S. C. Massari, Association of Chilled Car Wheel Manufacturers, Chicago, Ill.

At this Round Table Conference, the discussion revolved around the subject of "Graphitization and Inclusions in Cast Iron."

Management — Apprentice Training
Tuesday, May 17, 2:00 P. M.

Chairman—C. J. Freund, University of Detroit, Detroit, Mich.
The following report and papers were presented and discussed:
Report of A.F.A. Committee on Apprentice Training, presented by committee chairman V. J. Hydar, Falk Corp., Milwaukee, Wis.
Wisconsin Training Plan, by J. E. Tepoorten, Coordinator, Occupational Extension Service, Wisconsin Vocational Schools, Madison, Wis.
Foundry Apprenticeship in a Commercial Shop, by E. N. Brough, Fairbanks Morse & Co., St. Johnsbury, Vt.
Canadian Apprentice Training Situation in Western Canada, by A. W. McCallum, Anthes Foundry, Ltd., Winnipeg, Can.

Management — Foundry Costs
Tuesday, May 17, 2:00 P. M.

Chairman—Sam Tour, Lucius Pitkin, Inc., New York, N. Y.
Reviews of the melting cost methods of the Non-Ferrous, Steel, Malleable and Gray Iron industries were presented and discussed as follows:
Non-Ferrous Melting Costs, by A. E. Grover, Berea, O.
Calculating Cost of Steel in Castings, by W. J. Corbett, Atlas Steel Castings Co., Buffalo, N. Y. This paper was presented by R. L. Collier, Steel Founders' Society, Cleveland, O.
Malleable Melting Costs, by R. E. Belt, Malleable Founders' Society, Cleveland, O., and C. S. Anderson, Belle City Malleable Iron Co., Racine, Wis. Mr. Belt presented this paper.
Melting Costs in Gray Iron, by P. E. Rentschler and V. L. Diefenbacher, Hamilton Foundry & Machine Co., Hamilton, O., and W. J. Grede and R. L. Lee, Liberty Foundry Co., Wauwatosa, Wis. This paper was presented by Mr. Rentschler.

Steel Castings
Tuesday, May 17, 2:00 P. M.

Chairman—L. E. Everett, Key Co., East St. Louis, Ill.
The following paper and reports were presented and discussed:
A Study of Steel Molding Sands, by H. W. Dietert, E. E. Woodliff and J. A. Schuch, Harry W. Dietert Co., Detroit, Mich. Mr. Dietert presented this paper.
Report of Committee on Methods of Producing Steel for Castings, presented by committee chairman, H. D. Phillips, Dodge Steel Co., Philadelphia, Pa.

Report of Committee on Heat Treatment of Steel Castings, presented by committee chairman, D. C. Zuege, Sivyer Steel Casting Co., Milwaukee, Wis.

CAST IRON SHOP COURSE (SESSION 2)

Tuesday, May 17, 4:00 P. M.

Chairman—L. G. Korte, Atlas Foundry Co., Detroit, Mich.

At this session a discussion on "Cupola Practice—Burnt Iron" was lead by Dr. Jas. T. MacKenzie, American Cast Iron Pipe Co., Birmingham, Ala.

SAND SHOP COURSE (SESSION 2)

Wednesday, May 18, 8:00 A. M.

Chairman—Donald Yost, Budd Wheel Co., Detroit, Mich.

A discussion of "Practical Problems of Gray Iron Sand Control" was lead by R. E. Wilke, John Deere Tractor Co., Waterloo, Iowa.

MALLEABLE CAST IRON

Wednesday, May 18, 9:00 A. M.

Chairman—Dr. H. A. Schwartz, National Malleable & Steel Castings Co., Cleveland, O.

Vice Chairman—Homer Wright, Michigan Malleable Iron Co., Detroit, Mich.

The following papers were presented and discussed:

Spheroidized Pearlitic Malleable Cast Iron, by Duncan Forbes, Gunite Foundries Corp., Rockford, Ill.

Production of Short Cycle Malleable Iron, by W. D. McMillan, International Harvester Co., Chicago, Ill.

Pearlitic Malleable—Tocco Hardened, by W. E. Benninghoff, Ohio Crankshaft Co., Cleveland, O.

STEEL CASTINGS

Wednesday, May 18, 9:00 A. M.

Chairman—F. A. Melmoth, Detroit Steel Castings Co., Detroit, Mich.

The following report and papers were presented and discussed:

Report of Committee on Coupon Tests, presented by R. A. Gezelius Taylor Wharton Iron & Steel Co., High Bridge, N. J.

The Effect of Deoxidation Treatments on the Ductility of Cast Steel, by A. P. Gagnebin, International Nickel Co., Bayonne, N. J.

Steel Casting Design for the Engineer and the Foundryman, by C. W. Briggs, A. R. Donaldson, Naval Research Laboratories, Anacostia, Washington, D. C. and R. A. Gezelius, Taylor-Wharton Iron and Steel Co., High Bridge, N. J. This paper was presented by C. W. Briggs.

SAND RESEARCH
Wednesday, May 18, 9:00 A. M.

Chairman—R. F. Harrington, Hunt-Spiller Mfg. Co., Boston, Mass.

Vice Chairman—Harold W. Johnson, Greenlee Foundry Co., Chicago, Ill.

The following papers and committee report were presented and discussed:

Effects of Different Electrolytes on A.F.A. Clay Determinations, by J. F. Haseman, Cornell University, Ithaca, N. Y. In the absence of the author, Dr. H. Ries, Cornell University, presented this paper.

What Happens to a Core in a Core Oven, by Jasper Willsea, The Willsea Works, Rochester, N. Y.

Modern Automotive Foundry Sand Practice, by A. S. Nichols, Illinois Clay Products Co., Chicago, Ill.

Report of Sand Research Committee, presented by Dr. H. Ries, director and chairman of the Committee on Sand Research.

APPRENTICE TRAINING DEMONSTRATION
May 18, 9:00 A. M.

A small but highly interested group met at the Cleveland Trade School. They were greeted by J. G. Goldie, foundry instructor, and Frank Cech, pattern shop instructor. After a tour of the building, they gathered in the related instruction room of the pattern shop, where Mr. Goldie and Mr. Cech outlined their instructional methods and co-ordination between their departments. Problems, both common to such a school and peculiar to their own situation, were discussed. Participation in this discussion, by the group, was very active.

The meeting was then moved to the foundry where a class was at work. After some minutes, the class was called into the related instruction room provided for foundry boys and was quizzed at some length by the instructor and the assembled guests. The quizzing was conducted by drawing, blindly, from a pack of cards a card bearing a question on foundry practice or material used.

From a similar stack of cards, the name of a boy was drawn and he was then required to answer the question. The performance of these lads under very unusual conditions was excellent, and a tribute to the instructional effort expended on them. The questions dealt with molding and core making practices, melting, calculating heats, furnace care and operation, cleaning room methods and foundry materials.

The school is operated, both for vocational instruction and trade finding purposes, and for the giving of related instruction to employed apprentices. Their primary purpose is to supplement apprenticeship and no pretense is made at teaching a trade or supplanting proper industrial training. In that sense, the name "Cleveland Trade School" is somewhat a misnomer.

In addition to the two departments discussed herein, the school operates a full automotive department and a building trades department

including carpentry, brick and masonry, plastering, plumbing and interior decorating.

The equipment in both the pattern shop and the foundry is quite complete and is adapted to the production of a much heavier line of work than is the rule in school shops. While funds for the carrying on of some of the work are very meager, Mr. Cech and Mr. Goldie enjoy the cooperation of a number of Cleveland metal trades employers and it must be said for them that they are doing an excellent piece of work with what they have.

ANNUAL BUSINESS MEETING

Wednesday, May 18, 1938, 11:00 A. M.

Presiding—President Hyman Bornstein.

On calling the meeting to order, President Bornstein announced as the first order of business the report of the Nominating Committee, which was made by Secretary D. M. Avey. The report was as follows: "Nominations made by your committee are:

For president to serve for one year:
> Marshall Post, Birdsboro Steel Foundry and Machine Co., Birdsboro, Pa.

For vice-president to serve for one year:
> Henry S. Washburn, Plainville Casting Co., Plainville, Conn.

For directors to serve three years each:
> Hyman Bornstein, Deere & Co., Moline, Ill.
> Frederick Ayres Lorenz, Jr., American Steel Foundries, Chicago, Ill.
> D. O. Thomas, Bendix Aviation Corp., South Bend, Ind.
> George A. Seyler, Lunkenheimer Co., Cincinnati, O.
> H. S. Hersey, C. O. Bartlett & Snow Co., Cleveland, O."

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Respectfully submitted,
1938 NOMINATING COMMITTEE

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Past-President—Frank J. Lanahan Fort Pitt Malleable Iron Co., Pittsburgh, Pa., *Chairman.*

Past-President—James L. Wick, Jr., Falcon Bronze Co., Youngstown, O.

Past-President—T. S. Hammond, Whiting Corp., Harvey, Ill.
W. M. Ball, Jr., Edna Brass Mfg. Co., Cincinnati, O.
R. R. Deas, American Cast Iron Pipe Co., Birmingham, Ala.
W. H. Doerfner, Saginaw Malleable Iron Div., General Motors Corp., Saginaw, Mich.
K. V. Wheeler, Lebanon Steel Foundry Co., Lebanon, Pa.

Secretary Avey moved "that the report of the Nominating Committee be accepted and that the Secretary be instructed to cast the unanimous ballot of the Association in favor of the nominees."

Being duly seconded, the motion was approved and President Bornstein declared the above named candidates elected to the offices named. He then introduced President-Elect Post, requesting him to assume the chair.

At this time, Secretary Avey presented a recommendation of the Board of Directors that President Bornstein and F. A. Lorenz, Jr., 1938 Medallist, be elected to Honorary Life membership. Mr. Avey moved that the recommendation be adopted, which motion on being seconded was unanimously approved.

President Bornstein then resumed the chair and called for the report of the committee to nominate for election four members to serve with the 1939 Nominating Committee. This report was presented by Fred J. Walls, International Nickel Co., Detroit, chairman of the committee. The report recommended the following as nominees and alternates:

> Alfred L. Boegehold, General Motors Research Laboratories, Detroit, Mich.
> *Alternate:* R. R. Deas, Jr., American Cast Iron Pipe Co., Birmingham, Ala.
>
> L. H. Rudesill, Griffin Wheel Co., Chicago, Ill.
> *Alternate:* Robert Gregg, Reliance Regulator Co., Los Angeles, Calif.
>
> Sam Tour, Lucius Pitkin, Inc., New York, N. Y.
> *Alternate:* W. C. Hartmann, Bethlehem Steel Co., Bethlehem, Pa.
>
> Walton L. Woody, National Malleable and Steel Casting Co., Sharon, Pa.
> *Alternate:* Joseph Sully, Sully Brass Co., Toronto, Ont., Canada.

> The report, in addition to Chairman Walls, was signed by the following members:
>
> H. W. Johnson, Greenlee Foundry Co., Chicago, Ill.
>
> R. J. Allen, Worthington Pump and Machinery Corp., Harrison, N. J.

A motion to approve this report was seconded and carried.

R. E. Kennedy, Technical Secretary, then announced the winners in the Annual Apprentice Molding and Pattern Making Contests. These were as follows:

Patternmaking Contest

First Prize, John Bloomquist, John Deere Harvester Works, East Moline, Ill.

Second Prize, Edward A. Menzel, Brown & Sharpe Mfg. Co., Providence, R. I.

Third Prize, William Tharp, Caterpillar Tractor Co., Peoria, Ill.

Non-Ferrous Molding Contest

 First Prize, Joseph Wyban, Art in Bronze Foundry, Cleveland, O.
 Second Prize, Harry Smith, Mississippi Foundry Corp., Rock Island, Ill.
 Third Prize, Casimir Kotowicz, Ampco Metal Corp., Milwaukee, Wis.

Steel Molding Contest

 First Prize, Richard Reynolds, West Steel Castings Co., Cleveland, O.
 Second Prize, John Paliwoda, West Steel Castings Co., Cleveland, O.
 Third Prize, Jalmer Nelson, American Steel Foundries, East Chicago, Ind.

Gray Iron Molding Contest

 First Prize, Chester Marzec, Cleveland Trade School, Cleveland, O.
 Second Prize, Robert Pohlman, Carondelet Foundry Co., St. Louis, Mo.
 Third Prize, Jerry Halek, Johnston & Jennings Co., Cleveland, O.

President Bornstein then called upon the chairman of the International Relations Committee, Frank G. Steinebach, The Foundry, Cleveland, O., who read cablegrams of greetings from two overseas foundry associations. These were:

> "The French Foundrymen's Association addresses its cordial greetings to its American colleagues. Best wishes for a successful convention."
>
>   *Signed* PRESIDENT SANDRE.

> "Institute of British Foundrymen send cordial greetings to American Foundrymen's Association and sincere wishes for a successful convention. Hope to welcome many American friends to International Foundrymen's Congress in London in 1939."
>
>   *Signed* C. W. BIGG, President
>   TOM MAKEMSON, Secretary.

Replies forwarded were:

> "American Foundrymen's Association appreciate greatly British and French greetings and reciprocate good wishes. Convention most successful."
>
>   *Signed* F. G. STEINEBACH
>   *Chairman,* International Relations

At this time Past-President Frank J. Lanahan, on behalf of the Board of Awards announced the first annual Board of Awards Lecture and introduced the lecturer, Charles R. Hook, president of the American Rolling Mills Co., Middletown, O. and president of the National Association of Manufacturers. Mr. Hook in his lecture discussed "Some Management Opportunities and Responsibilities." This address was later published by the Association.

STEEL DIVISION LUNCHEON AND ROUND TABLE CONFERENCE

Wednesday, May 18, 1:00 P. M.

Chairman—Ralph West, West Steel Casting Co., Cleveland, O.

Vice-Chairman—A. H. Jameson, Malleable Iron Fittings Co., Branford, Conn.

At this meeting the subjects discussed were "Designing Steel Castings—Converter Steel Casting—Cement Mold Process."

CAST IRON SHOP COURSE (SESSION 3)

Wednesday, May 18, 4:00 P. M.

Chairman—K. H. Priestley, Eaton-Erb Foundry Co., Vassar, Mich.

The following paper was presented and discussed at this meeting.

Gates and Risers for High Test and Alloyed Cast Iron, by C. C. Brisbois and A. E. Cartwright, Robert Mitchell Co., Ltd., Montreal, Canada. Mr. Cartwright presented this paper.

MALLEABLE SHOP COURSE (SESSION 1)

Wednesday, May 18, 4:00 P. M.

Chairman—D. I. Dobson, General Malleable Corp., Waukesha, Wis.

"Practical Metallurgical Problems," was the subject for discussion at this meeting, lead by J. H. Lansing, Malleable Founders' Society, Cleveland, O., and assisted by A. M. Fulton, Northern Malleable Iron Co., St. Paul, Minn.

ANNUAL BANQUET

Wednesday, May 18, 1938, Hotel Cleveland

President H. Bornstein *presiding.*

President Bornstein, in opening the proceedings of the annual banquet, announced as the first order of business the presentation of the Joseph S. Seaman Gold Medal of the Association. He informed the meeting that due to a serious illness, Frederick A. Lorenz, Jr., who had been awarded the Joseph S. Seaman Medal, was unable to be present, but that Albert Walcher, vice-president of the American Steel Foundries, an associate of Mr. Lorenz, would respond in Mr. Lorenz's behalf. He then introduced Past-President James L. Wick, Jr., chairman of the Board of Awards.

PAST PRESIDENT WICK: *President Bornstein, honored guests, ladies and gentlemen:*

We are gathered here tonight to give testimony to our faith in a great industry, to rededicate ourselves to the high purposes and the lofty ideals of the American Foundrymen's Association, and to pay tribute to the officers, committeemen and those who have given so unstintingly of their time, thought and effort to translate into reality the exalted objectives of our Association.

It is my pleasant privilege at this time to preside over a ceremony that was sponsored in 1921 by four devoted members who are so reverently remembered for their notable contributions revitalizing the activities of our Association and enhancing the welfare of our industry in general, who originally conceived the thought of stimulating a more intense interest in the solution of the problems at hand and establishing a means by which conspicuous performance by reaching that end might be recognized in a tangible way through the award of a Medal of Merit. These four men have given ample proof of their sublime devotion to this great cause and through this Award continue to touch and influence the lives of all of us.

To carry out the idea that they had in mind, it becomes the duty of th. Board of Awards to select a person or persons upon whom they may bestow this signal honor, an Award for Meritorious Service in the Advancement of Our Industry provided from the proceeds of an endowment fund which these four created. The four to whom I refer and the qualities for which they will always be remembered and revered, are: Joseph S. Seaman, for his kindliness; John A. Penton, for his vision; John H. Whiting, for his industry; and William H. McFadden, for his outstanding achievements.

So tonight, with humble gratefulness in our hearts for the things that they did in an era gone by, we through the circumstances which they created join with them in honoring one of our associates of the present.

It was on a Friday noon, June 25, 1933, to be exact, that I first observed Frederick Ayres Lorenz, Jr., in action. A coalition meeting of the foundry industry-at-large had been convened. The regimentation of the entire foundry industry under groups representing the steel, malleable, gray iron and non-ferrous divisions had gathered to swing forward in full cooperation under the National Industrial Recovery Act. Mr. Lorenz was invited to preside. And I was greatly impressed by the manner in which he conducted the meeting. Shortly thereafter he was elected president of the Steel Founders' Society of America and has continued in that office to the present day. He guided the steel casting industry throughout the NRA code period with rare judgment and vision. At the same time he served as a member of the Durable Goods Committee. He has carried on in an even more effective manner since. Because of his inherent talent to inspire a spirit of cooperation to the end of establishing the observance of principles of fair competition, the steel division of the foundry industry has emerged solidly entrenched in a healthier commercial position than any other industry, though he blazed the trail and made available to others that by which they in a similar manner could have improved their individual economic condition.

In the past emphasis seems to me to have been placed by the Board of Awards upon recognition of the service of those who have made notable contributions to the expansion of this great industry with respect to its technical aspects only.

Tonight we rejoice in being able to honor one who to an unusual degree represents a combination of all. From a technical standpoint, he

has presented papers before the American Foundrymen's Association embracing the subject of "Design of and Efficient Grinding Methods for Steel Castings." He prepared a section on Steel Casting Design and it was included in the symposium on Steel Castings published by the American Foundrymen's Association in collaboration with the American Society for Testing Materials. More recently he developed a thesis entitled, "An Engineering Approach to the Use of Steel Castings," that was presented before a meeting of the American Society of Mechanical Engineers, the Machine Shop Practice Division. He has been a member and chairman of the Executive Committee of the Steel Division of the American Foundrymen's Association.

Had Mr. Lorenz's activities been restricted to and confined within the sphere of technical eras, he undoubtedly would have loomed large in that horizon. But those with whom he was associated fortunately detected his genius for organization and insofar as the foundry industry is concerned, therein lies his greatest contribution.

To say that Mr. Lorenz has done an outstanding job would be putting it mildly. By all odds, from a non-technical standpoint at least, it is the best that has ever been performed in this industry. His accomplishments epitomize the motto of the American Foundrymen's Association: *Coming together is a beginning; working together is progress; keeping together is success.*

Mr. Lorenz was born in Lansing, Michigan, the son of Frederick A. and Florence Ayres Lorenz. He was educated at the University of Chicago and the University of Illinois, from which he received the degree of bachelor of science. He began his business and professional career with the Chicago & North Western Railroad as inspector engineer. From there, he went to the Republic Iron & Steel Company at East Chicago, Indiana, as plant engineer, and later became superintendent of the rolling mills. In 1911 he became connected with the American Steel Foundries in Chicago, and today is one of its vice-presidents. He is a lieutenant-colonel in the ordinance reserve of the United States army.

Frederick Ayres Lorenz, Jr., your native abilties have been so clearly demonstrated that they stand out like a clear-cut cameo and your individual services in the cause are recognized by every member of the steel casting fraternity. Such a splendid example have you set for the other branches of our industry that in admiration thereof and upon the recommendation of the Board of Awards of the American Foundrymen's Association and on their behalf and on that of the donor, I have the distinct pleasure of advising you that you have been awarded the Joseph S. Seaman Medal for 1938. And I take pleasure in presenting a bronze replica to your representative, Mr. Alfred Walcher.

May I read again the citation on the Certificate of Award: "Frederick A. Lorenz, Jr., for courageously espousing and valiantly bringing to fruition, a broad conception of the commercial equation in the foundry industry."

Mr. Walcher, it gives me great pleasure to present this to you who represent our esteemed friend and associate.

MR. WALCHER: President Bornstein, ladies and gentlemen:

It is, of course, not possible for me to speak for Lorenz, but the awarding of this medal by the American Foundrymen's Association to Frederick Lorenz is sufficient evidence of his integrity and ability to serve the industry over a long period of time. It is indeed unfortunate he is not able to be with us here this evening. I think it would have been the happiest moment of his life if he could have been here with us.

I take very great pleasure in accepting this honor which has been conferred upon him, knowing he will appreciate it with very deep gratitude. Thank you.

Following the presentation of the Medal Award, President Bornstein introduced the speaker of the evening, H. V. Kaltenborn, news editor, Columbia Broadcasting Co., New York, who gave an exceptionally interesting review of current world events and trends in the internal situation.

Following Mr. Kaltenborn's address, the meeting was adjourned.

SAND SHOP COURSE (SESSION 3)

Thursday, May 19, 8:30 A. M.

Chairman—P. Charles Fuerst, Falk Corp., Milwaukee, Wis.

At this session a discussion of the "Practical Problems of Core Room Sand Control" was led by H. L. Campbell, American Hoist and Derrick Co., St. Paul, Minn.

MALLEABLE CAST IRON

Thursday, May 19, 10:00 A. M.

Chairman—L. N. Shannon, Stockham Pipe Fittings Co., Birmingham, Ala.

The following papers were presented and discussed:

Factors Influencing Annealing Malleable Iron, by A. L. Boegehold, General Motors Research Laboratory, Detroit, Mich.

Atmosphere Control in Annealing Malleable Iron, by E. G. deCoriolis and R. J. Cowan, Surface Combustion Corp., Toledo, O. Mr. deCoriolis presented this paper.

The Design of Conveyor Equipment for Malleablizing Furnaces, by H. H. Harris, General Alloys Co., Boston, Mass.

Discussion of these papers was continued at the Malleable Luncheon and Round Table Conference, which convened immediately following this session.

MALLEABLE DIVISION LUNCHEON AND ROUND TABLE CONFERENCE

Thursday, May 19, 12:30 P. M.

Chairman—J. H. Lansing, Malleable Founders' Society, Cleveland, O.

"Annealing Malleable Iron" was the subject discussed at this meeting.

MATERIALS HANDLING

Thursday, May 19, 10:00 A. M.

Chairman—E. W. Beach, Campbell, Wyant & Cannon Foundry Co., Muskegon, Mich.

Vice-Chairman—Jas. Thomson, Continental Roll & Steel Foundry Co., East Chicago, Ind.

The following papers were presented and discussed:

Foundry Maintenance, by Carter Bliss, Scullin Steel Co., St. Louis, Mo.

Tramrail Distribution of Core Sands and Cores, by A. F. Anjeskey, Cleveland Tramrail Div., Cleveland Crane & Engineering Co., Cleveland, O.

Conveyors in a Modern Foundry, by J. B. Silver, Palmer-Bee Co., Detroit, Mich.

MANAGEMENT — FOUNDRY SAFETY AND HYGIENE

Thursday, May 19, 2:00 P. M.

Chairman—E. H. Ballard, General Electric Co., West Lynn, Mass.

The following paper was presented and discussed:

Silicosis and the Foundry Industry, by Dr. Leonard Greenburg, Executive Director, Div. of Industrial Hygiene, Dept. of Labor, State of New York. The discussion was under the leadership of Dr. J. H. Chivers, Crane Co., Chicago, Ill.

CAST IRON SHOP COURSE (SESSION 4)

Thursday, May 19, 4:00 P. M.

Chairman—Harry Rayner, Dodge Bros. Corp., Detroit, Mich.

At this meeting "Cupola Operation" was discussed, G. P. Phillips of the International Harvester Co., Chicago, leading the discussion.

MALLEABLE SHOP COURSE (SESSION 2)

Thursday, May 19, 4:00 P. M.

Chairman—D. I. Dobson, General Malleable Corp., Waukesha, Wis.

The following paper was presented and discussed:

Design of Straightening Equipment for Malleable Iron Castings, by C. W. Weedfall, Saginaw Malleable Iron Division, General Motors Corporation, Saginaw, Mich.

A discussion of *Shop Practice Problems* and a discussion of *Current Shop Problems* followed the presentation of the above paper, under the leadership of Ed. Meyer, Chain Belt Co., Milwaukee, Wis.

Annual Report of Executive Vice President

Members of the American Foundrymen's Association:

The following summary of general Association activities for the year ending June, 30, 1938, is submitted for the records and as a supplement to special reports and the official acts of the Board of Directors and Executive Committee as presented in the minutes of their meetings and as shown in this volume.

Membership

The book membership as of June 30, 1937, was 2,452; new members elected, 603; number of delinquents, 44; removed by death 4; resigned, 54—total, 102; net gain, 501. (Greatest membership gain in any previous year was 530 in the year 1920.)

The book membership as of June 30, 1938, was 2,953. (Previous high was 2,477 on December 31, 1930.) One hundred and twenty-two members are located outside of the United States and Canada, distributed over 22 countries.

Chapters

During the year, two new chapters were formed—the Metropolitan New York-New Jersey Chapter and the Northern Illinois-Southern Wisconsin Chapter, with headquarters at Rockford, Illinois. A third chapter was formed by members of our Canadian Section, residing in Ontario. This is known as the Ontario Chapter with headquarters at Toronto. We still have a Canadian Section comprising members located outside of Ontario.

For the records we show below name and location of chapters and number of members of all classes in each as of June 30:

Birmingham District	211	Northern California	62
Buffalo	119	Northern Illinois-	
Chicago	337	Southern Wis.	35
Detroit	170	Quad City	91
Met. New York-		St. Louis District	95
New Jersey	107	Southern California	119
Met. Philadelphia	181	Wisconsin Chapter	175
Northeastern Ohio	234	Canadian Section	113

A large measure of credit is due chapter officers and membership committees for their promotion work in increasing membership.

Finances

The books of the Treasurer and the Association were audited by Robert T. Pritchard, C.P.A. On pages xxv, xxvi appears the balance sheet and statement of income and expense. For the purpose of comparison, the balance sheet for the preceding year is also shown. It will be seen that the assets of the Association, exclusive of award funds, total $70,392.86 as of June 30, 1938, compared with $59,927.55 at the end of the preceding fiscal year. In the income and expense statement excess of income over expense is given as $11,432.94. Of this amount $7,123.49 is in cash and the balance in inventory.

Award funds covering the McFadden, Seaman, Penton, Whiting and Obermayer funds of which the principal funds are invested, total $21,520.10, with a balance of $3,233.36 as of June 30 in the award fund interest account.

The complete report of the auditor presented by the Finance Committee at the annual meeting of the Board of Directors, July 26, is on file at the office of the Association subject to examination by any member interested in seeing it.

In considering the financial status of the Association it must be borne in mind that the books were closed and audited less than six weeks after the close of a successful exhibit and the further fact that operations will be carried on for the year 1938-39 without the aid of income from exhibits as none will be held in conjunction with the 1939 annual convention. To meet this situation, economies in operation will be necessary and a determined effort made to increase income from dues of members, sale of publications and in other ways. This will be reflected in the 1938-39 operating budget.

Regional Meetings

Regional foundry conferences, usually of two- or three-day duration, are playing an increasingly important part in the Association's activities. Held generally in cooperation with engineering schools and under the sponsorship of the Association and chapters, they are bringing about closer understanding between the engineering schools and the foundry industry, and are in other ways fostering the education of those attending. These meetings are covered more fully in the report of the Technical Secretary.

Safety and Hygiene

The work of our very important section on Safety and Hygiene, is covered elsewhere in the report of our Director of Safety and Hygiene, E. O. Jones.

Amendment to By-Laws

On the recommendation of a special Fact-Finding and Policy Committee unanimously approved by the Board of Directors, amendments to the By-Laws providing changes in membership classification and in rates of dues were submitted to the membership for letter ballot on April 23, 1938 and approved by them to become effective as of July 1,

1938 for all new members with no changes in rates of dues for members of record June 30, 1938, until July 1, 1939.

The amendment provided for a new class of membership known as Sustaining members with minimum annual dues of $50.00. It is expected that this provision will considerably increase income for support of Association activities, providing as it does for the larger operators to subscribe in proportion to tonnage produced and benefits received.

1938 Foundry Show

From the report of the Manager of Exhibits presented at the annual meeting of the Board of Directors, the following summary of facts is presented for the records:

The total space sold was 58,835 sq. ft. as compared to 50,732 sq. ft. at Milwaukee in 1937 and 55,741 sq. ft. at Detroit in 1936. The total number of exhibitors at Cleveland was 209 as compared to 209 in Milwaukee and 214 at Detroit.

One hundred and twenty-three, or 58 per cent, of the exhibitors at the Cleveland Show also exhibited at Milwaukee and Detroit. The balance, or 42 per cent, was about the average of floating exhibitors, that is, those who exhibit when the convention is staged in their home city or in a district which they serve.

While the Cleveland Show was not the largest Foundry Show ever staged, it was practically the unanimous opinion of convention goers that it was the best ever, with a greater amount of new equipment and new developments displayed than at any previous show.

Greater Service

With increased membership, with chapters extending from coast to coast operating effectively in various ways to further the program of A.F.A., with steadily increasing contacts in engineering fields offering opportunities for presenting the utility of cast metals, with a new publication policy approved and in effect, with an efficient headquarters staff and hundreds of capable and willing committee workers, we look forward to rendering, with the helpful cooperation of other organizations, greater service to the foundry industry of America.

Respectfully submitted,

C. E. HOYT, *Executive Vice President*

American Foundrymen's Association, Inc.
Balance Sheet

AS AT JUNE 30, 1938

Assets
ASSOCIATION ASSETS

	JUNE 30, 1938	JUNE 30, 1937
Cash in Bank, Schedule III	$45,646.86	$38,523.37
Accounts Receivable	916.00	1,668.78
Supplies on Hand	4,301.72	1,850.35
Furniture and Fixtures, less Depreciation Reserve	1,874.98	1,597.47
Due from Award Funds	684.00	
Total Association Assets	$53,423.56	$43,639.97

RESERVE FUND — Schedule II

Investments at Cost	$14,865.55	$14,865.55
Cash in Bank on Savings Account	1,776.07	1,099.19
Total (Market Value $14,892.00)	$16,641.62	$15,964.74

STEEL CASTINGS TEST FUND

Cash in Bank on Savings Account	$ 327.68	$ 322.84
Grand Total	$70,392.86	$59,927.55

Liabilities
ASSOCIATION LIABILITIES

Accounts Payable	$ 50.00	
Accrued Social Security Taxes	522.48	
Dues Paid in Advance	2,532.74	$ 3,147.45
Unexpected Appropriation		381.07
Total Association Liabilities	3,105.22	3,528.52
Surplus — Exhibit C	50,318.34	40,111.45
	$53,423.56	$43,639.97

RESERVE FUND — Schedule II

Principal of the Fund	$16,641.62	$15,964.74
	$16,641.62	$15,964.74

STEEL CASTINGS TEST FUND

Unexpended Appropriation	$ 327.68	$ 322.84
Grand Total	$70,392.86	$59,927.55

American Foundrymen's Association, Inc.
Income and Expenses

FOR THE YEAR ENDED JUNE 30, 1938

INCOME		This Year	Last Year
Dues		$ 32,445.10	$26,712.01
Publications — Subscriptions and Sales		4,436.70	3,202.72
Exhibit Permits		5,200.00	5,275.00
Space Rentals		64,927.13	56,941.24
Registration Fees		3,466.50	3,835.50
Cast Metals Handbook, receipts		544.64	1,132.95
Sand Book — Advertising	$2,654.40		
Sales	65.24	2,719.64	
American Founders Compensation Group		2,945.01	2,124.71
Safety and Hygiene Service		955.83	
Total Income		**$117,640.55**	**$99,224.13**

EXPENSES	This Year	Last Year
Salaries — Officers and Assistants	36,172.25	29,130.00
Stenographic and Clerical	7,457.46	6,251.73
Committee Traveling and Expense	2,013.45	1,993.32
Publications	13,118.05	12,947.80
Printing	4,242.86	3,638.33
Postage	4,101.59	3,381.03
Office Rent	3,675.00	3,350.00
Telephone and Telegraph	872.39	926.73
Office Expense	1,313.70	784.13
Taxes	50.87	10.72
Traveling Expense	2,222.91	1,329.70
Insurance	208.33	264.95
Rental of Building for Exhibit	7,055.95	3,150.00
Exhibit Installation and Expense	874.50	2,052.38
Booth Work	2,678.93	2,675.35
Power for Exhibit — Net	311.98	392.66
Advertising and Promotion (Exhibit)	5,588.92	5,393.49
Dues — Other Associations, etc.	60.00	77.00
Legal and Professional Services	252.00	100.00
Convention Expense	1,387.79	1,618.58
Banquet Expense — Net	105.79	51.81
Registration Expense	239.69	268.10
District Chapters — Prorata of Dues	3,812.88	2,526.87
Cast Metals Handbook — Expenses		376.73
Sand Book — Cost	2,750.14	
Sand Research	765.75	
Safety and Hygiene Service	656.89	617.44
Social Security Taxes	1,490.44	
Depreciation of Furniture and Fixtures	752.45	742.12
Exchange	146.30	131.91
General Expense	161.92	132.25
Chapter Meetings, etc.	1,666.43	691.72
Total Expenses	**$106,207.61**	**$85,006.85**

EXCESS OF INCOME OVER EXPENSES FOR THE YEAR.....$ 11,432.94 $14,217.28

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Report of Director,
Safety and Hygiene Section

Members of Board of Directors, American Foundrymen's Association:

We are pleased to report many of our problems which six years ago seemed like mountains unsurmountable, have been ironed out into mole hills. Many States, during this period of time, have amended or passed new legislation which has helped to clarify some of the difficulty. The present problem in some instances, consists of smoothing out the wrinkles which will be accomplished with the passing of time. Your Association, during this period, has cooperated with National, State, Employer and Employee representatives which has resulted in a clearer understanding of our safety and hygiene problems.

The Industrial Hygiene Codes Committee headed by J. R. Allan as chairman has completed the following three codes:

> Tentative Code of Recommended Practices for Testing and Measuring Air Flow in Exhaust Systems,
>
> Tentative Code of Recommended Practices for Grinding, Polishing and Buffing Equipment Sanitation,
>
> Tentative Recommended Good Practice Code and Handbook on the Fundamentals of Design, Construction, Operation and Maintenance of Exhaust Systems,

all of which have been published within the last two years. These codes have been well received by the foundry and allied industries and have filled a much needed want.

This Committee is, at the present time, working on a Metals Cleaning Code which will be published some time after the first of the year.

It is difficult for anyone to appreciate just how helpful this material is unless they have been called upon to serve on committees in their own States which were presented with the problem of setting up standards as far as hygiene in the foundry is concerned.

Too many times industry has been unprepared and has had to accept what other interests were well prepared with. In October, 1936 when the Illinois Occupational Diseases Act became effective, our industry found itself unable to purchase coverage at any price inasmuch as very few carriers wished to underwrite the business. A group of foundrymen sought assistance from the Association and the Board of Directors finally consented to have the Association administer a cooperative insurance plan which had been submitted to them, the Association to be reimbursed for the time spent in this activity. This group has been in existence for two years and has increased its membership steadily.

During the Annual Convention we were fortunate in having Dr. Leonard Greenburg, Executive Director, Division of Industrial Hygiene, New York State Department of Labor present a paper entitled: "Silicosis in the Foundry Industry" which is indicative of the fact that our particular problems have been considerably overestimated. As an industry, however, we will continue the good work which has been accomplished.

We would like to pay our respects and extend our thanks to those who have been on the firing line giving generously of their time, wisdom and money. They, no doubt, have held the umbrella over the heads of many others and the industry can be thankful they had the courage and intestinal fortitude to do so.

Respectfully submitted,

E. O. JONES, *Director of Safety and Hygiene Section.*

Report of Technical Secretary

CHICAGO, JULY 26, 1938.

Members of Board of Directors, American Foundrymen's Association:

Your Technical Department has functioned this past year largely in carrying out the following major activities: (1) Preparation of program for 42nd Annual Convention, working with committees formulating the program for the various sessions; (2) Editing and publication of: (a) Bound Volumes of Transactions No. 45 (1937), containing Proceedings of 41st Annual Convention, (b) Bimonthly Transactions (6 issues), containing Association news and convention proceedings, (c) Management series of reprints of Milwaukee convention, (d) Special publications, (e) Preprints for Cleveland convention — 34 preprints issued, (f) Fourth edition of book on Testing and Grading Foundry Sands, (g) Alloy Cast Iron publication; (3) Technical committee activities; (4) A.F.A. fall technical conference; (5) Regional foundry meetings; (6) Chapter program development.

Forty-Second Annual Convention

While we said last year that the Milwaukee convention sessions marked a record in extent of interest and quality of papers, we thoroughly believe that the Cleveland convention papers and reports so far exceeded those of the Milwaukee convention that it will be difficult to develop another convention program of similar quality and interest. This last program was outstanding in balance of interest between the management, metallurgical and practical phases. This increase in quality of papers, we believe, is due to the increasing reputation of technical work of the Association, and general improvement in the ability of foundry workers to prepare good papers.

Annual Convention Sessions

A comparison of the number and types of sessions for the past three years shows a gradual increase in number — 25 in 1936, 28 in 1937 and 33 in 1938 — the increase for 1938 over 1937 being accounted for by the added sessions of the Shop Operation Courses. Another reason for the gradual increase is that we have endeavored to have fewer papers at any one session, this making for better presentation and more time for discussion.

An innovation in this year's schedule was the establishing of a published set time for each paper. This met with an excellent reception and it is proposed to follow this scheme as a regular policy. In addition to the time schedule, another feature to be credited for the better handling of the sessions was the revised set of rules and regulations for the conduct of sessions and presentation of papers which was furnished to each chairman and author.

xxx FORTY-SECOND ANNUAL MEETING

Number of Papers — Annual Convention

While there has been an increase in the number of sessions at the Annual Convention, the total number of papers has for several years remained about the same. While it is rather surprising to note that only three more papers and reports were presented in 1938 than in 1937, there can be no question but what the papers for 1938 were of an unusually high grade and of outstanding interest in all classes, the practical as well as the management and technical. For this, the greatest credit is due to the chairmen and members of the various program committees.

Instructors Dinner

The annual dinner gathering of engineering instructors during convention week is proving increasingly popular. It is believed that this meeting is doing much to raise the standard of foundry instruction in engineering schools and promoting the interest of engineering instructors in spreading information on casting uses amongst engineering students.

Shop Instruction Courses

This year three shop courses were held — Gray Iron (4 sessions), Sand Control (3 sessions) and Malleable (2 sessions). The Malleable Course was the first ever held by that division and its development was due to the efforts of D. I. Dobson, General Malleable Corporation, Waukesha, Wis. Some shop courses have been held each year since they were started in 1928.

Round Table Luncheons

Each division — Gray Iron, Malleable, Non-Ferrous and Steel — held a round table luncheon conference at the Cleveland convention. The chief difficulty this year in holding these luncheons was the lack of adequate restaurant service. We recommend that next year these meetings be held where good meeting rooms and service are available.

Fall Technical Conferences

The first Fall Technical Conference of the Association was held September 30, October 1, 1937, at Battelle Memorial Institute, Columbus, O. With an attendance of approximately 175 from all sections of the country, and with the discussions limited to a few outstanding problems of the gray iron, steel and non-ferrous foundries, the meeting was voted a very successful affair. At the time of this conference, all divisions held meetings of their executive committees. The various division committees voted approval of a similar conference for the fall of 1938. An invitation to hold this year's meeting at Ann Arbor, Michigan, was received from the University of Michigan and accepted by the Board of Directors. The future of this conference as an annual affair, should, we believe, be left to a vote of the divisions involved, with their recommendation being the guide for your approval.

Regional Foundry Conferences

This past year has witnessed a decided increase in the numbers of regional conferences as sponsored by chapters. During the previous year four were held, while this past year seven have been held. Those for the past year were:

Rolla, Mo. (Missouri School of Mines) Sponsored by St. Louis Chapter.

Iowa City, Ia. (University of Iowa) Sponsored by Quad City Chapter and Northern Illinois Foundrymen's Association.

Milwaukee, Wis. Sponsored by Milwaukee Chapter with the University of Wisconsin cooperating.

Ithaca, N. Y. (Cornell University) Sponsored by A.F.A. members in Northwestern New York and Buffalo Chapter.

Birmingham, Ala. Sponsored by Birmingham Chapter, A.F.A., and Birmingham Section of American Society of Mechanical Engineers.

East Lansing, Mich. (Michigan State College) Sponsored by Detroit Chapter.

Cambridge, Mass. (Massachusetts Institute of Technology) Sponsored by New England Foundrymen's Association with A.F.A. cooperating.

As an evidence of the popularity of regional meetings, it can be said that of those held in past years, the attendance this year increased in every case, the largest attendance being that at Birmingham with over 600 registered. The combined attendance at all the regional meetings was over 2200.

It is our recommendation that these meetings be encouraged, with the limitation that each group endeavor to secure their speakers from their own district, and to not expect attendance from outside their own districts. This recommendation is made because we feel that these meetings build up chapter interest and membership and bring practical information to many who do not have the opportunity to attend the annual A.F.A. conventions as well as developing talent and papers for our annual conventions.

Cast Metals Handbook

The Cast Metals Handbook, first edition (1935) is now under revision. All copies of the first edition have been distributed either through sales, or to members. Our schedule calls for the second edition to be off the press this winter and its preparation will be a major activity of the technical department until issued.

Exchange Papers

During the past year, three official exchange papers have been prepared by A.F.A. members for presentation before foreign associations: Dr. Jas. T. MacKenzie, American Cast Iron Pipe Co., Birmingham, Ala., for the Institute of British Foundrymen, James R. Allan, International Harvester Co., Chicago, for the French Foundry Association, and that for the International Foundry Conference at Warsaw, Poland, was prepared by W. W. Kosicki, Metalloy Products Co., and G. K. Eggleston, Detroit Lubricator Co., Detroit. Papers from foreign associations to our Cleveland convention were prepared by Professor W. J. Rees, Sheffield University, on behalf of the I.B.F., and by M. Dudouet, on behalf of the French Foundry Technical Association. These bring to 88 the number of papers presented by this exchange arrangement.

General Interest Committees

International Relations Committee

The International Relations Committee, under the chairmanship of Frank G. Steinebach, has carried on its normal function—that of forwarding greetings to conferences of the overseas associations. This next year it has before it the work of organizing a party of delegates to attend the International Congress to be held in London next June.

Foundry Cost Committee

With representation from the Steel Founders' Society, Gray Iron Founders' Society and Malleable Founders' Society, the Cost Committee, under the chairmanship of Sam Tour, Lucius Pitkin, Inc., developed a convention session for the discussion of melting costs. The committee plans to hold future sessions covering other phases of cost practices, the hope being that a greater uniformity in cost methods may be developed by the various trade associations.

Apprentice Training Committee

Again our Apprentice Training Committee has been one of the most active of Association committees. This past year it has held two convention sessions, issued a revision of the standards for four-year foundry apprenticeship, issued a preliminary report on standards for pattern making apprentices, and held the annual competition in gray iron, steel and non-ferrous molding and pattern making. It has further interested chapters in organizing apprentice committees and the holding of meetings to discuss training methods.

It is recommended to your Board that with the approval of the Awards Committee $350 be appropriated from the Awards Fund for the prizes and certificates for the Apprentice Contest for 1939.

Of interest in connection with the work of this committee have been booths at the Milwaukee and Cleveland conventions, sponsored by the Milwaukee Vocational School and the Cleveland Trade School, respectively. The Cleveland Trade School booth, as developed under the direction of Frank Cech, pattern instructor, and James G. Goldie, foundry instructor, was a decided attraction of the Cleveland convention

Foundry Refractories

The A.F.A. is cooperating with the Joint Committee on Foundry Refractories, the other sponsoring body being the American Ceramic Society. The Joint Committee has aided the A.F.A. in holding annual sessions on foundry refractories, and increasing interest is shown by the attendance and quality of the papers presented. The Joint Committee has developed standard classifications for cupola and malleable furnace refractories, these being approved by the Division of Simplified Practice, Department of Commerce. This year, the Joint Committee is being reorganized to carry on more active work than it has for the past three years.

Foundry Sand Research

A principal work of the Foundry Sand Research Committee this year has been the preparation and publication of the fourth and revised edition of the book "Standards and Tentative Standards for Testing and Grading Foundry Sands and Clays." This is now being distributed to all members on request. The cost of this publication has been met by the inclusion of advertising pages. In connection with this revision, we especially wish to commend the work of Norman Hindle, who attended to the great amount of editorial details involved. The assistance of Dr. Ries and his committee, particularly C. P. Randall, Hunt-Spiller Mfg. Co., W. G. Reichert, Singer Mfg. Co., C. M. Saeger, Jr., U. S Bureau of Standards, and H. W. Dietert cannot be too highly praised.

The Sand Research Committee, under Dr. Ries, has also prepared and has ready for publication a second pamphlet on properties of sand, a treatise to explain in a practical way the application of the test methods, in reality a preparatory pamphlet for the use of the Handbook on Standard Tests. A third important piece of work has been the investigation of sands at high temperatures, funds for which were appropriated by your Board. Explanation of this work, started in the interests of the Steel Division, will be made in a separate report.

Chapters

Our chapter roll now stands at 14. This past year there have been added the Metropolitan New York-New Jersey Chapter, the Northern Illinois and Southern Wisconsin Chapter, and the Ontario Chapter. The Ontario Chapter has been developed from the membership of the Canadian Section, with the Canadian Section still remaining in force. These chapters are all holding regular meetings and are most certainly showing their effect on membership as the large majority of new members are coming in from chapter territory. Total attendance at the regular meetings of all chapters is estimated conservatively at 13,000. We feel that the chapter officers and directors are in all cases to be highly commended for their excellent work.

Cooperation with Other Associations

The A.F.A., since 1921, has had an agreement with the A.S.T.M., recognizing it as the specifications making organization of the United States. This agreement permits the A.F.A. to have official representation on all A.S.T.M. committees dealing with subjects of interest to A.F.A. We now have 16 such representatives, two new representatives being appointed this past year—on the Committees on Radiography and Corrosion. The A.F.A. is represented on 15 committees functioning under the procedure of the American Standards Association. We have sponsored with the N.F.A. Project B-8, Safety Code for Protection of Industrial Workers in Foundries and sole sponsor of project B-45.1, Foundry Pattern Color Standards, both of which are approved standards. The A.F.A. is cooperating with the Division of Simplified Practice in sponsoring recommended practices for Foundry Refractories and Standard Pattern Colors. Three practices have been published and are periodically sent out by the Division for reapproval or revisions. These three are (1) Standards for Malleable Refractory Shapes and Sizes, (2) Standards for Cupola Refractory Shapes and Sizes, (3) Pattern and Core Box Color Markings.

Metal Divisions

Our four Metal Divisions—namely, Steel, Malleable, Cast Iron and Non-Ferrous, are our most important activities, each headed by an advisory committee, and are carrying on extensive committee work and take a most prominent part in developing the program for the annual convention.

Steel Division (Chairman—John Howe Hall)

In the past two years, the activities of the Steel Division have been put on a good basis, with ten technical committees functioning under the direction of an executive committee. The reports of these committees are now providing much of the material for the convention Steel Sessions. The committees are (1) Methods of Producing Steel for Castings, (2) Heat Treatment, (3) Design for Steel Castings, (4) Radiography, (5) Test Coupons, (6) Impact Testing, (7) Sand Research, (8) Program, (9) Round Table Conference, (10) Handbook Revision.

Malleable Division (Chairman—Leon J. Wise)

The Malleable Division must be given credit for the greatest advance in its convention program over preceding years. The division held two technical sessions, a round table conference and a two-session shop course, at Cleveland, the papers and discussions being outstanding.

Non-Ferrous Division (Chairman—Harold J. Roast)

The Non-Ferrous Division operates under its own set of regulations, providing for election of officers and committee appointments. Active committees are: (1) Research, cooperating with Ingot Metals Institute, (2) Recommended Practices, (3) Analysis of Defects, (4) Program (5) Round Table Conference, (6) Handbook Revision.

Two sessions and a round table conference were held at Cleveland. The papers presented were probably the most interesting group ever gathered at an A.F.A. convention.

Gray Iron Division (Chairman—G. P. Phillips)

The division has as its most important piece of work the preparation of a publication covering alloy cast iron.

Other important committees are (1) Program, (2) Shop Operation, (3) Round Table Conference, (4) Handbook Revision, (5) Definition of Foundry Terms, cooperating with the International Dictionary committee, (6) Cooperation with Engineering Schools, (7) Joint Committee on Pig Iron Qualities, cooperating with A.S.T.M. and A.I.M.E., (8) Fluidity Testing, (9) Welding Cast Iron, (10) Classification of Graphite in Gray Cast Iron, cooperating with A.S.T.M. Committee A-3, and International Foundry Committee.

Acknowledgment

In closing, your Technical Secretary, wishes to acknowledge the real cooperation of the many committee members, division and committee chairmen and authors of papers, who, working together, have developed the reputation of our Association to a point where it stands with the best of technical organizations. We further wish to stress the excellent cooperation and diligent efforts of our associate, Norman Hindle, who has borne the major share of the detail work of the Technical Department.

Respectfully submitted,
R. E. KENNEDY, *Technical Secretary*

Reports of Board of Directors and Executive Committee Meetings

FOLLOWING the annual meetings of the Board of Directors on September 22, 1937 in Cleveland, which were reported in Bound Volume 45, official meetings were held as follows:

February 9, 1938, Chicago—Meeting of Executive Committee.

May 15, 1938, Cleveland—Special Meeting of Board of Directors.

May 20, 1938, Cleveland—Annual Meeting of 1937-38 Board of Directors.

May 20, 1938, Cleveland—First Meeting of 1938-39 Board of Directors—(Recessed to re-convene in Chicago, July 26).

July 26, 1938, Chicago—Continuation of First Meeting of 1938-39 Board—(Recessed from meeting in Cleveland, May 20).

Minutes of these meetings are given below.

Minutes of Meeting — Executive Committee

CHICAGO, FEBRUARY 9, 1938.

The meeting was called to order in the office of the Association.

Members present: President Hyman Bornstein, Executive Vice-President C. E. Hoyt and Directors H. S. Washburn, J. R. Allan, Jas. L. Wick, Jr., and L. N. Shannon, members of the Executive Committee.

Absent: Vice-President Marshall Post.

Present by invitation: Secretary-Treasurer D. M. Avey and Technical Secretary R. E. Kennedy.

The meeting was called by President Bornstein to consider and act upon various items of business.

Technical Secretary Kennedy presented and discussed the forthcoming revised publication "Testing and Grading Foundry Sands and Clays." Mr. Kennedy reported that this revised edition had been recommended by the Committee on Foundry Sand Research for approval as standard. On motion of Mr. Wick, seconded by Mr. Allan, the Executive Committee approved the forthcoming publication as a standard of A.F.A.

Discussion on policy relating to advertising with particular reference to accepting two-page spreads in this revised sand testing publication was referred to Mr. Hoyt, Mr. Kennedy and Mr. Avey for decision.

Mr. Allan suggested the inclusion of advertising in engineering codes being prepared under direction of the A.F.A. Committee on Safety and Hygiene Engineering Codes. After discussion, the question was referred to a committee consisting of Mr. Allan, Mr. Hoyt and Mr. Avey.

Discussion of the proceedings to date in relation to the revision of By-Laws of the Association, a recount of the presentation of the revised By-Laws to each chapter board and discussion of the procedure to be followed in presenting the By-Laws, followed. President Bornstein requested that decision on these matters be referred to the Executive Committee at its afternoon session.

The Secretary reported on chapter relationships with the A.F.A. headquarters office. President Bornstein gave a detailed report of the situation prevailing in the newly established Metropolitan Chapter, embracing northern New Jersey and eastern New York. On motion of Mr. Shannon, seconded by Mr. Washburn, the Executive Committee voted to refund 15 per cent of the dues of members in the new chapter area covering the entire year from July 1, 1937 to June 30, 1938.

The Executive Committee directed the Secretary to write the officers of the new chapter suggesting that some term denoting geographical location be coupled with the present name of Metropolitan. Territorial area requested by the new chapter was approved, subject to future change by the Board, should this prove necessary.

At the start of the afternoon session, the suggestion was made that the By-Laws be submitted to the membership for approval at sometime prior to the annual convention at the discretion of the officers. On motion of Mr. Shannon, seconded by Mr. Allan, the Executive Committee recommended the early adoption of the new By-Laws, but decided to advise the Board to defer the application of the new dues rates to January 1, 1939 and directed the Secretary to refer this decision to the Board for letter ballot.

The Secretary was directed to contact chairmen and secretaries of all chapters simultaneously with submitting the By-Laws to vote, soliciting their assistance in securing full vote on this question.

Further discussion on chapter matters brought the suggestion from Director Shannon that older chapters be invited to stimulate and encourage the formation of new chapters where a concentration of membership on the outer fringes of their territory would permit this formation. He proffered his personal assistance and that of Birmingham Chapter men in stimulating action in Cincinnati. The Secretary was asked to visit Cincinnati, secure the cooperation of members for a called meeting to which Mr. Shannon will be invited.

The Secretary-Treasurer presented the report of receipts and disbursements covering the period from July 1, 1937 to January 31, 1938. President Bornstein referred this report to the Finance Committee and requested that copies of the monthly reports be supplied regularly to the Finance Committee and that quarterly reports be submitted by mail to the Board.

The question of budget for the remainder of the year was discussed by members of the Executive Committee and the details of budget for the remainder of the year were referred to the Finance Committee.

Executive Vice-President Hoyt reported on the status of the forthcoming Cleveland exhibition and discussed obtaining a lecturer for the business meeting and a speaker for the annual banquet. The Executive Committee requested the Technical Secretary to arrange to schedule the business meeting, with a special lecturer to be designated and compensated by the Board of Awards at their pleasure, for a morning session. The Technical Secretary reported that by a rearrangement of this tentative program, it will be possible to schedule the business meeting for 11:00 A.M., Wednesday, May 18. Mr. Hoyt reported that the Music Hall, seating 3100, will be available for this occasion.

The Secretary discussed the necessity for acquiring new addressograph equipment to facilitate handling the increasing membership list and also to permit the establishment of a personnel list of foundry executives in cooperation with manufacturers of equipment and producers of supplies. On motion of Mr. Shannon, seconded by Mr. Allan, the Secretary was directed to secure the required equipment and personnel for this listing service.

Technical Secretary Kennedy presented the invitation of the University of Michigan to hold a fall regional conference of the A.F.A. at Ann Arbor in 1938. On motion of Mr. Shannon, seconded by Mr. Wick, this invitation was accepted and the Technical Secretary was directed to proceed with the arrangements.

Executive Vice-President Hoyt presented for discussion the invitation of the Rosenwald Museum of Science and Industry to provide fifty-six bronze panels for the doors of the Museum to be cast from plaster patterns supplied by the Museum and to cost in the neighborhood of $1,500.00. After discussion, it was decided that A.F.A. does not have funds for a local undertaking of this nature.

Discussion of the forthcoming publication on alloy cast iron followed. On motion of Mr. Shannon, seconded by Mr. Wick, the Executive Committee authorized the office to solicit advertising from manufacturers of alloy castings, alloying materials and such other products as might be advertised in this publication.

A communication from the Milwaukee Chapter requesting that A.F.A. allow expenses for officers of chapters attending the annual convention was discussed. It was pointed out that funds were not available for such undertaking and on motion of Mr. Shannon, seconded by Mr. Allan, the request was disallowed.

Respectfully submitted,

D. M. AVEY, *Secretary*

Minutes of Meeting of Board of Directors

CLEVELAND, O., MAY 15, 1938.

This special meeting of the Board of Directors was called to order by President Hyman Bornstein at the Hotel Statler, 6:30 P.M.

Present: President H. Bornstein, Vice-President Marshall Post, Executive Vice-President C. E. Hoyt, Secretary-Treasurer D. M. Avey, Technical Secretary R. E. Kennedy and Directors W. J. Cluff, L. S. Peregoy, H. S. Washburn, C. C. Gibbs, L. N. Shannon, C. E. Sims, T. Kaveny and J. L. Wick, Jr.

President Bornstein called upon Executive Vice-President Hoyt for any report or recommendation in connection with the convention. None was offered.

On motion of Mr. Wick, seconded by Mr. Peregoy, the Board of Directors recommended to the members the election to honorary life membership of retiring President Hyman Bornstein, and Seaman Medalist Frederick Ayres Lorenz, Jr.

There being no further business, the meeting adjourned, subject to the call of the President.

Respectfully submitted,

DAN M. AVEY, *Secretary*

Minutes Annual Meeting 1937-1938
Board of Directors

CLEVELAND, O., MAY 20, 1938.

Members present: President Hyman Bornstein, Vice-President Marshall Post, Executive Vice-President C. E. Hoyt, Secretary-Treasurer D. M. Avey and Directors W. J. Cluff, J. R. Allan, C. C. Gibbs, H. B. Hanley, L. N. Shannon, C. E. Sims, D. P. Forbes, T. Kaveny and Jas. L. Wick, Jr.

Others present: Directors-Elect H. S. Hersey, D. O. Thomas and G. A. Seyler and Technical Secretary R. E. Kennedy.

The annual meeting of the Board of Directors was called to order by President Bornstein at 9:00 A.M., at the Hotel Cleveland.

Secretary D. M. Avey read the minutes of the preceding annual meeting, held in Chicago, July 20, 1937. On motion of Mr. Wick, seconded by Mr. Cluff, the minutes were approved.

Executive Vice-President Hoyt reported in brief form matters relating to the business of the Association and the annual convention and exhibition, requesting the privilege of filing with the minutes a more complete and detailed report.

Secretary-Treasurer Avey presented a statement of income and disbursements covering the period from the start of the fiscal year to May 1, 1938; and also a statement of the membership status covering that period.

In reviewing matters connected with chapter organization, the Secretary referred a communication from the Northern Illinois Foundrymen's Association which has petitioned for admission as a chapter. This petition has been approved. On motion of Mr. Wick, seconded by Mr. Shannon, the Secretary was directed to arrange with the new chapter for the proper name designation and territorial limits.

The Secretary reported the actions of the Executive Committee and letter ballots of the Board of Directors in the period following the last annual meeting. These actions were as follows:

Time and place—1938 Convention and Exhibition   July 27, 1937
    Approved........15   Not Received......2
    Disapproved 0

Bound Volume (Prices—Non-Members $10, Members $3)   Nov. 25, 1937
    Approved........15   Not Received......2
    Disapproved 0

Dues Refund to Canadian Section (Executive Comm. only) Dec. 18, 1937
    Approved........ 7

International Foundry Congress in U.S.—1942   Dec. 29, 1937
    Approved........15   Not Received...,..1
    Disapproved 1

Petition for New York-New Jersey Chapter   Jan. 3, 1938
    Approved........15   Not Received.,....2
    Disapproved..... 0

Approval of By-Laws for Submission to Membership   Jan. 5, 1938
       Approved........16   Not Received......1
       Disapproved 0
Appropriation of $1,650 for Research Project   Jan. 14, 1938
       Approved........16   Not Received......1
       Disapproved 0
By-Laws Dues Rates to Apply as of Jan. 1, 1939   Feb. 22, 1938
       Approved........16   Not Received......1
       Disapproved 0
Recommendation of New Code as a Tentative Standard   April 6, 1938
       Approved........15   Not Received......1
       Disapproved 0
Petition for Northern Illinois Chapter   April 29, 1938
       Approved........16   Not Received......1
       Disapproved 0
Seaman Medal Award to F. A. Lorenz, Jr.   May 7, 1938
       Approved........15   Not Received......1
       Disapproved 0   *   *   *

Technical Secretary Kennedy presented a report in brief form on technical activities, announced the apprentice awards winners and requested the privilege of extending his report for inclusion with the complete minutes of the annual business meeting.

On motion of Mr. Post, seconded by Mr. Shannon, the Board approved the report of the Executive Vice-President and Manager of Exhibits subject to supplementary extension.

On motion of Mr. Allan, seconded by Mr. Post, the Board approved the report of the Secretary-Treasurer subject to detailed extension and inclusion of the annual financial report and audit.

On motion of Mr. Shannon, seconded by Mr. Sims, the Board approved the report of the Technical Secretary, subject to extension in the minutes.

Secretary Avey reported on the work of the A.F.A. Industrial Hygiene Codes Committee consisting of the following:

Chairman, J. R. Allan, International Harvester Co., Chicago, Ill.
       R. J. Aldrich, Sloan Valve Co., Chicago, Ill.
       F. H. Amos, International Harvester Co., Chicago, Ill.
       C. P. Guion, W. W. Sly Mfg. Co., Chicago, Ill.
       E. E. Birkland, Crane Co., Chicago, Ill.
       Carl F. Larsson, American Air Filter Co., Inc., Chicago, Ill.
       S. McMullan, Western Electric Co., Chicago, Ill.
       J. G. Liskow, Claude B. Schneible Co., Chicago, Ill.
       Nathan Lesser, Deere & Co., Moline, Ill.
       R. W. McCandlish, Research Corp., Chicago, Ill.
       John F. Tobin, American Blower Corp., Chicago, Ill.
Secretary, E. O. Jones, American Foundrymen's Association, Chicago, Ill.

Conferees: G. A. Kluge, Central Blow Pipe Co., Chicago, Ill.
       J. M. DallaValle, U. S. Public Health Service, Washington, D. C.
       L. C. Stokes, Dept. of Labor, State of Illinois, Chicago, Ill.
       C. S. Anderson, Belle City Malleable Iron Co., Racine, Wis.
       L. S. Peregoy, Sivyer Steel Casting Co., Milwaukee, Wis.
       J. O. Houze, National Malleable & Steel Castings Co., Cicero, Ill.

On motion of Mr. Avey, speaking as a director, and seconded by Mr. Forbes the following resolution was unanimously adopted:

Whereas, the Industrial Hygiene Codes Committee, through its work on the preparation of engineering codes is performing a distinct service to the industry, and

Whereas, these codes present valuable engineering data, much of which previously was not available, and

Whereas, the use and application of these codes will be of tremendous service to the industry directly in improvement of plant and operation practice, now, therefore, be it

RESOLVED: That the Board of Directors commend and express the appreciation of the American Foundrymen's Association to Chairman Allan and the members of the Industrial Hygiene Codes Committee, for their fine work.

The Secretary was directed to report this resolution in the minutes and to provide copies for each of the members of the Industrial Hygiene Codes Committee.

Technical Secretary Kennedy reported on the alloys of iron research program and also on the status of the American Standards Association relations.

President Bornstein and Technical Secretary Kennedy reported the proceedings of the chapter conference held during the convention. On motion of Mr. Post, seconded by Mr. Kaveny, the following resolution commending the Northwestern Ohio Chapter and its committees was adopted:

Whereas, the Northwestern Ohio Chapter of the American Foundrymen's Association, host to the 1938 convention and foundry show, has shown unusual initiative and diligence in assisting in the presentation of these events, and

Whereas, the Northwestern Ohio Chapter and its committees have extended hospitality to the members of the American Foundrymen's Association and guests, remarkable for its extent and sincerity, now, therefore, be it

RESOLVED: That the Board of Directors of the American Foundrymen's Association, acting for the membership, do hereby offer and extend to the Northeastern Ohio Chapter and the various committees thereunder, sincere thanks and appreciation.

The Secretary reported that by individual letter, all members of the Board had gone on record approving deferring the new dues structure provided in the new By-Laws, and to permit all members of record on July 1, 1938, to continue on the same dues rate until July 1, 1939. On motion of Mr. Post, seconded by Mr. Shannon, this action was confirmed by formal vote.

The Secretary requested direction of the Board in the matter of billing for dues of members of record on July 1, 1938, mentioning the elimination of the Limited Firm and Subsidiary Plant classifications and the provision of the By-Laws which permits affiliation of non-

resident salesmen reporting to plants of Company members. On motion of Mr. Shannon, seconded by Mr. Cluff, the Board directed the Secretary to bill all members of record on July 1, 1938, in accordance with their past dues rate.

President Bornstein offered a letter, comprising recommendations of the retiring president to the incoming Board of Directors.

On motion of Mr. Gibbs, seconded by Mr. Allan, the Treasurer was authorized to pay expenses of all board members present at the meeting for the additional day required by remaining to attend the meeting after the convention.

On motion of Mr. Wick, seconded by Mr. Shannon, it was recommended that the incoming Board study the complete report of the Fact Finding Committee, appointed during 1937.

Director Hanley reported on the Cornell meeting and plans for 1938.

On motion of Mr. Cluff, seconded by Mr. Post, the Board commended the work of all authors, committee members, chairmen, chapter officers, firms and individuals who contributed by their work to the success of the 1938 convention.

Mr. Shannon reported on the progress of the Birmingham Chapter and the success of the annual two-day spring conferences. .

On motion of Mr. Sims, seconded by Mr. Forbes, the Board recommended preparation and printing of a directory of members of A.F.A. during the forthcoming year.

President Bornstein introduced the incoming president, Marshall Post, and the new directors, Mr. Hersey, Mr. Thomas, and Mr. Seyler.

Meeting adjourned.

Respectfully submitted. .

DAN M. AVEY, *Secretary*

Minutes First Meeting 1938-1939 Board of Directors

CLEVELAND, O., MAY 20, 1938.

President Post called the new Board to order at 11:15 A.M., Friday, May 20, 1938, at the Hotel Cleveland.

Present: President Marshall Post, Executive Vice-President Hoyt, Secretary-Treasurer D. M. Avey, and Directors J. R. Allan, C. C. Gibbs, L. N. Shannon, C. E. Sims, D. P. Forbes, H. B. Hanley, T. Kaveny, Jas. A. Wick, Jr., H. S. Hersey, D. O. Thomas, G. A. Seyler, H. Bornstein and Technical Secretary R. E. Kennedy.

President Post addressed the Board briefly expressing his pleasure in the responsibilities and his desire to further the work of the Association. Retiring Director Cluff commented upon the pleasure of his work as a director, and expressed his willingness to cooperate as a member of the alumni.

Mr. Shannon moved that the election of officers as provided for in Article 4, Section 1, of the By-Laws, the election of the Executive Committee and all other actions in completing the organization for the succeeding year, be deferred to an adjourned meeting of the Board of Directors to be called after the end of the fiscal year, July 1, 1938.

This motion was seconded by Mr. Hanley and unanimously approved.

Mr. Bornstein moved the adoption of the resolution customarily required by the Harris Trust and Savings Bank for the establishment of accounts and withdrawal of funds. It was further resolved that checks for withdrawal of funds, general checking accounts, all interest savings accounts, all securities of the sinking fund of the American Foundrymen's Association shall require the signature of any two of the following officers of the Association: President, Vice-President, Executive Vice-President and Secretary-Treasurer. On being seconded by Mr. Hanley, this resolution was approved.

On motion of Mr. Gibbs, seconded by Mr. Kaveny, a Petty Cash Checking Fund of $1,000 was approved, to be used in paying incidental expenses and reconciled at the end of each month by a full statement of expenditures, the withdrawal check to be signed by the Secretary-Treasurer or Executive Vice-President.

On motion of Mr. Shannon, seconded by Mr. Hersey, the following resolution was unanimously adopted:

Resolved: The officers of this Association are hereby authorized to open a checking account in the name of the Association in the convention city in 1939 and that the Executive Vice-President or the Secretary-Treasurer are hereby authorized to execute the resolutions required by the bank and to determine how this account shall be opened and the signatures required for the withdrawal of funds.

On motion of Mr. Wick, seconded by Mr. Sims, it was unanimously voted to continue in effect the resolutions adopted at the last annual board meeting, authorizing the Secretary-Treasurer to reimburse committee members and board members for traveling expenses at regular called meetings with the exception of such meetings as are held during convention week, unless they be specifically authorized by the Executive Committee.

On motion of Mr. Hanley, seconded by Mr. Allan, the Board directed that funds be made available for apprentice contests prizes at the 1939 convention to the extent of appropriations made available by the Board of Awards from awards interest fund account, and that the Board of Awards be advised of this action and requested to make suitable appropriations.

Director Kaveny urged consideration of castings promotion by A.F.A. Director Thomas expressed his interest in the promotion phase of Association activity pointing to the decline in tonnage of malleable castings represented in the automotive industry. A general discussion of the desirability of A.F.A. partcipation in this work followed but no action was taken.

On motion of Mr. Kaveny, seconded by Mr. Seyler, the meeting recessed to reconvene at the call of the President, after the end of the fiscal year, June 30, 1938.

Respectfully submitted,

DAN M. AVEY, *Secretary*

Minutes of Meeting — Board of Directors

(Recessed from May 20, 1938)

CHICAGO, JULY 26, 1938.

Present: President Marshall Post, Vice-President H. S. Washburn, Directors J. R. Allan, L. N. Shannon, C. E. Sims, D. P. Forbes, H. B. Hanley, Thomas Kaveny, Jas. L. Wick, Jr., H. Bornstein, H. S. Hersey, Executive Vice-President C. E. Hoyt, Secretary-Treasurer and Director D. M. Avey, Technical Secretary R. E. Kennedy.

Absent: Directors C. C. Gibbs, C. J. P. Hoehn, G. A. Seyler and D. O. Thomas.

This meeting, a continuation of the meeting of May 20 at Cleveland, and recessed until after June 1, the end of the fiscal year, was called to order with President Post presiding.

In calling the meeting to order, President Post announced the death of Director Frederick A. Lorenz, Jr., on July 23, and following appropriate remarks by the President, Mr. Bornstein moved that resolutions be prepared for the records, expressing the sympathy of the Board and that a copy of the resolution be forwarded to Mrs. Lorenz. Motion duly seconded and unanimously carried.

The Secretary read a letter from Director Thomas, presenting his resignation, stating that since his election he had severed his connection with the Saginaw Malleable Iron Division of the General Motors Corporation and was no longer actively connected with the foundry industry.

Mr. Wick moved that Mr. Thomas' resignation be accepted and that a letter be sent to him expressing the Board's regret and extending best wishes for his success in his new connection. Motion seconded and carried.

Approval of Minutes

The minutes of the meeting at Cleveland, May 20, 1938, were read by the Secretary and approved as read.

Statement by Secretary Avey

President Post announced that Mr. Avey had asked the privilege of addressing the Directors and would do so at this time.

Mr. Avey read a prepared statement reviewing A.F.A. policies, including publications, indicating he was not in accord with the recently adopted publication program and some other Association policies and concluded by saying that if the present program of activities was to be continued, he felt his services were not needed and he should resign.

Following Mr. Avey's presentation, he, Mr. Hoyt and Mr. Kennedy retired and the Directors went into Executive Session.

Tribute to the Memory
of
Frederick Ayres Lorenz, Jr.

Whereas, in the death of Frederick Ayres Lorenz, Jr., July 23, 1938, the Foundry Industry was deprived of an outstanding leader, and

Whereas, in his career, Mr. Lorenz established for all foundrymen a high standard and concept of public service, and

Whereas, his associates and many friends were deeply impressed by his readiness in conference, by his aptitude for stripping an involved situation and reducing it to its essentials by his unerring judgment, and

Whereas, in all of his work through the American Foundrymen's Association, and other organizations, Mr. Lorenz won recognition exemplified by high office and by such honors as the Joseph S. Seaman Medal awarded to him in 1938, all testifying to his high character, ability and capacity for human understanding, now, therefore, be it

Resolved, that the Board of Directors of the American Foundrymen's Association by unanimous action declare their regret and sorrow at his loss and their gratitude for his life and service, and be it further

Resolved, that this expression be recorded in the transactions of the American Foundrymen's Association and a copy thereof conveyed to Mrs. Lorenz, that she and her family may preserve this token of the affection of the many friends of FREDERICK AYRES LORENZ, JR.

AMERICAN FOUNDRYMEN'S ASSOCIATION
MARSHALL POST, *President*
DAN M. AVEY, *Secretary*

Executive Session

Election of Directors

President Post called for the report of the Nominating Committee appointed to nominate directors to fill vacancies on the Board. The report was presented by the chairman, Mr. Wick, nominating A. Walcher, Vice-President, American Steel Foundries, to succeed Frederick A. Lorenz, Jr., deceased, and W. H. Doerfner, Works Manager, Saginaw Malleable Iron Division, General Motors Corporation, to succeed D. O. Thomas, resigned.

Their election was moved by Mr. Wick, seconded by Mr. Bornstein and carried unanimously. The President announced Mr. Doerfner had previously been notified by the Nominating Committee of his selection to succeed Mr. Thomas and invited to come to Chicago that he might, following the election, sit in with the Board at this meeting. He was then introduced to the members present.

The Board, continuing in executive session, took under consideration Mr. Avey's comments and supporting documents he had left with them. Following discussion, the Directors present by unanimous vote accepted Mr. Avey's resignation, effective at once, and authorized his salary to be paid to September 30, 1938. Messrs. Avey, Hoyt and Kennedy returned to the room, and the Directors proceeded with the next order of business.

Organization of New Board

President Post announced that the Board would proceed to organize in accordance with the provisions of the By-Laws, Article V, Section 1 and Article XI, Section 1, and called for the report of the Nominating Committee.

The Committee, consisting of Mr. Wick, Chairman, and Messrs. Bornstein and Washburn, presented the following nominations for officers:

For Executive Vice-President and Manager of Exhibit—C. E. Hoyt
For Secretary—R. E. Kennedy
For Treasurer and Director of Safety and Hygiene—E. O. Jones
For Technical Secretary—N. F. Hindle
For Assistant Secretary-Treasurer—Jennie Reininga

There being no further nominations, Mr. Wick moved election of officers nominated. Motion seconded and unanimously carried, whereupon the chair announced officers as submitted by the Nominating Committee, elected to hold office to the end of the fiscal year, June 30, 1939, or until their successors were elected and qualified.

Mr. Wick, continuing, reported that the nominations of the Committee for four members of the Executive Committee to serve with the President, Vice-President and Executive Vice-President were Directors L. N. Shannon, H. S. Hersey, W. H. erfner and H. Bornstein. On motion of Mr. Wick, duly seconded, these members were declared elected as members of the Executive Committee.

Change in Resolution Authorizing Signature for
Withdrawal of Association Funds

Mr. Bornstein moved that the resolution passed at the meeting of the Board, May 20, 1938, on withdrawal of funds be changed to read as follows:

"Checks for the withdrawal of funds, general checking accounts, all interest savings accounts and all securities of the sinking fund of the American Foundrymen's Association shall require the signature of any two of the following officers of the Association: President, Vice-President, Executive Vice-President and Treasurer."

Motion seconded by Mr. Shannon and carried.

Mr. Bornstein moved also that resolution passed at the May meeting covering signature for withdrawal of petty cash fund be changed to read as follows:

"Withdrawal checks on petty cash fund shall be signed by the Executive Vice-President, Treasurer or Assistant Secretary-Treasurer."

Reports of Officers

The following reports were called for and presented:
Report of Secretary on membership.
Report of Treasurer.

On presentation of these reports which were read by Mr. Avey, it was duly moved, seconded and carried that they be accepted and ordered filed.

Report of Technical Secretary by R. E. Kennedy.

Mr. Kennedy stated that to save time, his 12-page report had previously been mailed to the Board, and that later he would be pleased to answer any questions members wished to ask.

Following discussion, it was moved, seconded and carried that the Directors extend congratulations on the accomplishments of the Technical Department during the past year, and that the report be accepted and ordered filed and consideration be given to having the report printed and made available to all members.

The report of the Manager of Exhibits was read by Mr. Hoyt, approved as read and ordered filed.

The report of the Director of Safety and Hygiene was presented by Mr. Jones, who advised that copies of his report had been mailed to all Directors and that he would be pleased to answer questions Directors wished to ask. Following discussion, it was moved, seconded and carried that the report be approved as presented and ordered filed.

Report of Finance Committee

The report of the Finance Committee was presented by Mr. Bornstein, Chairman. He stated that his committee had examined the auditor's report for the fiscal year ending June 30, 1938, and the more

detailed report prepared by Assistant Treasurer J. Reininga, and submitted a summary of the balance sheet, covering general fund, award funds and reserve fund.

Mr. Bornstein stated that the Finance Committee had prepared a tentative budget of expenses which was subject to revision after the program as well as the place and character of the 1939 annual convention had been determined.

He stated that it was the recommendation of the Committee that the budget when approved, be observed to the extent that expenditures under any heading should not be exceeded without the approval of the Finance Committee, and that the Treasurer submit copies of budget statement each month to members of the Finance Committee.

On motion duly seconded, the report of the Finance Committee was accepted and ordered filed.

Mr. Bornstein then submitted the recommendations of the Finance Committee on salaries of officers for the year beginning July 1, 1938, and moved that the recommendations of the Finance Committee on salaries be approved. On motion duly seconded they were approved, and the report of the Committee made a matter of record.

Mr. Bornstein then offered a resolution authorizing payment of stenographic and clerical compensation as recommended by the Executive Vice-President and approved by the Finance Committee. Resolution seconded by Mr. Shannon and approved.

Place of 1939 Convention

The Executive Vice-President presented a report on invitations received from various cities for holding the 1939 convention, reviewing facilities offered in various cities for accommodating a convention with or without exhibits. He stated that the only invitations received supported by chapters of A.F.A. was one from St. Louis, supported by the chapter there and one from Los Angeles supported by the Southern California Chapter as well as the Northern California Chapter, who by unanimous resolution had withdrawn their invitation and endorsed the invitation extended by Los Angeles.

Following discussion and consideration of the various invitations extended, Mr. Wick moved that all invitations be referred to a special committee to be appointed by the President, which would give further consideration to invitations and submit recommendations to the Board as to time, place and character of the 1939 convention for letter ballot, and that it also be the recommendation of the Board that the Executive Vice-President visit Los Angeles and personally inspect the accommodations offered there. Motion seconded by Mr. Kaveny and approved.

Membership Campaign

President Post, addressing the Directors, stated that plans were under consideration for a membership campaign under the direction of a general membership committee and special committees, with a special committee on the securing of Sustaining members of which Vice-Presi-

dent Washburn would be chairman, with Board members and members
at large appointed as members of this committee.

It was understood that the general committee and special committee
on the above class of membership would cooperate closely with the mem-
bership committees of the various chapters.

Meeting of the 1939 Nominating Committee and Board of Awards

Consideration was given to a time for holding on the same day,
meetings of the Nominating Committee and the Board of Awards, re-
sulting in Wednesday, January 18, 1939, being suggested as a day for
these meetings.

Mr. Bornstein, Chairman of the Board of Awards, invited sug-
gestions from the Board of Directors for medallists for the coming year
and also names of possible speakers for the annual lecture at the 1939
convention.

Appointment of Standing and Special Committees

On motion duly seconded the President was authorized to make and
approve appointments for all standing and special committees not
provided for in the By-Laws or by special act of the Board.

Ontario Chapter Petition

Mr. Kennedy presented a petition from the members of the Province
of Ontario for permission to form an Ontario Chapter. On motion duly
seconded, the petition was approved.

Change in Chapter Name

Mr. Hoyt presented a request from the Milwaukee Chapter that it
be given permission to change its name from the Milwaukee Chapter
to Wisconsin Chapter, and that its territory be extended to take in the
State of Wisconsin. On motion duly seconded, change in name was ap-
proved and the question of territory extension deferred for further
consideration.

Chapter Delegate Meeting

Mr. Bornstein stated that he had discussed with other members of
the Board the desirability of holding a chapter delegate meeting at some
other time than the annual convention, but that for this year he would
recommend that arrangements be made for staging a meeting at the
next convention, with representation limited to not more than two
delegates from each chapter board, following presentation of a proposed
agenda for discussion at the meeting, the items in the agenda to be
discussed by the Board with the delegates prior to the chapter
conference.

Chapter Chairman Trophy or Award

Mr. Kennedy presented a request from the Chicago Chapter of
A.F.A. that the Board of Directors authorize the procuring of a master
pattern for a bronze plaque from which castings could be made and pur-
chased by chapters desiring to present such plaques to retiring chairmen.

Following discussion Mr. Shannon moved that this matter be made an item for discussion and placed on the agenda for the chapter delegate conference they to consider and make recommendations to the Board. Motion seconded and carried.

Chapter Dues Refund

Mr. Hoyt explained the situation regarding dues refund as provided for in the old By-Laws and revised in the new By-Laws and asked for a resolution of instruction as to the per cent of refund that should be made on dues for the current year.

On motion by Mr. Wick, duly seconded, the Treasurer was authorized to refund to chapters 15 per cent of dues from members of record prior to July 1, 1938 and paying the same rate of dues as last year, and a refund of 20 per cent on dues of new members paying the rate of dues provided for in the amended By-Laws this 20 per cent to apply to Sustaining as well as to other classes of membership. Motion seconded by Mr. Bornstein and unanimously approved.

Foundry Sand Research

Mr. Kennedy submitted a recommendation made by the Sand Research Committee that an appropriation of $750 be authorized for the continuation, until January 1, 1939, of the investigations into the effect of high temperatures on steel sands and binders being conducted at Cornell University. On motion by Mr. Shannon, duly seconded, appropriation was authorized.

Membership for Life

Mr. Hoyt stated that Mr. Jesse M. Darke who attended the organization meeting of A.F.A. in Philadelphia in 1896, and who has continued all these years an active interest in A.F.A., would retire from active duty as Chief Metallurgist, Lynn Works, General Electric Company, on August 31 after 46 years of service.

Mr. Hoyt stated that the election of Honorary Life members, as provided for in Section 5, Article II of the By-Laws, could only be made by a three-fourths vote of the Board of Directors, acting upon recommendation of the Board of Awards that such membership be conferred. He then recommended that in recognition of Mr. Darke's long years of service and membership he be made a member for life with payment of dues remitted.

On motion of Mr. Kaveny, duly seconded, it was unanimously voted to make Mr. Darke a member for life without payment of further dues.

The names of other longtime members were considered for life membership and it was recommended that further consideration be given to such memberships. President Post made the suggestion that arrangements be made for an old-timers' gathering at the annual convention.

Preprint Distribution Policy

Mr. Kennedy reviewed the present policy as to method of distribution of preprints under which all preprints available are sent to all members in advance of the convention. Following discussions as to other methods of distribution and the practice of other societies and associations, it appeared to be the consensus of opinion that for the present the method of distribution of recent years be continued.

Subscription Rates to Regular Publications

The Directors were told that with the change in publication policy, it would be necessary, in order to comply with postal regulations, to show in the By-Laws that a portion of members' dues were for subscription to publications if the second class mailing rates are to be available.

On motion by Mr. Sims, duly seconded, the showing of a subscription rate of $4.00 per year for Transactions and $1.00 per year for American Foundryman, and notice to the postal authorities to that effect was authorized.

Adjournment

President Post expressed his appreciation for the attendance and cooperation of all members of the Board and stated that a motion to adjourn would be in order. On motion, duly seconded, the recessed annual meeting of the 1938-39 Board of Directors of A.F.A. stood adjourned.

Respectfully submitted,

R. E. KENNEDY, *Secretary*

Atmosphere Control in Annealing Malleable Iron

BY E. G. DE CORIOLIS[1] AND R. J. COWAN[2], Toledo, Ohio

Abstract

A review of malleable annealing practice reveals the importance of surrounding the castings with materials of suitable composition. These are known to function by their control of the atmosphere which surrounds the metal. This atmosphere is usually too active to prevent decarburization. Because of this, the annealing rate is retarded, particularly at the decarburized edges. This results in the formation of white pearlitic rims on the castings. By separating the atmospheres from the packing materials, a wide range of gas compositions is made available with which to surround the castings during annealing. A discussion of fundamental principles indicates the applicability of these ideas to malleable annealing and their use for the prevention or retardation of decarburization. The proper use of atmospheres reduces the annealing time considerably. Their use is illustrated by a description of the operation of two types of apparatus, one being an intermittent and the other being a continuous furnace.

INTRODUCTION

1. The annealing of American black heart malleable iron has been given a lot of study during recent years. The purpose of this study has been the acceleration of the annealing rate. The commercial requirements of industrial operations have made necessary an increasing efficiency in all stages of the manufacturing processes. The time consumed in annealing malleable iron long has seemed to be a fruitful field for research, since any improvement which would reduce the time required for this operation would be a welcome step in the right direction.

2. It long has been known that the gases, which surround a malleable iron casting during the process of annealing, have a very great effect upon the resulting metal. Many annealing room

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superintendents will recall the care used in former days for the satisfactory preparation of packing material. These materials usually contained iron in an oxidized form. After having been used in annealing pots, it will be recalled how this material was spread out on the floor of the annealing room and re-oxidized carefully before being used again.

3. Frequently salammoniac was sprayed over this material and the whole exposed to air until a suitable oxidization had taken place. At other times, when the packing material had become too highly oxidized, so that it would adhere to the surface of the castings, it was customary to reduce this oxide by means of charcoal added to the packing material. These old procedures, now very much obsolete, indicate that there was an appreciation on the part of the early workers of the importance of surrounding the casting with a material of suitable character.

4. Considering the nature of the materials then in use, and in the light of our present knowledge, it appears evident that the packing had to be of a nature to combine with the carbon of the castings to produce an atmosphere of a protective character, in immediate contact with the casting's surface. We know now that this atmosphere was largely composed of carbon monoxide, as evidenced by the fact that it commonly was seen to burn at the surface of the pot. The volume and intensity of this flame acted as a gauge to the operator of the proper conduct of his operation.

5. When the annealing art was subjected to closer scientific scrutiny, it was found that the process of graphitization, which is the basis of the annealing art, was itself dependent upon a gaseous reaction. The idea became current that graphitization would not occur in the absence of carbon monoxide gas. It even was suggested that the use of carbon monoxide gas as an atmosphere with which to surround the metal during the annealing process would have an accelerating effect upon the graphitizing rate. Furthermore, these scientific studies had shown that the white rim or pearlitic edge, so frequently found upon the early malleables, was traceable to an atmospheric condition surrounding the metal during annealing.

6. It will be recalled when it was commonly said that "the strength of malleables is in the skin." The strength of the skin was shown to be due to a pearlitic edge, which was thought to be due to the fact that oxidization of the carbon during annealing had been more rapid than graphitization.

VARIOUS APPLICATIONS OF THE IDEA

7. There have been many attempts in the past to anneal malleable iron in special atmospheres. In addition to the example given, in which attempts were made to control this atmosphere by controlling the packing material, there have been cases where the casting to be annealed has been packed in closed refractory muffles. Frequently, there was placed within these muffles some combustible material such as wood or coke, which would combine with the air enclosed within the muffle to produce an atmosphere having non-scaling characteristics, which would serve to surround the metal during all stages of the annealing with a non-oxidizing atmosphere. Muffles of this type were not very efficient from the standpoint of fuel economy or tonnage capacity, but they did produce good malleable iron, usually free from objectionable scale.

8. Among the early attempts to use an atmosphere of modern type for the annealing of malleables, one having a composition that was made to vary in accordance with the temperature cycle was used. The reason for such an atmosphere was based upon a consideration of physical chemistry relationship between the composition of the atmosphere and the metal composition. An attempt was made to change the atmosphere surrounding the metal in accordance with the temperature changes during annealing, so as to have at all times an atmosphere that would be neutral to the metal. These ideas were not adopted widely, although they would seem to be fundamentally correct. The only limitation was a practical one which involved changing the atmosphere in accordance with changes in the temperature cycle.

9. It must be borne in mind also in any consideration which involves a controlled atmosphere, that malleable iron, during the annealing process, evolves rather large volumes of gases. An analysis of these gases would indicate principally carbon monoxide, but also some carbon dioxide. The liberation of these gases during the annealing operation has a very important effect upon the atmosphere problem and renders still more difficult the matter of an artificial change of composition.

10. It is apparent from a consideration of the foregoing that there were serious limitations to the wide acceptance of special atmospheres for malleable annealing. Many of the fundamental requirements were known, but there was not available any commercial equipment of suitable flexibility.

RADIANT TUBE HEATING AND ATMOSPHERE CONTROL

11. While these researches were going on in the malleable field, there had been developing in other fields, a wide-spread use of special atmospheres for the heat treatment of metals. At first, this work was done within heated metal muffles, but it soon became apparent that a furnace was required of greater cross-section than could possibly be used when depending upon cast metal muffle sections. These requirements led to the application of fuel fired radiant tubes as heating elements. This method of firing has been described in the literature with considerable detail, so that it is necessary only to mention it in this connection to indicate its use where large muffles are obviously excluded.

12. In addition to a special method for heating, there had been developed also an atmosphere generator suitable for use in connection with the bright annealing of steel. The atmosphere produced for this purpose came to be widely known as "DX" gas. It consisted of the products of combustion of mixtures of fuel gas and air, which were washed and dried previous to their use. The composition of this atmosphere could be varied within rather wide limits, extending from about 12 per cent carbon dioxide on the one side to 10 per cent carbon monoxide on the other side. When used in conjunction with a hydrocarbon gas, this "DX" gas had been suitably used for carburizing metals and found wide use for this purpose.

13. There had occurred simultaneous developments in the art of furnace construction involving large gas tight units equipped with mechanical devices for progressively moving work through these furnaces. In this connection, great strides had been made and considerable experience developed in the art of casting heat resisting alloys which were capable of standing up under high temperature and load conditions for long periods of time.

14. With this background of experience and accomplishment, it was but natural to project the application to the annealing of American malleable cast iron.

DECARBURIZATION AND ANNEALING

15. The presence of a pearlitic rim on malleable iron castings has been associated with the manufacture of the metal since the early days of the industry. It seemed that any attempt to

reduce the annealing time usually resulted in the formation of a pearlitic rim. It has been known for a long time that this condition could be traced to the decarburization of the metal at some stage of its heat treatment. It is known that when decarburization occurs, it is progressive from the surface toward the metal core, that is to say, while the metal in the interior of the casting may have a carbon content of 2.5 per cent, if it is decarburized during heat treatment, the carbon content at the surface may be reduced to a much lower value, depending upon the intensity of the decarburizing action.

16. It might be expected that a difference in carbon content between surface and core might result in a difference in the annealing rate between the two. The work conducted by Schneidewind[3] at the University of Michigan would suggest such a difference. Certain of his results have been condensed and plotted in Fig. 1 to illustrate this effect.

17. Since these rates are so different, the casting edge will be under annealed, while the interior will be thoroughly annealed and, to thoroughly malleableize the edge, it becomes necessary to extend the annealing time, especially that required for so called second stage annealing.

18. In overcoming the problem of the pearlitic edge, it is apparent that it may be undertaken in either one of two different ways. It may be done by prolonging the time of second stage

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3 Schneidewind, R. and White, A. E., University of Michigan Engineering Research Bulletin No. 24, August 1933, p. 13.
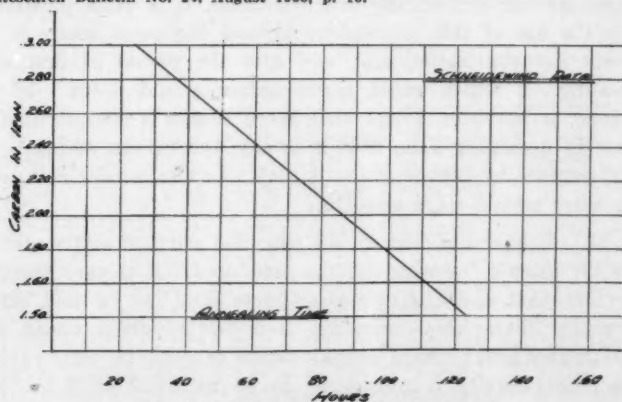
FIG. 1—SCHNEIDEWIND'S RESULTS CONDENSED.

annealing or, preferably, by preventing or even by retarding the rate of decarburization during first stage annealing. If decarburization is prevented entirely, it will be found that temper carbon is present at the extreme outer edge. This is desirable from the standpoint of machineability. It is not, however, always necessary to reach this degree of excellence, in which case it is necessary only to hold decarburization to an acceptable minimum.

19. Since the primary object of all this work is to reduce annealing time as well as to improve the product, our attention was concentrated on the problem of preventing decarburization during first stage annealing. Our studies soon indicated that decarburization occurs most rapidly during the first heating stages of the metal.

20. We observed active decarburization at temperatures below 1300°F. in an atmosphere of the "DX" type. It was found that this continued during the heating to maximum temperature and also during the first holding stage. After graphitization had begun, however, and as its rate increased, decarburization became less and less a factor.

21. It was found that, as these reactions progressed, there came a time when graphitization became more rapid than decarburization. This has been found to be a very important point and one which must be taken into account in the operation of this process.

22. If, in order to prevent decarburization, an atmosphere is used having carburizing characteristics, it is quite possible to carry the use of this atmosphere beyond the point where it will prevent decarburization and well into the period of first stage annealing, in which rapid carburization would occur. If this be done, it has been found that there results a considerable decrease in annealing rate, that is to say, the carbon added to the metal surface by carburization is very likely to persist as a white edge after second stage annealing.
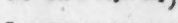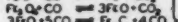
23. This would destroy the intended purpose and would not cure the disease for which it has been used. It is very essential, therefore, that a carburizing atmosphere shall not be used beyond the point where decarburization is a factor, which would seem to be at that point where graphitization is going on very rapidly. This point usually is considered to be within 2 or 3 hr. after the time maximum annealing temperature is reached.

ATMOSPHERE DIAGRAM

24. When iron is heated in the presence of an atmosphere, it tends to combine with the constituents of this atmosphere. If the atmosphere contains oxygen, the iron will scale. Carbon dioxide will similarly scale the iron. The presence of carbon monoxide will retard the scaling effect. If, however, carbon monoxide be in excess of carbon dioxide or air, it will prevent scale formation and if in sufficient excess, it will impart carbon to the iron surface. Under these conditions, the carbon monoxide reduces the oxygen of the air to carbon dioxide.

Authorities:-

$$2\ CO \rightleftharpoons C + CO_2 \qquad \text{Boudouard}$$
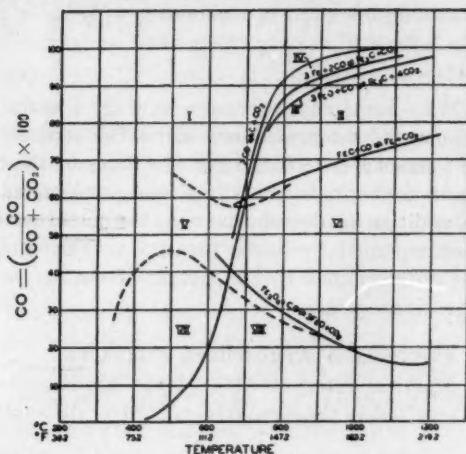
$$\left.\begin{array}{l} FeO + CO \rightleftharpoons Fe + CO_2 \\ Fe_3O_4 + CO \rightleftharpoons 3FeO + CO_2 \\ 3FeO + 5CO \rightleftharpoons Fe_3C + 4CO_2 \\ 3Fe + 2CO \rightleftharpoons Fe_3C + CO_2 \end{array}\right\} \quad \text{Matsubara}$$

THE DOTTED LINES ARE THOSE OF BAUR AND GLAESSNER

REACTIONS		$2CO$ $\rightleftharpoons$ $C + CO_2$	$FeO + CO$ $\rightleftharpoons$ $Fe + CO_2$	$Fe_3O_4 + CO$ $\rightleftharpoons$ $3FeO + CO_2$	$3FeO + 5CO$ $\rightleftharpoons$ $Fe_3C + 4CO_2$	$3Fe + 2CO$ $\rightleftharpoons$ $Fe_3C + CO_2$	Solid End Products in Equilibrium
	I	→	→	→	→	→	Fe_3C and Solid C
	II	←	→	→	←	←	Metallic Fe without Solid C
THE ARROWS INDICATE THE DIRECTION IN WHICH THE REACTIONS OCCUR IN THE FIELDS I, II, ETC.	III	←	→	→	→	←	Metallic Fe without Solid C
	IV	←	→	→	→	→	Fe_3C without Solid C
	V	→	←	→	DATA INCONCLUSIVE	DATA INCONCLUSIVE	FeO and Solid C
	VI	←	←	→	→	→	FeO without Solid C
	VII	→	←	←	DATA INCONCLUSIVE	DATA INCONCLUSIVE	Fe_3O_4 without Solid C

TEMPERATURE - CONCENTRATION DIAGRAM OF IRON-CARBON EQUILIBRIA AT A CONSTANT PRESSURE OF 1 ATMOSPHERE ABS.

FIG. 2—REACTION OF FREE OR COMBINED CARBON WITH SURROUNDING ATMOSPHERE, ACCORDING TO CLEMENTS.

25. If carbon be present, either as a free solid or in combination with iron, it will enter into reaction with the surrounding atmosphere to produce effects which will vary with the temperature and concentration of the constituents in the atmosphere. This can be best illustrated by consideration of Fig. 2 taken from Clements[4].

Gaseous Reactions

26. In this diagram, the proportion of carbon monoxide and dioxide are shown at the left and the temperature is the base line. For sake of simplification in considering only those reactions which directly affect annealing, Clements' diagram has been simplified as shown in Fig. 3.

27. It will be noted that there are now three main zones of reaction. In Zone *A*, at temperatures above the critical point of the metal, the tendency is to carburize. In Zone *B*, the tendency is to decarburize without, however, oxidizing the metal surface. In Zone *C*, in addition to decarburization, the metal surface also is oxidized accompanied by scale formation. The intensity of these reactions is determined by the distance from the lines dividing these respective zones.

CONTROLLED ATMOSPHERE FURNACES

28. The practical interpretation of the fundamentals above described has resulted in the development of two distinctive types

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4 Clements, Fred, "*Blast Furnace Practice*," Vol. 1.

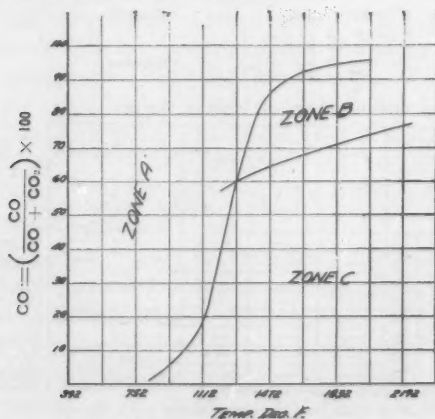

FIG. 3—CLEMENTS DIAGRAM SIMPLIFIED.

of furnace: the intermittent and the continuous. Each has a field
of application depending upon individual requirements of the
foundry. Each type has its advantages and its limitations.

29. As applicable to both types, the furnace is comprised
essentially of a gas tight shell, lined with suitable refractory and
heated by radiant tubes disposed both above and beneath the
work line. Supporting piers of heat-resisting alloy or refractory
or both are placed at necessary intervals between the bottom
radiant tubes. On the piers are mounted work supporting rails to
carry the load.

30. Hard iron is packed in alloy trays or baskets, usually
provided with a solid liner at the bottom to prevent sand from
the hard castings filtering downward and accumulating around
the radiant tubes. The design of the whole alloy structure, in-
cluding tubes, piers, rails and trays, requires careful engineering
to insure ease of operation and long life of these alloy parts,
since they constitute a major part in the cost of installation and
eventual maintenance.

31. The work is charged into the furnace through tight-
fitting doors which will insure minimum leakage, as it is the
intent of the operation to maintain a slight positive pressure in
the furnace chamber and exclude air infiltration. Suitable inlets
are provided for admitting the atmosphere gas. The temperature
of the furnace is controlled by pyrometer couples suitably dis-
posed at strategic points in the furnace structure.

ATMOSPHERE PREPARATION

32. It has been shown that the principal element concerned
in controlling the atmosphere for malleableizing is to secure the
proper balance between CO and CO_2. The most suitable source
for such atmosphere is fuel gas, either manufactured or natural.
The gas is burned in an enclosed chamber with a controlled
proportion of air which may be varied to yield a product as
high as 12 per cent CO_2 and no CO to a product as low as 5 per
cent CO_2 and 10 per cent CO. The hot products of combustion
are passed through a water cooled tower to remove excess mois-
ture and are delivered to the furnace at controlled pressure.
Atmosphere preparation units of this type have been developed
which operate automatically to hold any desired flow and com-
position within the limits set forth.

THE INTERMITTENT FURNACE

33. In the intermittent type of operation the hard iron castings are loaded into suitable alloy containers and charged into the furnace as a batch. The operation is periodic. After the doors are sealed and heat turned on, the first step is to purge all air from the furnace by means of a prepared gas which is non-combustible. As the castings approach the critical temperature, the composition of the atmosphere is changed to reduce the CO_2 content. There also is added at this time a quantity of raw unburnt gas to raise the CO content of the mixture above that which it is possible to produce in the preparation unit and to
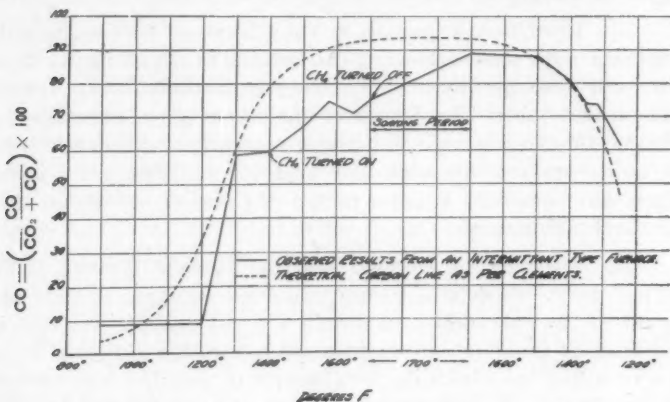
FIG. 4—RESULTS OF GAS ANALYSIS IN INTERMITTENT ANNEALING FURNACE.

lower the CO_2 in the atmosphere. The flow of raw gas is maintained up to the point of maximum temperature and for a short period thereafter. It is then turned off and the remainder of the annealing cycle, including the cooling period and second stage anneal, is carried through by the use of prepared gas alone.

34. Fig. 4 is a chart prepared from actual results obtained in the operation of one of these furnaces.

35. For sake of clarity, this chart shows the relative ratios of CO and CO_2 in the atmosphere as determined at the several stages of the operation and plotted in comparison with the curve illustrated in Clements' chart. It will be observed that these two curves are in reasonable concordance, although the operating curve shows values that are somewhat lower. These curves take no account of hydrogen-methane values. The raw gas that is add-

ed not only builds up the carbon monoxide but also produces a methane content which offsets the departure of the operating curve from the theoretical. It is known that methane is a strongly carburizing gas and its use in this connection is a practical expedient for accomplishing the desired result.

36.   By operating in this manner, it has been possible to prevent decarburization during the heating up and holding periods and to prevent oxidization or scale formation during the remainder of the cycle. The results obtained are superior in quality. Temper carbon is found at the extreme edges of the casting. There is a complete absence of pearlite. The complete cycle embraces approximately 48 hr.

37.   The furnace may be charged and discharged in 30 min. There are times, however, when it is not necessary to produce a casting free of decarburization, and in such cases the atmosphere can be modified in such a way as to produce the desired results. This may vary all the way from a thin feather edge to a deep pearlitic rim, in accordance with the purpose for which the work is intended.

THE CONTINUOUS FURNACE

38.   In the continuous type of operation, cold hard iron is introduced at intervals at one end of the furnace and finished annealed castings are removed at the other. As at present developed, there is no dividing partition or wall in these units between the successive zones. Hence the composition of the atmosphere tends to become uniform throughout the whole length of the furnace. This makes it impractical to change the atmosphere in accordance with the requirements of the annealing cycle. A compromise, therefore, must be established between the non-decarburizing atmosphere of the first stage and the non-oxidizing atmosphere of the second stage. The preference is toward slight decarburization during the first which is less injurious than recarburization during the second stage.

39.   It also is necessary to control very accurately the volume of gas entering the furnace. If the composition of the entering gas is changed, it then becomes necessary to change the volume of gas used. This is all determined by its effect upon the metal. Too large a volume of gas or too lean a gas will bring about excessive decarburization and a consequent pearlitic rim. An intermediate flow of gas will not completely avoid decarburi-

zation as the furnaces are operated at present, but will hold it at a satisfactory minimum, which is very much less than that encountered in the usual annealing practice.

40.　Fig. 5 shows a series of gas analyses taken at a number of points along the furnace length. These indicate that the carbon monoxide content is sufficiently high to prevent scaling and oxidization. A slight amount of decarburization, about 0.012-in. does occur. Natural gas addition has been made to this atmosphere, and it has been found that too large a proportion of natural gas will produce an atmosphere too high in hydrogen for
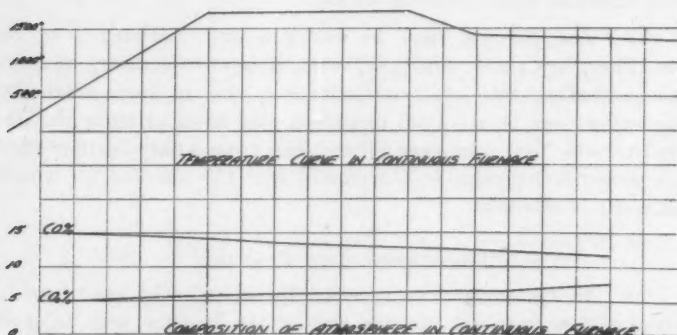
FIG. 5—CARBON DIOXIDE AND MONOXIDE CONTENTS OF CONTROLLED ATMOSPHERE CONTINUOUS FURNACE.

best results. An atmosphere with less than 12 per cent hydrogen seems to work best in a continuous furnace.

41.　From the foregoing facts, it is apparent that the continuous furnace is not as flexible as the intermittent type, from the standpoint of atmosphere control. This fact is offset, however, by the increased heating rate of the continuous furnace and the fact that the furnace structure itself does not have to pass through a series of temperature gradients. Accordingly, the total time for annealing can be reduced in the continuous furnace to somewhere between 30 and 36 hr. for the complete cycle.

RESULTS OBTAINED

42.　The results obtained from the operation of these furnaces has been very gratifying. The castings come from the furnaces free from scale. The presence of temper carbon at the

edge of the casting has improved the machineability of the part, and the time of annealing, in each instance, has been much reduced.

43.   It should be borne in mind that what has been said applies to the manufacture of fully-annealed, blackheart malleable of specification grade. The reduced annealing time does not refer to short cycle pearlitic malleables. For the manufacture of these materials, the same fundamental principles may be applied, with consequent improvement of operation.

DISCUSSION

Presiding: F. L. WOLF, Technical Supt., Ohio Brass Co., Mansfield, Ohio.

H. BORNSTEIN[1]: Sometime ago we installed a one-ton radiant tube malleable annealing furnace, largely for experimental work. We have carried on experiments with temperature cycles and also on various gas mixtures. Some of our results may be of interest in connection with this paper. Refer to our Fig. 6 which gives full information on the temperatures, gas mixture used, and analysis of muffle gas. This is the

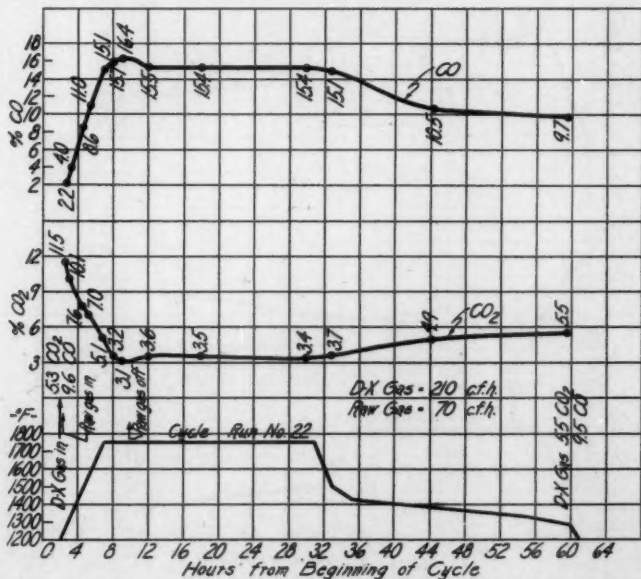[1] Director of Laboratories, Deere & Co., Moline, Ill.

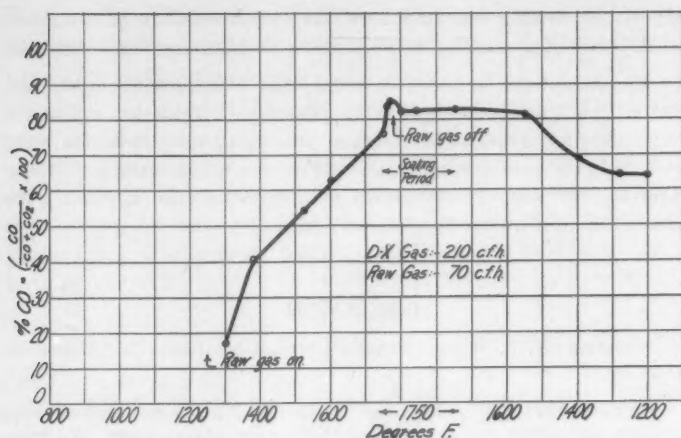FIG. 6—DATA ON MUFFLE GAS OF RADIANT TUBE FURNACE, RUN NO. 22.

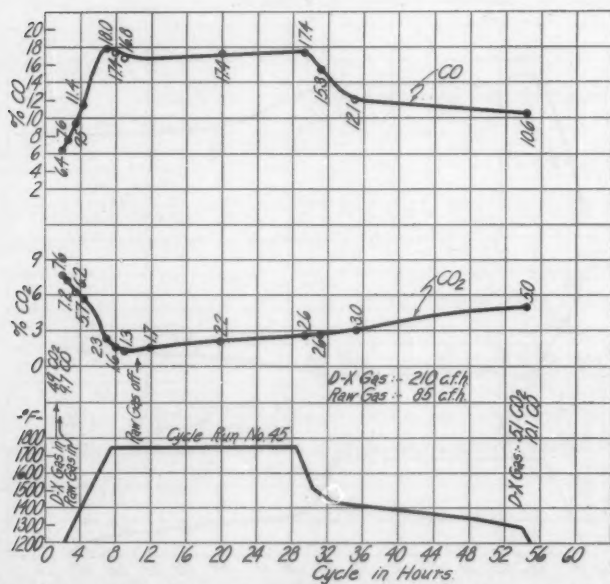Fig. 7—Data of Fig. 6, Replotted According to Method of Fig. 4.

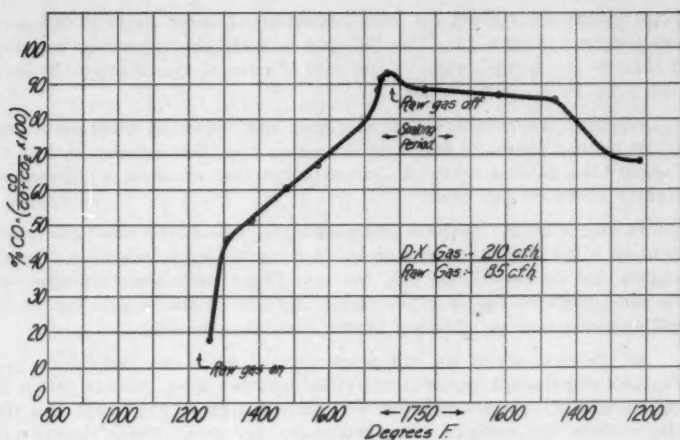Fig. 8—Data on Muffle Gas of Radiant Tube Furnace, Run No. 45, with Increase of Raw Gas Over Run No. 22 of Fig. 6.

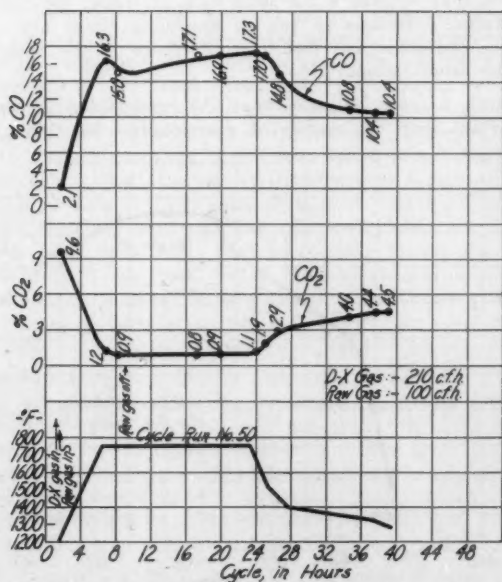FIG. 9—DATA OF FIG. 8, REPLOTTED ACCORDING TO METHOD OF FIG. 4.

FIG. 10—DATA ON MUFFLE GAS OF RADIANT TUBE FURNACE, RUN NO. 50.

mixture that we started out with consisting of three parts of DX gas and one part of raw gas. The DX gas is made by the partial burning of the raw gas using a ratio of one part of natural gas of approximately 1000 b.t.u. to six parts of air.

We took the results shown in Fig. 6 and replotted them according to the method shown in Fig. 4 of the paper and this is shown in Fig. 7. Compare the holding zones. It will be noted that we show a percentage slightly above 80 per cent.

In our Fig. 8, we have increased the amount of raw gas from 70 to 85 cubic feet per hour and by so doing we have increased the CO content and decreased the CO_2 content. These data were replotted on the same basis as Fig. 4 in the paper and are shown in our Fig. 9. It will be noted that we obtain a higher ratio than in Fig. 7.

In the next series we ran practically a two-to-one ratio; raw gas was 100 and the DX gas was 210 cubic feet per hour. This is shown in Fig. 10 and it will be noted that we obtain a higher CO content and the CO_2 content has dropped down below one per cent. These results are replotted in Fig. 11. We want to call particular attention to the curve shown in Fig. 11. It will be noted that it comes pretty close to the theoretical curve shown in Fig. 4 of the paper.

The data which are given here are to amplify the results shown in the paper. Within the range that we have tried these particular atmospheres, that is from a 3-1 ratio to a 2-1 ratio, we could not see any substantial difference so far as annealing times are concerned. Furthermore, we could not find any difference in the results in the matter of decarburization.

Our work to date indicates that the composition is considerably more important than the matter of atmosphere. We think that it is
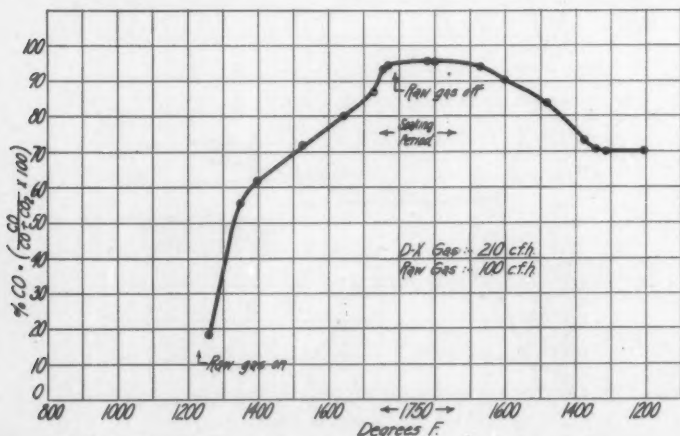

FIG. 11—DATA OF FIG. 10, REPLOTTED ACCORDING TO METHOD OF FIG. 4.

essential to have an atmosphere that will not produce scale. We would like to have an atmosphere which would have a minimum of decarburization.

We find that the silicon content of the iron has a major effect on the annealing cycle. We started out with an iron of approximately one per cent silicon and then increased the silicon to approximately 1.25 per cent and we found a tremendous difference in the response to annealing.

We believe that the paper would be of more value if the authors would add the composition of the irons in both the batch and continuous furnaces. We might also ask the authors as to the effect on the iron of a definitely carburizing atmosphere during the first portion of the cycle. Would we get away from some of the decarburizing that we now have?

MR. DE CORIOLIS: We are somewhat hesitant about appearing before this group and telling them how to malleableize. We are dealing, as I said before in this paper, entirely with the surface of the metal. There is a great deal going on inside of that metal concerning which we do not claim to know as much as you do. The crux of the paper, as I have said before, is to indicate fundamental facts involved and how to proceed to utilize this knowledge in malleableizing. Up to a certain point in the malleableizing cycle, that is to say, before graphitization has set in, you could hardly get an atmosphere which would be carburizing to as much as 2½ to 3 per cent of carbon in the metal. We know the carburizing potential of these gases hardly reaches that point. When carburizing low carbon steel we reach a maximum of 1.40 per cent carbon, and that is our top, and I can hardly conceive of an atmosphere that will proceed to carburize a 2½ per cent carbon iron. So that although such atmsophere as we have indicated here, particularly in the batch anneal furnace, is definitely carburizing to a low or even a medium carbon steel, it certainly would not be carburizing to such a high carbon product as 3 per cent. Beyond the point where graphitization begins, it is my understanding that you do not desire to have a carburizing effect because at that point your carbon begins to leave the surface and to concentrate itself into the nodules of carbon. Were we to carburize the metal at this point, when actually the carbon is trying to reach a certain nucleus, we would simply be adding more carbon to the metal surface which would have to migrate to a nucleus.

So, although I have no definite experience along these lines, I would say there would be no particular advantage, as far as I know, to try and carburize white iron. All you are trying to do is to have an atmosphere which will prevent the carbon from leaving the surface. Otherwise, particularly at high temperature, a low carbon iron would be formed which, according to Schneidewind, would take a longer period of time for the carbon to migrate from the surface into the nucleus of temper carbon.

R. J. COWAN (*Written closure*): We appreciate the contribution of Mr. Bornstein and desire to consider further some of the points he

has raised. First as to the effect of the atmosphere on annealing time. Complete absence of decarburization will result in temper carbon at the extreme edge of the casting. This is probably desirable from the standpoint of machineability of the casting. To obtain this it is necessary to employ a rich atmosphere, one of the 2 to 1 type mentioned by Mr. Bornstein. A small amount of decarburization during first stage annealing would result ordinarily in a white edge but this may be removed by oxidation during second stage annealing, the result being a ferritic edge free from temper carbon. The time required to accomplish either of these results may not be far different but the results will be different and perhaps of different value.

As to metal composition we are glad to include for purposes of record some typical hard iron analyses from both intermittent and continuous type operations:

	Intermittent	Continuous
Carbon (Per Cent)	2.50 to 2.60	2.70 to 2.75
Silicon (Per Cent)	0.90 to 1.00	1.15 to 1.20
Manganese (Per Cent)	0.30 to 0.35	0.38 to 0.42
Sulphur (Per Cent)	0.13 to 0.14	0.09 to 0.11
Phosphorus (Per Cent)	0.13 to 0.14	0.05

A Description of the Age Hardening Process as Applied to Castings

By L. W. Kempf,* Cleveland, Ohio

Introduction

1. Precipitation hardening is a tool so recently made available to the metal industry that the question quite naturally arises as to whether foundries are giving its possibilities sufficient consideration as an aid in widening the field of application of castings. On the assumption that a new process can be sufficiently utilized only if it is thoroughly understood, it is proposed in this paper to consider first the fundamentals of the process in terms of structural and property changes, using for purposes of simplicity illustrations drawn from the field of pure binary alloys or even entirely hypothetical materials. The modifications of the fundamental processes necessary for their application to commercial materials will then be considered together with some alloys at present finding application in the field of castings. The factors of most importance in commercial application of the process will be enumerated, and finally the possibilities for future development will be given attention.

"As Cast" Structure of a Binary Alloy

2. Considering the "as cast" condition of an alloy of relatively pure metal with an element of limited solid solubility in it, a structure is found somewhat as illustrated in the "as cast" micrographs in Fig. 1. There are a number of features which should be noted. First, the metal is divided up into polygonal cells by a network constituent which is revealed by the polishing and etching operations. Also, within the grain, there will be found colonies of a constituent, similar to that occurring in the grain boundary, distributed apparently at random throughout the body of the grain.

3. Closer examination will disclose a pattern in the distribution of this constituent within the grain which, with some imagination, can be considered as resembling a section through a pine tree. The tree-like or dendritic nature of the crystallization

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is more clear from Fig. 2, showing a section of an alloy from which the constituent between the trees, or at the grain boundaries, and between the branches of the trees, or within the grains, has been eaten away by etching with acid.

4. Proceeding from a point in this network of constituent toward the clear, constituent-free areas, a field is traversed which appears to have been darkened and roughened by the etching procedure. This darkened zone gradually grows lighter with increasing distance from the constituent network. If this structure is considered in terms of the mechanism of solidification of an alloy of two phases, its characteristics become explicable.

THE SOLIDIFICATION PROCESS

5. To follow more easily the course of solidification, it may be well to review the constitutional diagram of a hypothetical two phase system as illustrated in Fig. 1. These diagrams are built up by observing the melting and freezing characteristics and microstructure of a large number of alloys with various relative concentrations of the two metals concerned.

6. It should be emphasized that such diagrams, because of the slow rates of solidification used in their determination, represent only approximately the conditions prevailing during the solidification of most commercial castings. This point will be referred to again later. Suppose the concentration of the alloy being considered is given by the position of the line R-S.

7. Solidification of an alloy of this composition starts at a temperature corresponding to point T. The first material to solidify does not have the composition of the hypothetical alloy but has a concentration of element B indicated by the point U. In other words, it will be noted that the first material to solidify contains a far lower concentration of alloying element than is represented by the average composition of the alloy. However, as solidification proceeds, the material solidifying increases in concentration of element B along line C-D; for example, at temperature W the composition of the solid and liquid are represented by X and Y, respectively. At V the remainder of the liquid solidifies by the simultaneous deposition of two constituents of the compositions indicated by D and F.

8. The temperature interval T-V is the solidification range

of the alloy; line *D-E* represents the final solidification, or eutectic, temperature, and the characteristic structure resulting from the simultaneous deposition of two constituents is designated eutectic. After solidification, the structure of the alloy will be somewhat as depicted in the micrographs in Fig. 1 at various magnifications.

9. The solidified metal is built up from materials of two fundamental types. One type forms the continuous background or matrix. The etching characteristics of this constituent varies from zone to zone undoubtedly because of variations in composition. This is the material which solidified throughout the temperature range *T-V*, Fig. 1, the varying composition of which is indicated by the line *U-D*.

10. This constituent is called a solid solution of metal *B* in metal *A*. The atoms of *B* are dispersed as such throughout *A* much as salt is dissolved in water. As might be expected, the effect on the properties of *A* is continuous with changes in the concentration of *B*. Within the solid solubility limits of *B* in *A*
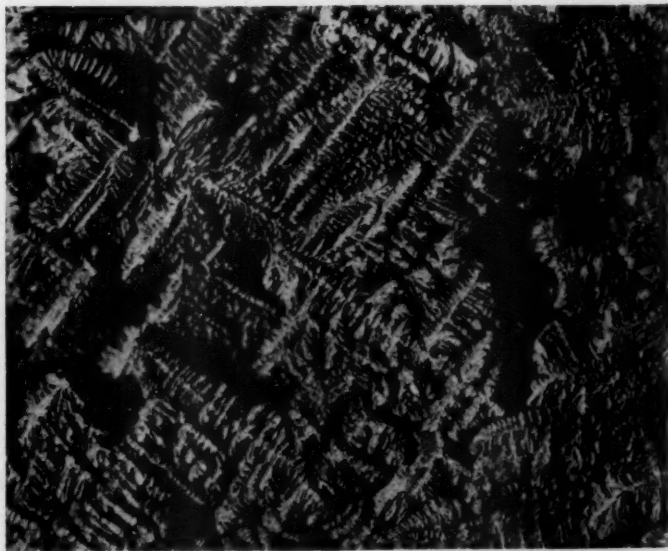

Fig. 2—ILLUSTRATING THE DENDRITIC NATURE OF CRYSTALLIZATION OF METALS.

increasing concentrations of B produce an alloy of gradually increasing tensile strength and decreasing ductility.

11. The other building block in the alloy appears from its etching characteristics to be of uniform and constant composition. In the diagram of Fig. 1, its composition is given by the line G-H. Largely because of its unvarying composition, it is called an intermetallic compound. Usually the relative ratios of A and B entering into the make-up of the compound can be represented by some simple formula such as A_2B. Intermetallic compounds are usually hard and brittle and, as might be expected, tend to impart their characteristics to the alloy as the concentration present in the structure increases.

12. Proceeding outward from the center of the grain or cell, small cores of the relatively dilute solid solution are surrounded by material with an increasing concentration of B, finally merging at the grain boundary with eutectic, of average composition indicated by point E and composed of solid solution with the maximum concentration of B, indicated by D and composed of composition indicated by F. Also, at final solidification the alloy rich liquid between the branches of the "trees" of relatively pure metal freezes in the form of small bits of eutectic. The variations in concentration in the solid solution will tend to smooth out by diffusion during the solidification process. There is, however, an important time factor in the process of diffusion, and it will be readily understood that, if the rate of solidification is sufficiently rapid, these concentration gradients will be more or less mantained to room temperature.

13. Another important diffusion reaction tends to take place during cooling to room temperature after solidification. The course of line D-I indicates that the concentration of B soluble in A decreases with temperature from the concentration corresponding to D at the eutectic temperature to that represented by I at room temperature. This change in solid solubility with temperature is the fundamental basis for precipitation hardening. With the ordinary rates of cooling, however, time is not available for marked changes in structure to occur; therefore, the structure at room temperature is not markedly different from that prevailing immediately after solidification.

14. A phenomenon important in commercial considerations is indicated by the dotted lines of the diagram of Fig. 1. The

solid lines represent the structures which should prevail under
equilibrium conditions or after very slow cooling to any specific
temperature. With cooling rates such as ordinarily prevail in
commercial castings, equilibrium is not attained and the structural
conditions may be more nearly as represented by the dotted lines.
Under-cooling tends to lower the temperature of initial and final
solidification. This results in relatively smaller volumes of solid
solution constituent and larger volumes of intermetallic compound
than would occur under equilibrium conditions and explains the
occurrence of secondary constituents in "as cast" alloys at con-
centrations of added element well within the equilibrium solid
solubility range. Rapid cooling also tends to preserve super-
saturated solid solutions as indicated by the dotted line, thus
accentuating coring.

THE SOLUTION HEAT TREATMENT

15. By holding the casting at, or more conveniently, by
reheating it to a temperature just below the eutectic temperature
for a sufficiently long period of time, it is possible to uniformly
saturate the solid solution with the concentration of B in equilib-
rium in A at the temperature of reheating. This solid solution
containing a relatively high concentration of B may be retained
by rapid cooling to room temperature.

THE PRECIPITATION HEAT TREATMENT

16. Because this is a supersaturated structure, there will be
a tendency for the alloy constituent or compound to precipitate.
This tendency in most commercial alloys is not very marked at
room temperature. In a few, however, particularly those in which
the base metal is of relatively low melting point, such for example
as lead or tin and even in some specific alloys of aluminum, the
progress of the reaction at room temperature is of commercial
importance. The tendency for the precipitation to occur will in-
crease rapidly with increase in temperature, and even those alloys
which show no evidence of precipitation at room temperature
will show definite effects at some temperature between ordinary
and those indicated by point K.

17. It might be worth while to consider in more detail the
structural changes accompanying precipitation at somewhat ele-
vated temperatures. The changes in structure on reheating for

extended periods of time to temperatures within the range indicated by K on the diagram are illustrated in Fig. 1. Unfortunately the structural changes usually occurring under conditions of time and temperature which give maximum tensile properties are quite obscure and difficult to illustrate in a general way.

18.   The micrographs of Fig. 1 illustrating precipitate structures are typical of alloys in which the time of treatment was so extended or the temperature so high that maximum mechanical properties are not obtained. The first structural evidence of precipitation is usually a change in etching characteristics; grain boundaries become more distinct and apparently wider; the color contrast between grains usually decreases and the solid solution grains seem to etch more easily, thus appearing rougher and darker. As the time increases at a specific aging temperature, small particles of precipitate may be identified in the grain boundaries. The tendency for precipitation to proceed at a relatively fast rate in the grain boundaries in some instances has an important bearing on the resistance of the alloy to corrosion in specific environments.

19.   It may be stated as a generality that tensile and yield strengths vary directly and elongation inversely with hardness. The usual course of hardness changes with variations in the conditions of precipitation is illustrated by the curves of Fig. 3.
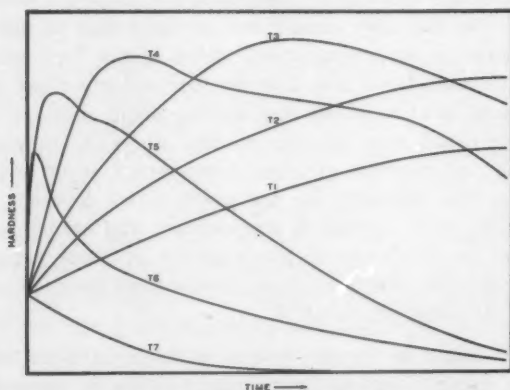

FIG. 3—TYPICAL AGE-HARDENING CURVES. INCREASING TEMPERATURE IS INDICATED BY THE INCREASING VALUE OF THE SUBSCRIPTS.

In Fig. 3 the temperature of treatment is to be assumed to in-
crease with the value of the subscript used for identifying the
curves.

20. As the temperature of reheating is increased, the rate
of change of hardness also increases rapidly. A certain minimum
temperature is necessary for the attainment of maximum hard-
ness, and at this temperature hardness increases at first quite
rapidly, the rate then decreases, a maximum hardness is reached,
and with increase in the time periods the hardness begins to
decrease at first quite rapidly, becomes fairly constant for quite
a period of time, and then decreases relatively rapid. These
general characteristics are probably present over quite a wide
range of temperature of reheating, but the change in rate with
temperature may completely suppress some of the details at
extremely low or high temperatures of reheating. It is important
to note that for maximum hardness a certain specific time period
at a specific temperature is necessary. Changes in time and tem-
perature from this optimum usually result in some sacrifice in
hardness.

MODIFICATIONS OF THE GENERAL TYPE OF PRECIPITATION HARDENING

21. It probably would be well to consider here two modifica-
tions of precipitation hardening which are commercially im-
portant. Referring again to the as-cast structure of Fig. 1, con-
siderable areas of supersaturated solid solution will be noted as
indicated by the dark etching portions. It is to be expected that
precipitation could be brought about with consequent change in
properties in these zones by reheating the as-cast material to
temperatures within the precipitation range. The curves of Fig. 4
illustrate the changes in hardness on reheating chill castings of
an aluminum alloy containing about 10 per cent copper. It will
be noted that even without the application of a high temperature
solution or homogenizing treatment, appreciable hardness increases
are brought about by precipitation treatments. Naturally these
effects will be the more accentuated the more rapid the cooling of
the casting.

22. Another modification of the precipitation hardening
process is the hardening which occurs in steel by rapid cooling
through the eutectoid transformation temperature. In this case

the solubility of carbon in iron changes abruptly at the eutectoid point because of the change in crystal structure of the iron. Large structural variations in the precipitate may be brought about by varying the conditions under which precipitation occurs on passing from gamma to alpha iron with consequent great changes in properties. A similar change in properties occurs in the beta alloy of copper with zinc or aluminum. This, however, is such a specific type of precipitation hardening that attention will be concentrated for this discussion on the more general type previously described.
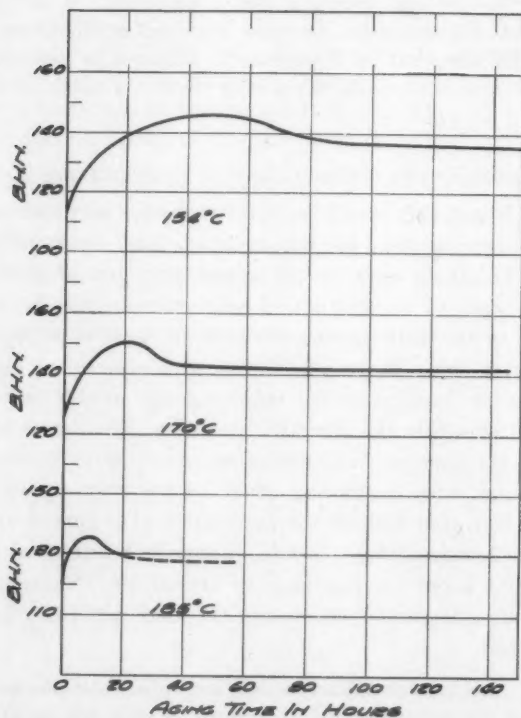

Fig. 4—The Course of Age-Hardening in an Aluminum Alloy Containing 10 Per Cent Copper in the "As-Cast" Condition.

COMMERCIAL APPLICATIONS OF PRECIPITATION HARDENING

23. Considerable commercial application has been made of precipitation hardening particularly in non-ferrous fields, and even more specifically in aluminum alloys. The change in solubility of copper in aluminum as indicated in Fig. 5 is the basis for a casting alloy containing about 4 per cent copper. Typical properties of this alloy in various conditions of heat treatment are given in Table 1.

24. The solid solubility of magnesium in aluminum varies from about 15 per cent at the eutectic temperature of 844°F. (450°C.) to less than 3 per cent at room temperature. Alloys containing more than about 6 per cent magnesium are quite amenable to heat treatment as indicated by the curves of Figs. 6 and 7 giving the properties of ½-in. diameter sand-cast test bars in the as-cast and heat-treated conditions.[1*]

25. Magnesium and silicon when present in aluminum in the ratio indicated by the formula Mg_2Si act to some extent like a single metallic element. The solubility of the compound in solid aluminum varies from 1.85 per cent at 1103°F. (595°C.) to less than 0.3 per cent at room temperature. Alloys containing only Mg_2Si within the limits soluble in solid aluminum are not easy to cast. By increasing the silicon content to 5-8 per cent, casting properties are markedly improved without greatly affecting the amenability of any Mg_2Si present to heat treatment. The properties of alloys containing various concentrations of silicon and magnesium with a solution treatment only and with a solution treatment followed by a precipitation treatment in Fig. 8.[2,3]

26. In the copper-base alloys, chromium shows solid solubility in copper varying from about 0.8 per cent at 1830°F. (1000°C.) to about 0.05 per cent at room temperature. Castings of an alloy containing about 0.5 per cent chromium, when of thin section and rapidly cooled to room temperature after pouring, may have their hardness increased from about 50 Brinell in the as-cast condition to about 100 by a precipitation treatment consisting of 2 hr. at 932°F. (500°C.). In such a condition, tensile strength will be about 50,000 lb. per sq. in. with 25 per cent elongation and an electrical conductivity of 80 per cent.[4] The solid solubility of the elements nickel, tin, zinc, and aluminum in copper does not greatly vary with temperature. It has been found, how-

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* Superscripts refer to references at the end of the paper.

Table 1

TYPICAL PROPERTIES[1] OF A COMMERCIAL HEAT TREATED ALUMINUM ALLOY CONTAINING ABOUT 4 PER CENT COPPER

Condition	Commercial Designation	Yield[2] Strength, lb./sq. in.	Tensile Strength, lb./sq. in.	Elongation Per Cent in 2 in.	Brinell Hardness, 500Kg. 10mm.
1. As Cast		10,500	22,000	4.0	53
2. Heat treated near the eutectic temperature and quenched....195-T4		16,000	31,000	8.5	65
3. Heat treated as in 2 and aged at about 300°F. 195-T6		22,000	36,000	5.0	80
4. Treated as in 3 but aged for a more extended period 195-T62		31,000	40,000	2.0	93

[1] Determined from standard half-in. diameter tensile test specimens individually cast in green sand molds and tested without machining off the surface.

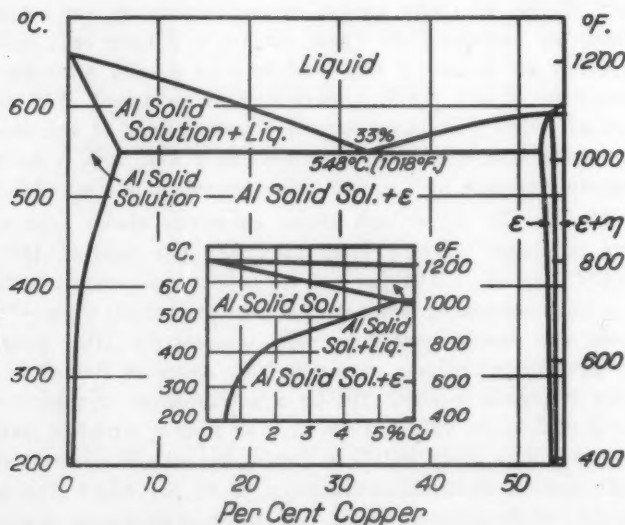[2] Stress which produces a permanent set of 0.2 per cent of the initial gauge length.

FIG. 5—CONSTITUTIONAL DIAGRAM OF ALUMINUM COPPER ALLOYS. THE SOLUBILITY OF COPPER IN SOLID ALUMINUM VARIES FROM ABOUT 5.6 PER CENT AT 1080°F. TO LESS THAN ABOUT 0.5 PER CENT AT ROOM TEMPERATURE
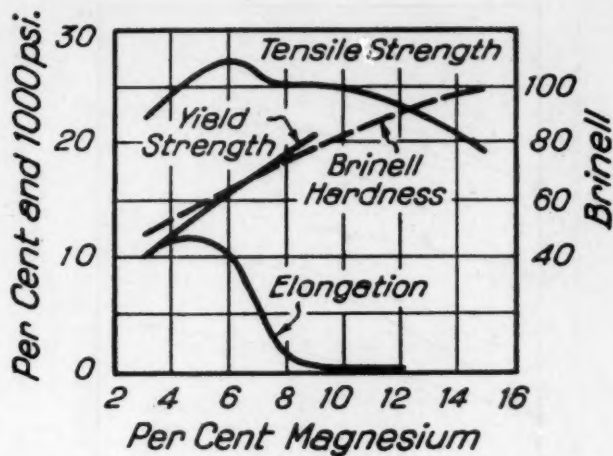
Fig. 6—Properties of Aluminum-Magnesium Alloys in the "As-Cast" Condition.
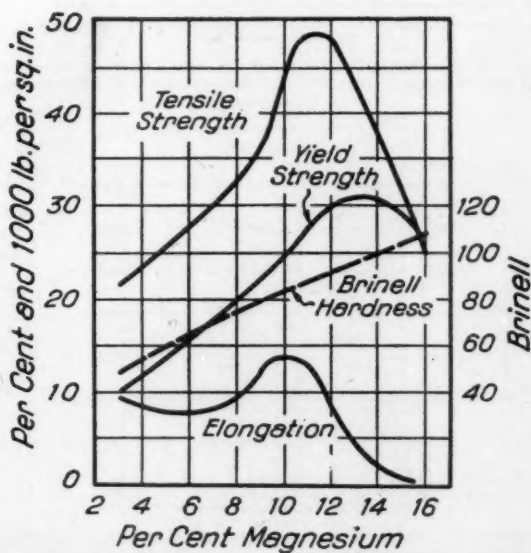
Fig. 7—Properties of Aluminum-Magnesium Alloys After Solution Heat Treatment.
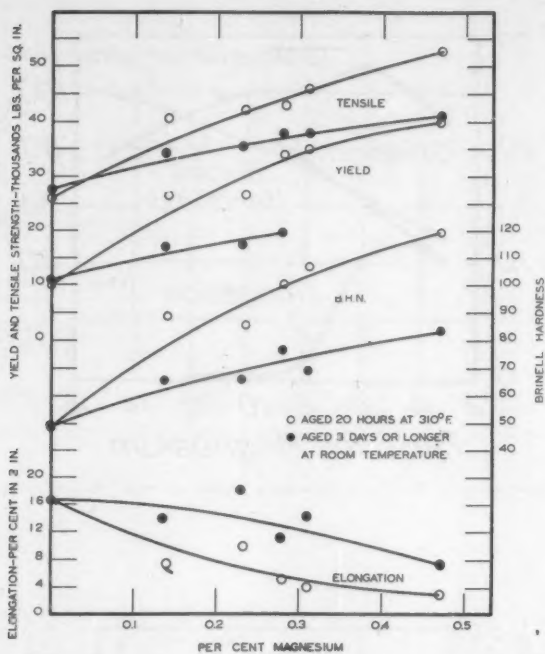
FIG. 8—EFFECT OF MAGNESIUM CONCENTRATION ON THE PROPERTIES OF AN ALUMINUM-10 PER CENT SILICON ALLOY IN THE SOLUTION HEAT-TREATED AND SOLUTION HEAT-TREATED AND AGED CONDITIONS.

FIG. 9—CONSTITUTIONAL DIAGRAM OF COPPER-BERYLLIUM ALLOYS. THE SOLUBILITY OF BERYLLIUM IN SOLID COPPER VARIES FROM ABOUT 2.4 PER CENT AT 1598°F. TO LESS THAN 0.5 PER CENT AT ORDINARY TEMPERATURES.

FIG. 10—EFFECT OF NICKEL ON THE SOLUBILITY OF TIN IN SOLID COPPER.

FIG. 11—EFFECT OF TIME AND TEMPERATURE ON THE HARDENING OF AN ALLOY OF 7.5
PER CENT NICKEL, 8 PER CENT TIN, BALANCE COPPER.

Fig. 12—Effect of Time at 1400°F. on the Properties of a Sand Cast 7.5 Per Cent
Nickel, 8 Per Cent Tin, Balance Copper Alloy Subsequently Quenched and Aged
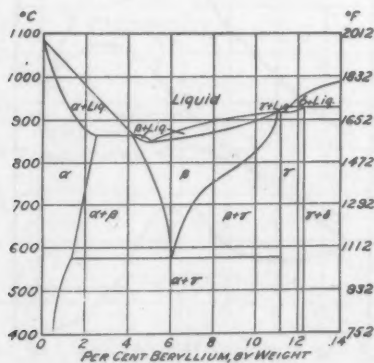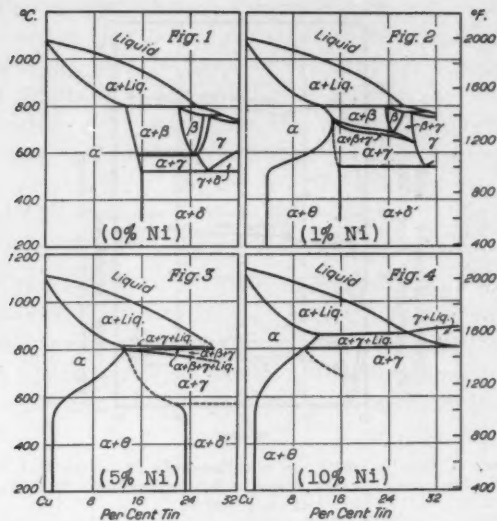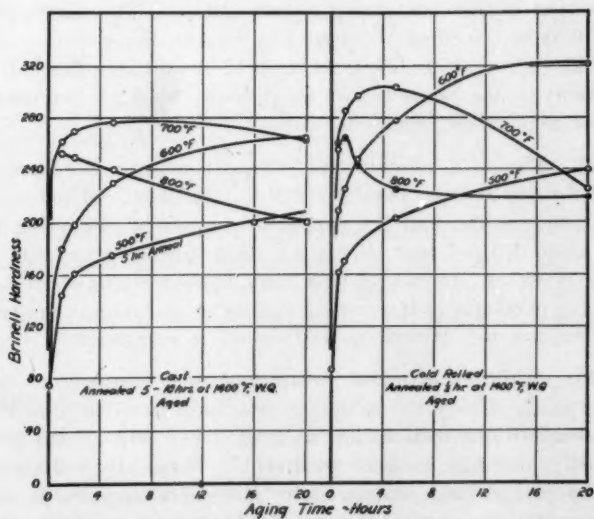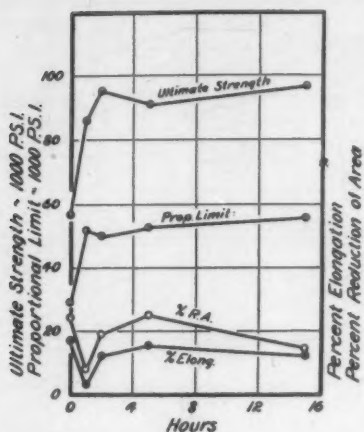5 Hours at 500°F.

ever, that alloys containing combinations of these elements are
affected markedly by heat treatment. Perhaps one of the most
interesting is that containing nickel and tin. The diagrams[5] of
Fig. 10 show the effect of nickel additions on the solubility of tin
in solid copper while Figs. 11 and 12 show the effect of heat
treatment on the properties of an alloy containing 8 per cent tin,
7.5 per cent nickel, remainder copper.[6]

27. The solid solubility of silicon in copper decreases from
about 6.8 per cent at 1472°F. (800°C.) to about 3.0 per cent at
room temperature, and the copper-silicon alloys containing more
than about 3.0 per cent silicon are quite amenable to heat treat-
ment. However, alloys with such high concentrations of silicon do
not have particularly interesting casting or mechanical properties,
and thus are not utilized to any extent in commerce.

28. Additions of nickel, cobalt, and other elements to copper-
silicon alloys change the solubility relationships to such an extent
that compositions with as low as 2 or 3 per cent silicon become
decidedly amenable to heat treatment[7]. These alloys apparently
are not yet utilized commercially, but there appears to be no
fundamental reason for their not having specific fields of ap-
plication.

29. Copper-base alloys commercially utilized in the heat-treated condition include one containing about 2 per cent zinc, 3 per cent nickel and 10 per cent tin. Another contains about 4 per cent aluminum, 4 per cent nickel, and 0.8 per cent silicon. The application to the latter alloy of a direct precipitation treatment, that is, without a preliminary solution treatment and quench, of about 2 hr. at 887°F. (475°C.) will increase the yield strength of small castings from 50,000 lb. per sq. in. to 65,000 lb. per sq. in. and increase the Brinell hardness from 140 to 190.

30. The solubility of beryllium in solid copper decreases from about 2.4 per cent at about 1598°F. (870°C.) to less than 0.5 per cent at room temperature. There has been some small commercial application of cast beryllium bronze[8]. The effect of heat treatment on alloys containing about 1 to 3 per cent beryllium is indicated in Table 2.

Table 2
AVERAGE PROPERTIES OF BERYLLIUM-COPPER CASTINGS

Composition	Elastic Limit, lb./sq. in.	Tensile Strength, lb./sq. in.	Elongation, Per Cent	Hardness, Brinell
As Cast:				
1.79% Be	42,600	61,800	7.0	110
1.86% Be	33,400	41,500	4.0	136
2.19% Be	44,800	60,500	10.0	120
Heat treated 2 hr. at 1472°F. (800°C.) quenched:				
1.67% Be	16,810	41,100	29.0	54
1.79% Be	15,800	33,400	16.0	65
2.19% Be	22,500	40,000	12.0	110
Aged at 572°F. (300°C.) for 3 hr.:				
1.79% Be	39,000	71,100	3.0	119
2.19% Be	86,200	110,000	1.0	375
2.45% Be	95,600	117,000	1.0	415

31. Another commercial copper-base alloy utilized in the heat-treated condition is aluminum-bronze which may contain about 8-12 per cent Al and 1-3 per cent Fe. The amenability of these alloys to heat treatment is based on changes in solubility at the eutectoid inversion.

Iron-Base Alloys

32. In the iron-base alloys, the variation in properties possible through variations in structure related to the allotropic change, and its accompanying change in solubility of carbon is so great that precipitation hardening of the more general type is not very widely used in commercial alloys. Copper, however, exhibits

a wide change in solubility in solid alpha iron as indicated by Fig 13[9], and this change in solubility can be utilized in increasing the properties of malleable iron and steel castings. The maximum effect is obtained by reheating the casting to temperatures just below the eutectoid temperature, rapidly cooling to room temperature, and reheating in the neighborhood of 932°F. (500°C.). Changes in properties accompanying such heat treatments are illustrated in Table 3[10]. The ordinary rates of cooling from the
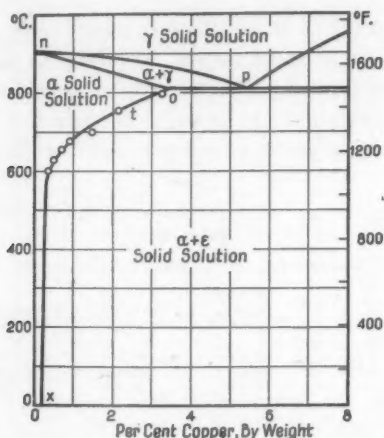
FIG. 13—THE SOLID SOLUBILITY OF COPPER IN ALPHA IRON.

Table 3

MECHANICAL PROPERTIES OF PRECIPITATION HARDENED MALLEABLE CAST IRON

Cu., Per Cent	Tensile Strength, lb./sq. in.	Yield Strength, lb./sq. in.	Elongation, Per Cent in 2 in.	
0.031	60,500	38,000	14.5	Annealed
	63,000	44,000	17.7	Aircooled 1350-1375°F.
	62,000	41,000	16.3	Reheated 930°F.
0.63	58,500	42,000	22.2	Annealed
	61,000	45,500	23.7	Aircooled 1350-1375°F.
	60,500	44,000	24.5	Reheated 930°F.
1.30	58,500	44,000	15.5	Annealed
	60,000	48,000	15.7	Aircooled 1350-1375°F.
	65,500	54,000	14.0	Reheated 930°F.
2.01	59,500	45,000	18.5	Annealed
	62,000	50,000	18.5	Aircooled 1350-1375°F.
	70,000	59,000	15.2	Reheated 930°F.

Carbon, 2.3 per cent; Si., 1 per cent; Mn., 0.3 per cent; P., 0.12 per cent.

malleablizing cycle are also sufficient to retain a certain degree
of supersaturation in irons containing 1.0 per cent or more of
copper and, by merely reheating in the neighborhood of 932°F.
(500°C.), appreciable changes in strength can be obtained.

33. The hardness of cast iron, containing upwards of 1.0
per cent copper can also be increased by reheating to temperatures
in the neighborhood of 500°C. This effect is, however, not great
and is not utilized to any extent in commercial practice.

34. The new iron-base alloys containing about 25 per cent
cobalt and 7-15 per cent aluminum are quite susceptible to heat
treatment, and, in fact, develop their optimum magnetic proper-
ties only after specific solution and precipitation treatments.

35. Probably at least 85 per cent of the magnesium alloy
castings produced in this country are heat treated. Most commer-
cial alloys contain aluminum as a major constituent which is
soluble to the extent of 12 per cent at the eutectic temperature of
815°F. (435°C.) and decreases with temperature to less than about
3 per cent at room temperature. The curves of Fig. 14 show the
effect of heat treatment on the properties of sand casting with
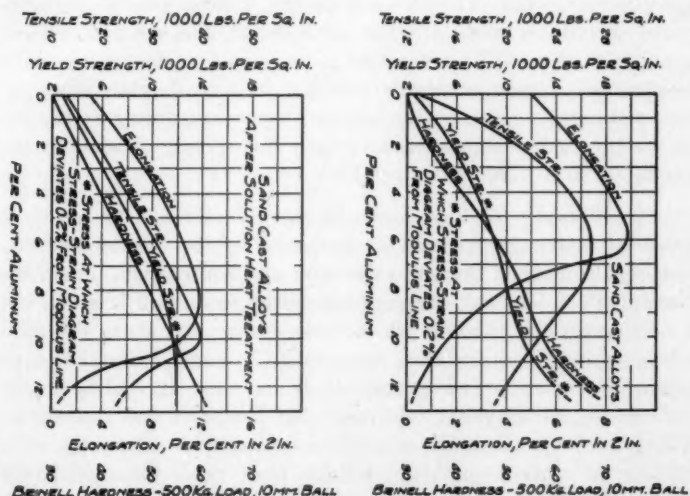various aluminum concentrations[11].

Fig. 14—Properties of Magnesium-Aluminum Alloys in the "As-Cast" Condition
(Above) and after a Solution Heat Treatment (Below).

COMMERCIAL CONSIDERATIONS

36. Perhaps the most important commercial consideration in the application of precipitation hardening to any specific alloy is the choice of time and temperature for the solution and precipitation treatments. Fortunately, the approximate maximum temperature for the solution treatment is one of the few things of practical use directly obtainable from an equilibrium diagram. The eutectic temperature indicated in the diagrams for the binary alloys concerned may be taken as the maximum possible solution temperature. This may be modified downward by the presence of impurities or additional elements.

37. By means of a few experiments, the optimum solution heat-treatment temperature can be readily determined. For maximum results, the time at temperature must be sufficient to saturate the solid solution. It will vary directly with the grain size of the casting, and thus fine structures are most desirable for heat treating.

38. It should be noted that the speed of diffusion increases very rapidly with temperature; for example, an increase of about 18°F. (10°C.) will, in some aluminum alloys, reduce by about 50 per cent the time necessary for complete solution of the intermetallic compound. For this reason the highest possible temperature should be used. On the other hand, the strength of the metal at the elevated temperature is an important factor inasmuch as the casting must in general support its own weight. Some assistance may be given by the use of proper supporting racks, but in complicated castings, the strength factor may determine the maximum heat-treating temperature.

39. Rapidly cooling any metal structure tends to set up a balanced system of internal stress caused by the plastic deformation resulting from thermal expansion and contraction. From the standpoint of internal stresses, it is better to quench from as low a temperature as possible, which means from just above the solid solubility line for the alloy concerned. In some cases this may mean heating at a temperature nearer the eutectic, cooling in the furnace to just above the solid solubility line, and then quenching. The rate of quench must be sufficiently rapid to retain the solid solution to room temperature, but the more rapid the quench, the higher the internal stresses and, therefore, the rate of cooling should be no more drastic than necessary. In some materials the

corrosion resistance is markedly affected by the rate of cooling and this must be considered in a choice of the proper heat-treating conditions.

40. It probably will be generally necessary to determine the optimum time and temperature for reheating for the precipitation treatment by means of cut and try experiments.

41. The starting point for such experiments may usually be obtained from the solid solubility curve of the principal element in the base metal. Because of the variation in cooling rate of various section thicknesses and also of structural features such as grain size, the rate of precipitation will vary in various parts of the casting and perhaps from casting to casting, particularly when the structure is being aged directly in the as cast condition. For these reasons it is generally desirable to age to maximum hardness or, if a hardness somewhat lower than the maximum is desired, to obtain this by over-aging rather than attempting to stop the aging treatment before the maximum is reached. The reason for this is found in the contour of the aging curve.

42. It will be noted that the rate of change of hardness up to the maximum is relatively rapid but that at or near the maximum hardness there is considerable range of time in which the hardness does not change appreciably. Also, beyond the maximum hardness, the rate of change of hardness with time is relatively slow, permitting the attainment of a uniform product in spite of inherent differences in aging rates. For the same reason aging periods of too short duration should be avoided inasmuch as the shorter the aging period, the more difficult it is to control the properties of the casting.

43. Occasionally the small density changes which accompany the changes in structure involved in solution treatment and precipitation become of importance commercially. These changes are usually small—never more than perhaps 2 or 3 per cent, but if they become of importance, a compromise must usually be effected between the maximum mechanical properties and maximum stability inasmuch as the maxima in the various properties, for example hardness and density, do not usually occur under the same conditions of time and temperature. Also, other properties such as resistance to corrosion or electrical conductivity do not usually vary at the same rates during the precipitation process and the specific application will dictate the compromise to be made.

Possibilities for Future Developments

44. The light alloy casting industry has obviously made relatively greater use of solution heat treatment and precipitation hardening processes than the other non-ferrous or ferrous industries. In the non-ferrous field this is perhaps not so much because of an inherent lack of possibilities in the alloy materials available as of a general lack of appreciation of those possibilities. In the field of ferrous castings the availability of a powerful tool for the modification of properties in the allotropic transformation of iron naturally retards the application of what appear at the present to be less effective methods of hardening.

45. It should be emphasized that the heat-treating possibilities of alloys of a specific base metal are not exhausted by a consideration of the solubility relationships in binary alloys alone. Small concentrations of additional elements often have a profound effect on solid solution ranges of binary alloys. For example, the addition of 0.1 per cent magnesium to an aluminum-base alloy containing 4 per cent copper changes its precipitation hardening characteristics under some conditions by a factor of two or more. In the opposite direction, relatively small concentrations of iron (in the tenths of a per cent) may greatly reduce the improvement in properties possible by the heat treatment of the same 4 per cent aluminum-copper alloy.

46. It is probably in the direction of small additions to the commonly used alloys that the development of heat-treatable alloys will proceed. In this field of complex alloys, no, or very few, guides are available at present for the indication of the most promising directions for investigation. Procedures must be quite empirical based principally on the experience and judgment of the operator with the greatest discoveries, all other conditions being equal, rewarding the one who investigates the greatest number of combinations.

REFERENCES

1—Kempf and Keller, *Nat'l Met. Handbook*, A.S.M., 1936, p. 942.
2—Archer and Kempf, Trans. A.I.M.E., Vol. 93, p. 448, (1931).
3—Kempf and Rowe, *Nat'l Met. Handbook*, A.S.M., 1936 Ed., p. 970.
4—Neave, Proc. Inst. of Auto Eng., Vol. 31, p. 647 (1937).
5—Eash, *Nat'l Met. Handbook*, A.S.M., 1936, p. 1053.

6—Wise and Eash, TRANS. A.I.M.E., Vol. III, p. 218 (1934).
7—Corson, PROC. INST. MET. DIV., A.I.M.E. 1927, p. 435.
8—Cone, *Foundry*, Vol. 61, No. 9, p. 20 (1933).
9—Gregg and Daniloff, *Nat'l Met. Handbook*, A.S.M., 1936, p. 262.
10—Lorig and Smith, TRANS. A.F.A., Vol. 42, p. 211 (1934).
11—Wood, *Nat'l Met. Handbook*, A.S.M., 1936 Ed., p. 1210.

DISCUSSION

Presiding: H. M. ST. JOHN, Detroit Lubricator Co., Detroit.

CHAIRMAN ST. JOHN: For many years it was well understood one
of the great advantages of iron-carbon alloys, in other words, steel over
the non-ferrous alloys, is the fact steel can be hardened and strengthened
by heat treatment. It was assumed that was the inherent difference
and nothing in particular could be done about it. More recently, some-
thing has been done about it, and the development which Mr. Kempf
has described in his paper this morning is perhaps the most important
one that has taken place in non-ferrous metallurgy for many years.
It is one which in the copper base alloys particularly no great advantage
has yet been taken, but the field is there, it is open. This disadvantage
of non-ferrous alloys which has existed all these years is disappearing.
It only waits for the metallurgist and the foundryman to take full ad-
vantage of the possibilities.

E. W. SAWYER[1]: One of the most difficult fields is probably the
brass alloys, and I think Mr. Kempf is to be congratulated on the
admirable restraint which he has shown in his treatment, which has
allowed him to cover his new element duralumin in which we are in-
terested. The diagram which he showed, the molybdenum diagram,
coupled with his remarks, that copper in general does not dissolve the
precipitation agent, very greatly interested me. I believe the diagram
should show quite a bit more solubility for duralumin at the higher
temperatures. Perhaps that is the reason why duralumin precipitated in
copper shows higher tensile properties than for other copper alloys.
In heat treated conditions, 150,000 pounds is not unusual.

Furthermore, Mr. Kempf's remarks that additions still further
general improvement in properties is borne out in this case also by the
addition of nickel, I believe one per cent heat treated beryllium-copper
will have tensile strengths as high as 180,000 pounds. It has been
interesting to me to see these predictions of Mr. Kempf's so carried
out in this instance.

C. H. LORIG[2]: Mr. Kempf, is it possible by means of micro-
structures to determine the extent of precipitation hardening? In other
words, is the precipitate always sub-microscopic and, therefore, im-
possible to show except, perhaps, by a shading in the micro-structure?
I wonder if there is not some way in which the micro-structure can

[1] Thompson Product Co., Cleveland, Ohio.
[2] Battelle Memorial Institute, Columbus, O.

serve in showing whether precipitation hardening is complete or in showing whether the alloy has received a precipitation hardening treatment.

MR. KEMPF: I think Dr. Sawyer's remarks need no comment.

With regard to Dr. Lorig's question, generally it is not possible to follow a precipitation hardening process by changes in micro-structure. I am referring now to the precipitation process itself. Changes in solid solubility can usually be followed microscopically, and changing solid solubility is usually a pretty good indication of whether subsequent reheating to a low temperature will bring about precipitation hardening.

In following the process of precipitation hardening, a good artisan at micro-structural analysis may be able to see structural changes in time periods shorter than required for maximum tensile strength, but this is not the usual case. It is only under conditions of time and temperature markedly exceeding those required for maximum mechanical properties that individual particles of precipitate can be identified.

Production and Properties of Age Hardenable Five Per Cent Nickel-Bronze Castings

By T. E. Kihlgren*, Bayonne, N. J.

Abstract

The properties and foundry production of an age hardenable nickel-bronze mixture of moderate cost and broad utility are presented. The mixture possesses physical properties somewhat superior to those normally obtained for 88-10-2 bronze in the "as cast" condition, and in addition, can be age hardened to secure the following range of properties: 70,000-90,000 lb. per sq. in. tensile strength, 50,000-74,000 lb. per sq. in. yield point (0.5 per cent extension under load), 25 to 5 per cent elongation, and 140-200 BHN. The effect of modifications of the normally employed heat treatment are discussed. Lead is a highly objectionable impurity and as little as 0.05 per cent will seriously retard age hardening and also cause hot cracking if the castings are rapidly cooled from the 1400° F. "solution treatment." Silicon retards aging, but the effect of small amounts of silicon (0.05-0.10 per cent) can be neutralized to a considerable extent by a moderate increase in the aging temperature and in the nickel content (to 5.5 per cent). The effects of other impurities are briefly discussed. The foundry procedure recommended for castings to be age hardened differs from that considered good practice in 88-10-2 bronze chiefly in the oxidation of the copper-nickel melt with cuprous oxide (or other suitable oxides) and a subsequent reduction with phosphorus, prior to the addition of remelt, and the introduction of tin and zinc. Where heat treatment is not contemplated, or maximum response to heat treatment is not required, the oxidation treatment may be omitted, and melting merely carried out in a moderately oxidizing atmosphere. The oxidation treatment, however, has been found to definitely increase the response to age hardening and to be generally helpful.

Introduction

1. The modification of commonly used bronze mixtures with small amounts of nickel (generally of the order of 0.5 to 2.0 per cent), to secure useful improvements in casting and physical properties has been practiced by many foundrymen for a number of years and is well recognized.

2. Somewhat less well known are the nickel-tin bronzes containing 5 per cent or more of nickel which may be age hardened by suitable heat treatment to develop strengths ranging from 75,000 to 110,000 lb. per sq. in., depending upon the composition and heat treatment employed.

3. The age hardenability of these alloys was reported in 1928 by Price, Grant and Phillips[1]* and discussed by Wise[2] at that time. A comprehensive treatment of the strength and aging characteristics of nickel-tin bronzes† containing up to 15 per cent of tin and 20 per cent of nickel, both cast and wrought, was presented by Wise and Eash[3] in 1934 and further data were presented by Fetz.[4] The equilibrium relationships existing in the copper corner of the copper-nickel-tin system have been carefully investigated by Eash and Upthegrove[5] and by Veszelka.[6]

4. It is not proposed, in this paper, to discuss the characteristics of the nickel-tin bronzes as a group, as these have been covered in the paper by Wise and Eash.[3] It is rather our purpose to concentrate on a single mixture of low cost and of broad utility, not only as cast but also in the heat treated condition, and which can be produced in any well conducted foundry. This casting mixture has a composition of 88 per cent copper, 5 per cent nickel, 5 per cent tin and 2 per cent zinc. This alloy will be referred to subsequently as 88-5-5-2 nickel-bronze and is essentially an 88-10-2 bronze with half of the tin replaced by an equal amount of nickel.

PROPERTIES OF 88-5-5-2 NICKEL-BRONZE—"AS CAST"

5. The average tensile properties listed in Table 1 are based on sand cast bars cast in the pattern indicated in Fig. 1 from 150-175 lb. oil-fired crucible and 30-40 lb. induction furnace virgin metal melts, over a range of temperature of 2200-2350°F.

Table 1

PROPERTIES OF 88-5-5-2 NICKEL BRONZE—"AS CAST"
Tensile Strength (lb./sq. in.) 48,500
Yield Point (0.5 per cent extension under load)
   (lb./sq. in.) 24,400
Elongation (Per Cent in 2 in.) 42
BHN (1000 Kg.) 86

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† U. S. P. 1,816,509, and U. S. P. 1,928,747 (E. M. Wise).
* Superior numbers refer to bibliography at end of paper.

6. The properties given in Table 2 were obtained on bars from two 100 per cent remelt heats using gates and risers from an oil-fired 150 lb. crucible heat, simply remelted in a clay graphite crucible, treated with 0.02 per cent phosphorus and cast:

Table 2

88-5-5-2 NICKEL-BRONZE—"AS CAST"—(100 PER CENT REMELT)

	Heat 1	Heat 2
Tensile Strength (lb./sq. in.)	48,750	48,500
Yield Point (0.5 per cent extension under load) (lb./sq. in.)	23,250	23,500
Proportional Limit (lb./sq. in.)	16,000	16,000
Elongation (Per Cent)	41	37

7. It is apparent that the 88-5-5-2 bronze yields properties somewhat superior to those normally obtained on the 88-10-2 bronze and its use as an alternative to the latter for general purpose castings merits consideration.

PHYSICAL PROPERTIES OF 88-5-5-2 BRONZE—HEAT TREATED

8. Much higher properties than those just indicated can be secured by applying a suitable heat treatment to the 88-5-5-2
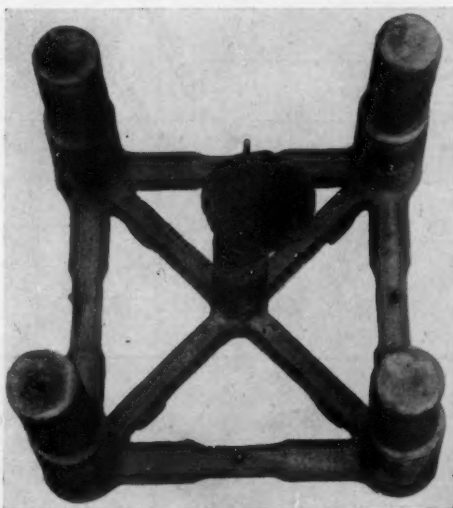
FIG. 1—TENSILE TEST CASTING

bronze. The properties of Table 3 can be obtained by appropriate heat treatment.

Table 3

PHYSICAL PROPERTIES OF 88-5-5-2 BRONZE—HEAT TREATED

Tensile Strength (lb./sq. in.) 70,000-90,000
Yield Point (0.5 per cent extension
   under load) (lb./sq. in.) 50,000-74,000
Proportional Limit (lb./sq. in.) 35,000-58,000
Elongation (Per Cent) 25-5
BHN 140-200

9. The heat treatment normally applied is that common to all age-hardening alloys and comprises a solution treatment (or

FIG. 2—PROPERTIES OF AGE HARDENED 88-5-5-2 NICKEL-BRONZE

homogenizing anneal) at an elevated temperature (after which
the bronze is in a "dead soft" condition) followed by an aging
treatment at a lower temperature for a suitable period rendering
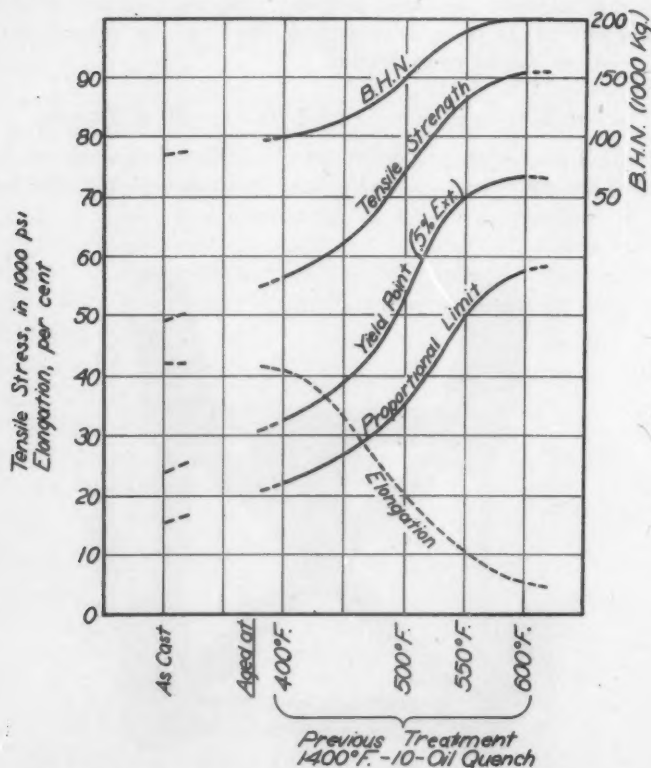the alloy hard, the temperature and time of the aging treatment
determining the properties finally secured. Various modifications
of the heat treatment are possible, such as air cooling the casting
after the high temperature anneal instead of quenching which
yields a slightly different combination of properties. These modi-
fications will be discussed in a later section.

10. Fig. 2 shows the physical properties of the 88-5-5-2 alloy
both in the "as cast" and heat treated conditions. The 550°F.
aging treatment gives an excellent combination of high strength
coupled with adequate ductility. The average properties observed
are noted in Table 4:

Table 4

AVERAGE PROPERTIES OF 88-5-5-2 BRONZE—HEAT TREATED AT 550°F.

Proportional Limit (lb./sq. in.) 50,000
Yield Point (0.5 per cent extension
　　under load) (lb./sq. in.) 70,000
Tensile Strength (lb./sq. in.) 87,000
Elongation (Per Cent) 10
BHN 196

11. Further increase in elastic properties can be obtained by
using a higher aging temperature, although at the expense of a
decrease in ductility.

12. While the data on which Fig. 2 is based were obtained
on bars given a 10 hr. anneal at 1400°F. prior to aging, a 5 hr.
anneal is adequate and little is gained by increasing the time of
the solution treatment.

Relationship of Physical Properties

13. Fig. 3 shows the relationship between tensile strength
and elongation, as well as proportional limit, resulting from vari-
ous treatments. These relations are useful to engineers in securing
a proper balance between strength, elastic properties and ductility.

Electrical Resistivity

14. Inasmuch as the 88-5-5-2 alloy may be found useful in
the electrical field as a non-magnetic material of fair conductivity
and high strength, some information as to the electrical resistivity
(in microhms-cms.) has been obtained. Fig. 4 shows the resistivity
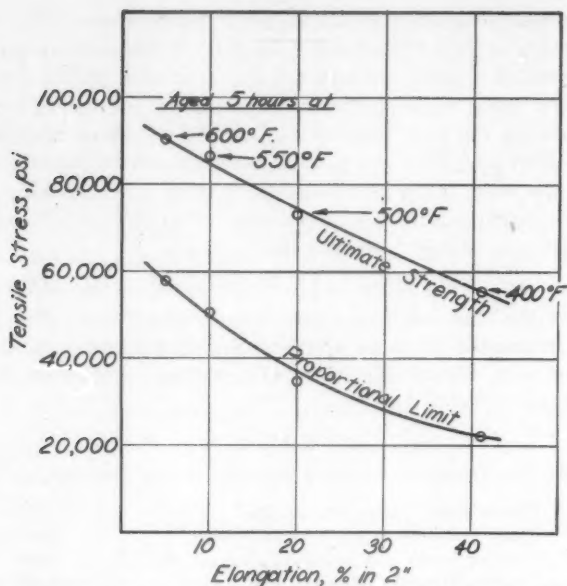
FIG. 3—RELATIONSHIP OF ULTIMATE STRENGTH AND PROPORTIONAL LIMIT TO ELONGATION
OF AGE HARDENED 88-5-5-2 BRONZE
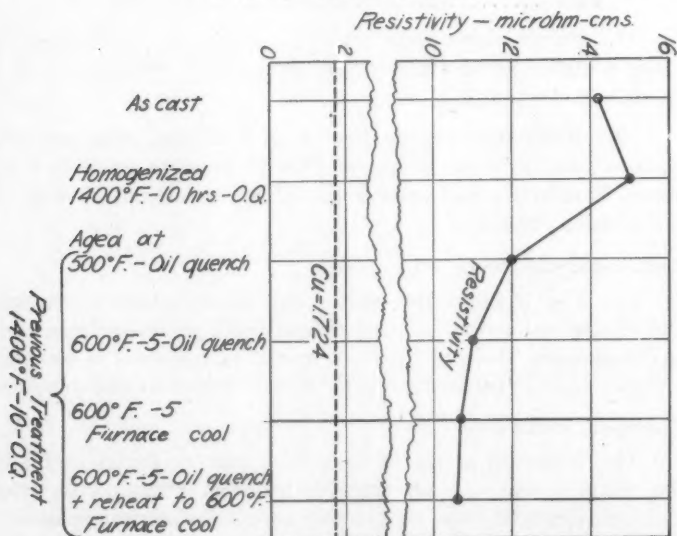
FIG. 4—ELECTRICAL RESISTIVITY OF 88-5-5-2 BRONZE AS CAST AND AGE HARDENED

of the 88-5-5-2 bronze in various stages of heat treatment. As
might be expected, the resistivity is highest in the fully homo-
genized or softened condition and is lower, the more vigorous the
age hardening treatment. Generally speaking, the resistivity will
range from 10 to 12 microhms-cms.

Causes of Failure to Age Harden

15.   Failure of the material to age harden to the same degree
as indicated in Fig. 2 for a given aging temperature may be
due to either lack of soundness, the presence of certain impurities,
or to a deficiency in tin or nickel. If, however, the elongation is
normal with respect to the tensile strength, as indicated by com-
parison with Fig. 3, the castings are no doubt satisfactorily sound

FIG. 5—EFFECT OF MODIFICATIONS OF NORMAL AGE HARDENING HEAT
TREATMENT OF 88-5-5-2

and a moderate increase in aging temperature will secure the desired properties.

MODIFICATIONS OF HEAT TREATMENT

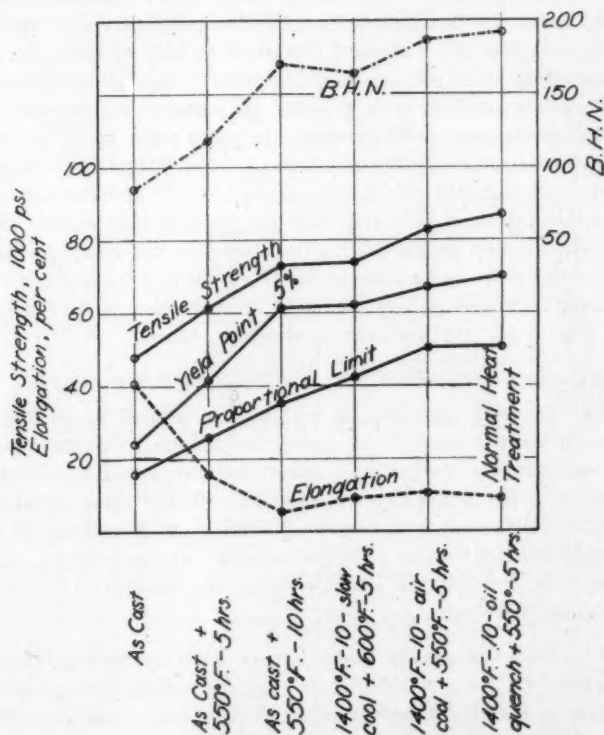16. To obtain maximum elastic and strength properties, a solution treatment at approximately 1400°F., with a fairly rapid cooling therefrom and subsequently aging at a lower temperature is generally necessary. However, there are a wide variety of applications where maximum properties are not required and with this in mind, a variety of heat treatments have been tried, the results of which are shown in Fig. 5, and are discussed below.

Age Hardening by Direct Aging of Castings

17. The properties of castings of moderate section comparable to that of a tensile test bar can be quite sharply improved by simply aging the "as cast" material at 550°F. for 5 hr. and oil quenching (air cooling would probably be still better). By this simple treatment, it is possible to increase the proportional limit of tensile bars by 60 per cent, the yield point by 75 per cent, the tensile strength by 30 per cent and the Brinell hardness by 40 per cent, and still retain adequate ductility. It is obvious that for castings of moderate size, this simple and inexpensive treatment will develop properties highly attractive for many purposes. By increasing the aging time at 550°F. to 10 hr., further increases in strength properties and hardness are obtained as may be seen from Fig. 5 but the ductility is sharply reduced.

Variation in Cooling Rate from the Solution Temperature

18. In some instances, it will be impractical to quench the casting after the solution or aging treatments. The fourth and fifth sets of points of Fig. 5 show the properties obtained on tensile bars by annealing at 1400°F.—10 hr.—and cooling in Sil-O-Cel (to simulate the rate of cooling of a casting of considerably heavier section than the tensile bar) or cooling in air, respectively, followed by suitable aging treatment for 5 hr., and air cooling therefrom.

19. The bar *slowly cooled* from 1400°F. responded only moderately to aging at 550°F. but by increasing the aging temperature to 600°F., it developed properties quite similar to those obtained on the bars *quenched* from 1400°F. and subsequently aged at 500°F. (Fig. 2). The specimen *air cooled* from 1400°F.

followed by the usual 550°F. treatment developed properties only
slightly inferior to the similarly aged bars oil quenched from
1400°F. and then aged (shown in the final set of points).

20. Apparently, the rate of cooling following the solution
or aging treatments is not critical. Castings of heavy section, air
cooled from 1400°F., might be expected to require somewhat
higher aging temperatures (600°F.) for effective hardening.
Castings corresponding to the tensile bar or lighter section, *air
cooled* from 1400°F., should have properties quite comparable to
those obtained by *oil quenching* from 1400°F. before aging.

Effect of Variations in Time of Solution Treatment on Properties after Aging

21. Fig. 6 shows the properties of bars given a solution
treatment at 1400°F. on 2, 5 and 10 hr., respectively, and then
aged at 550°F. for 5 hr. It will be seen that the 2-hr. solution
treatment developed properties after aging quite comparable to
the 10-hr. treatment. At aging temperatures above 550°F.
(e.g. 600°F.), a longer prior solution treatment appears necessary
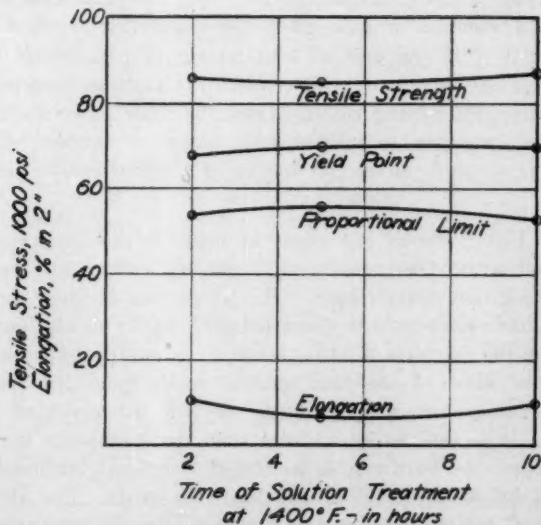and 5 hr. is suggested to cover the range of aging temperatures
apt to be used.

Fig. 6—Effect of Varying Time of Solution Treatment on Properties after a
Subsequent Aging at 550° F.

EFFECT OF IMPURITIES

22. When the 88-5-5-2 alloy is to be used in the "as cast" condition, control of impurities is apparently no more critical than for the 88-10-2 composition. However, when the castings are to be age hardened, control of composition is very important if suitable properties are to be secured on heat treatment.

Lead

23. As far as the "as cast" properties are concerned, the effect of lead is to lower the strength and elastic properties moderately, and the ductility quite vigorously, although even at one per cent lead, the properties (Table 5) are still rather good.

Table 5

EFFECT OF LEAD ON 88-5-5-2—AS CAST

Lead %	Tensile Strength lb./sq. in.	Elongation Per Cent in 2 in.	BHN 1000 Kg. Load
0.00	47,150	45	92
0.25	42,900	22	89
0.50	37,800	16	78
1.00	39,850	16	75

24. When 88-5-5-2 castings, however, are to be heat treated, very small amounts of lead affect the properties drastically. As little as 0.05.-0.10 per cent of lead retards or completely inhibits response to age hardening and renders the castings susceptible to hot cracking, on cooling rapidly from the 1400°F. solution treatment, due probably to molten lead being in contact with the thermally stressed metal producing a "solder embrittlement" effect.

25. Fig. 7 shows the effect of lead on the properties "as cast" and after heat treatment involving quenching from the 1400°F. solution temperature. The properties of the "as cast" material have already been discussed. The highly deleterious effect of lead on the response to heat treatment is readily apparent from the graph. Most of the lead content melts quenched from the 1400°F. homogenizing temperature showed hot cracking in the fracture. It is interesting to note that, by increasing the aging temperature, the hardness is increased somewhat, although it is still well below the level of the lead-free melts. The attendant presence of molten lead at the higher aging temperatures is no doubt undesirable as in other alloys.

Table 6

EFFECT OF LEAD ON TENSILE PROPERTIES USING MODIFIED
HEAT TREATMENTS

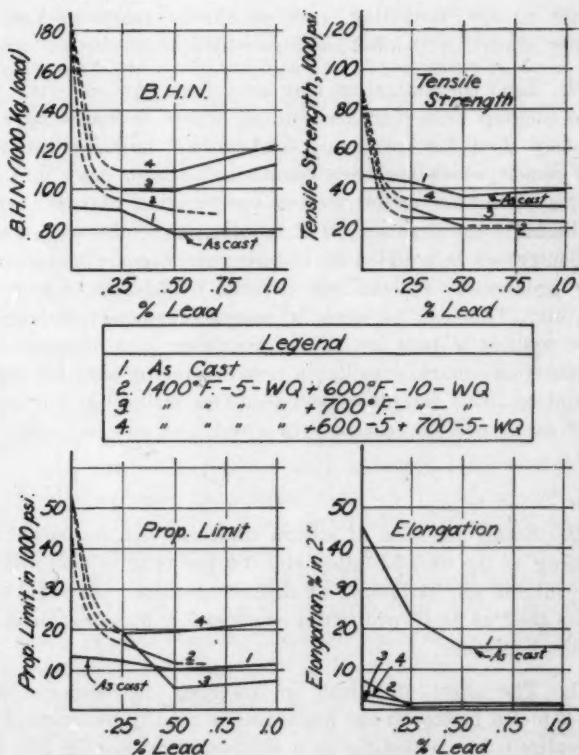|		Tensile Strength lb./sq. in.			Elongation in 2 in. Per Cent		
Heat Treatment	0 % Pb	0.15 % Pb	0.50 % Pb	0 % Pb	0.15 % Pb	0.50 % Pb
As Cast + 550° F.— 10 hrs.—air cool	72,450	62,750	58,650	2.5	9	13
1400° F.—10 hrs.—air cool + 550° F.—5 hrs.—air cool	79,100	55,200	51,200	7	22	15
1400° F.—10 hrs.—cool in Sil-o-cel + 600° F. —5 hrs.—air cool	74,000	59,800	59,500	9	7	9

FIG. 7—EFFECT OF LEAD ON THE RESPONSE OF 88-5-5-2 BRONZE TO AGE HARDENING

26. As might be expected, hot cracking, due to the "solder embrittlement" effect in lead content alloys, can be fairly well avoided by slow cooling of the casting from the 1400°F. solution treatment, or by simply aging the casting directly, without homogenization, but the response to age hardening is still unsatisfactory as may be seen from Table 6.

27. The reason for the inhibiting effect of lead on the response to age hardening is not yet understood and requires further investigation. Certain elements have shown some indication of partially neutralizing the deleterious effects of traces of lead but these effects are not as yet sufficiently well established to merit discussion.

28. It is obvious from the above that to insure suitable response to age hardening, care should be taken to keep lead virtually absent in 88-5-5-2 castings which are to be heat treated.

29. Lead contamination may arise from raw materials, or be due to pick-up from furnace linings, where furnaces have been previously used for producing lead-content mixtures, or to the use of remelt which has been contaminated with gates and risers from leaded bronze melts. Proper specification of raw materials will eliminate the first source of trouble, while the second source of difficulty can be avoided by re-lining the furnace and reserving it for production of lead-free bronzes 88-5-5-2, 88-8-4, 88-10-2, 89-11, etc. This can be done if several furnaces are available. Where melting is done in unlined crucibles, little chance of lead contamination occurs, provided a new crucible is used for the first melt and set aside for subsequent lead-free melts. Careful segregation of scrap will be necessary to avoid lead contamination from remelt.

Silicon

30. Small amounts of silicon diminish the response to age hardening of the 88-5-5-2 alloy and 0.2 per cent silicon will practically inhibit age hardening of this composition, although it does increase the "as cast" properties considerably over the phosphorus deoxidized mixture.

31. The effect of silicon in lowering the response of the 88-5-5-2 nickel bronze to age hardening is readily understood when it is realized that nickel forms a silicide, thus making less nickel available to form the compound responsible for age hardening of

the copper-nickel-tin alloys. By increasing the nickel content by
an amount sufficient to unite with the silicon present, leaving at
least 5 per cent of nickel for the normal age hardening, one might
expect that the modified mixture would then age harden similarly
to the silicon-free 88-5-5-2 composition. This actually does occur,
and by raising the nickel to 6 per cent, the alloy with 0.2 per cent
silicon aged was found to harden to a considerable degree on
aging at 600°F., although the combination of properties obtained
were still somewhat below those of the silicon-free, 5 per cent
nickel alloy deoxidized with phosphorus. An increase in the aging
temperature of the silicon-modified mixture to 650°F. is also help-
ful in raising the general level of properties.

32. Where contamination with silicon is difficult to avoid, it
may be found desirable to add an excess of 0.5 per cent of nickel,
over the customary 5 per cent, as the extra one-half per cent will
be sufficient to take care of 0.05 to 0.10 per cent of silicon. It will
be found generally desirable, however, to attempt to keep silicon
as low as possible preferably below 0.05 per cent.

Other Impurities

33. Small amounts of manganese are not objectionable, up
to 0.25 per cent having been satisfactorily added, but under the
melting procedure recommended for the 88-5-5-2 bronze this
element is not a necessary addition. Zinc can be reduced consider-
ably without difficulty, good results having been obtained with
0.75 per cent zinc, and also in the absence of zinc, although in the
latter instance, the phosphorus addition should be increased to
about 0.15 per cent.

34. Wise and Eash[3] reported the effect of various impurities
on the aging response of the 7.5 per cent nickel — 8 per cent tin
mixture and their findings probably apply to some extent to the
88-5-5-2 alloy as well. They found that 2 per cent of iron, though
it retarded the aging response, nevertheless developed very good
strength properties and a reasonable amount is probably tolerable
in the 88-5-5-2 bronze. Wise and Eash[3] also found that increasing
the zinc to 10 per cent greatly retarded the aging response and
small amounts of chromium (0.15 per cent) were also deleterious
in this respect.

35. While the effects of some of the impurities which retard
aging response can be partially offset by changing the aging tem-
perature (usually by an increase of 50 to 100°F.), it is de-

sirable to avoid complications of this sort by suitable control of composition.

36.   Reasonable adherence to the melting procedure outlined in the following section, should yield castings quite low in total impurity content which will consistently develop properties on aging quite comparable to those described in this paper.

FOUNDRY PRACTICE

37.   When the 88-5-5-2 nickel-bronze is to be used simply in the "as cast" condition, the foundry practice considered desirable for 88-10-2, i. e., melting under moderately oxidizing conditions followed by treatment with 0.05 per cent phosphorus will be found satisfactory under most melting conditions and moderate amounts of such impurities as lead, silicon (though they should not be co-present), etc., are no more objectionable than in the 88-10-2 bronze.

38.   However, where advantage is to be taken of the age hardening characteristics of the alloy, further precautions in melting are required if good tensile properties, particularly good ductility, are to be obtained after heat treatment. The principal deviation from the normal 88-10-2 melting practice consists of an oxidation treatment of the copper-nickel melt with copper oxide or other suitable oxides such as those of manganese and nickel prior to further additions, and an increase of about 100°F. in pouring temperature. Greater care in choice of raw materials must also be exercised as contamination with lead must be avoided and other impurities will alter the aging response to some extent. The foundry practice to be described will deal with production of castings to be age hardened.

39.   It should be appreciated that each foundry represents a different problem due to differences in the melting equipment used, the raw material available, and the type of castings produced. While the melting procedure herein described has been found to work well in high frequency induction furnace and 150-175 lb. oil-fired crucible melts, some modifications may be found necessary to meet the conditions existent in a particular foundry. This paper is intended to aid the foundryman in arriving at the optimum procedure for the particular conditions existing in his own plant.

Raw Materials

40.   *Copper.* Most of the data in the present paper were obtained from melts made from electrolytic copper or from a high

grade of remelted electrolytic copper of about 99.95 per cent purity. The 150 lb. foundry heats were all from electrolytic copper.

41. *Nickel.* While 2:1 shot and 50:50 nickel-copper shot, provided they are low in silicon, sulphur, and free of lead, may be used successfully, complete insurance against contamination with undesirable impurities may be obtained by the use of electrolytic nickel (one in. squares are convenient to use), and the latter is a more economical method of adding the nickel where efficient melting equipment is available.

42. *Zinc.* A low lead content zinc must be used and zinc of a grade meeting A. S. T. M. specification No. 1A, serial designation B-6-33 will be found very satisfactory ("special high grade" — 0.007 per cent maximum lead).

43. *Tin.* A good commercial grade of "pure bar tin" should be used with lead 0.02 per cent maximum and preferably well under that figure.

44. *Remelt.* Clean gates and risers from melts made from the above grades of nickel, copper, zinc and tin can be included in the charge, and 40 to 50 per cent can be readily handled. As a matter of fact, 100 per cent remelt heats can be quite successfully used, providing the gates and risers are from melts made from virgin metals of the above grades. In the latter instance, oxidation treatment is, of course, undesirable. The use of oily turnings should not be attempted, it being desirable to dispose of this type of scrap in some other way.

45. *Melting Equipment.* All of the melts discussed in this paper are from oil-fired 150 lb. crucible heats, using a low sulphur content fuel oil, or from 30-40 lb. induction furnace heats. However, there is no reason why efficient gas-fired crucible furnaces (low sulphur gas) will not also function satisfactorily. Rocking type indirect arc furnaces and tilting oil-fired furnaces of the Swartz type can also be used. However, in the latter two instances, it should be emphasized that furnaces of these types with permanent linings must be used for lead-free melts only. Any attempt to use such furnaces for both the 88-5-5-2 alloy and for leaded bronzes will result in lead contamination of the former and destroy its ability to age harden.

46. *Cover.* Excellent results have been obtained on melts made without a cover, but generally a moderate charcoal cover is

added after the copper and nickel have begun to melt down and
prior to deliberate oxidation of the melt and the addition of zinc
and tin. Glass covers have also been used but are not necessary.
In some instances,[8] crucibles have been covered by an inverted
old crucible bottom previously used for lead-free melts during
melting.

47. *Molding Sands.* The molding sands should be quite open
and, in general, a permeability of about 30 should be used for
moderate size castings. In working out the melting procedure and
in determining the strength and aging characteristics of the
bronzes reported herein, a synthetic sand of about 60-80 perme-
ability and 4.5 per cent moisture was used.

48. The molds were faced with graphite, dusted on lightly
and brushed off, and in most instances have given a satisfactory
cast surface. The sand employed was more open than is necessary
and a sand of 30 permeability is suggested. It is, of course, realized
that the type of sand to be used will vary with the class of work.

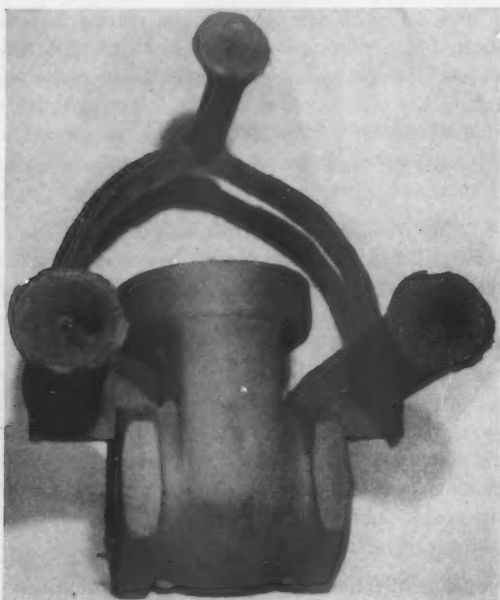49. Where the cast surface must be very smooth, a tighter

FIG. 8—VALVE BODY CAST IN 88-5-5-2 NICKEL-BRONZE (POURING TEMPERATURE 2275°F.)

facing sand will have to be used and the less satisfactory venting properties of the closer sand must be supplemented by more vigorous venting of the molds.

50. For pressure castings of moderate size, it is generally preferable to sacrifice a bit in cast surface smoothness and to use a more open sand and thereby decrease the per cent of rejects due to porosity. We believe this statement applies to a considerable gamut of bronze mixtures.

51. *Gating and Heading.* There are no specific methods to insist upon, as each type of casting requires individual consideration. The important thing to observe is that the risers be properly located on the pattern so that they will function to *feed* rather than *bleed* the casting. The general principles described in a previous paper[7] will be found to work well. This type of feeding, involving the use of gate risers which receive the hottest metal last, requires a good self-venting sand or generous venting of the mold as the air in the mold cavity must be ejected through the whistler vents and the sand itself, but is an efficient method of feeding the casting and is reliable.

52. A common foundry error is to attempt to feed a heavy section with a riser separated from the heavy section by a constriction. No matter how large the riser is, it will be unable to do a good job. A little advance consideration of the casting design will help to avoid errors of this sort.

53. Fig. 8 shows a valve body, tested at 1000 lb. per sq. in. hydraulic pressure, cast in the high permeability synthetic sand. Sections made through critical areas showed no signs of shrinkage.

54. *Melting Procedure.* The following procedure has been found to work well in oil-fired crucible furnace heats from virgin metals:

    (1) Melt down copper and nickel using a moderately oxidizing flame and when last of copper is charged in the melt-down, add moderate charcoal cover and continue melting.

    (2) When copper and nickel are melted, oxidize heat with 3 oz. of cuprous oxide to 100 lb. of melt, stirring in oxide with an iron rod. Hold heat 3 to 5 min.

    (3) Reduce melt with 3 oz. of 15 per cent phosphor copper per 100 lb. melt.

    (4) Add zinc, and shortly afterward add tin and bal-

ance of zinc. (A portion of zinc is added first to complete the reduction and improve the tin recoveries.)

(5) Add 4 oz. of phosphor copper per 100 lb. melt, pull crucible and pour castings.

55. When the charge is to consist of 50 per cent remelt, balance new metals, the procedure will be modified as follows:

(1) Melt down copper and nickel as before.

(2) Oxidize with 1.5 oz. of cuprous oxide per 100 lb. of charge.

(3) Reduce with 1.5 oz. of 15 per cent phosphor copper per 100 lb. of charge.

(4) Add remelt gates and risers.

(5) When molten, add zinc and tin to obtain the desired tin and zinc contents in the finished casting. It is not necessary to add zinc for volaitilization losses but about 0.10 to 0.20 per cent excess tin may be added to make up for tin losses.

(6) Add 3 oz. of phosphor copper per 100 lb. charge just prior to drawing the crucible.

56. Under some conditions of melting using remelt, it may be found that step (2) can be omitted without seriously impairing the results. The oxidation treatment will not be found detrimental, however, and, in general, the oxidized melts respond more vigorously to age hardening.

Acknowledgments

57. The author desires to express his thanks to A. J. Wadhams, manager, Development and Research Department, International Nickel Company, for permission to present this paper, and to N. B. Pilling and E. M. Wise, manager and assistant manager, respectively, Research Laboratory, International Nickel Company, for their encouragement and helpful criticisms. Sincere thanks are also due W. R. Hempstead for his considerable part in carrying out the experimental foundry and laboratory tests reported herein.

BIBLIOGRAPHY

1—Price, W. B., Grant, C. G., and Phillips, A. J., *"Alpha Phase Boundary of the Copper Nickel Tin System,"* Trans. A.I.M.E. (1928), vol. 78, Inst. of Met. Div., pp. 511-513.

2—Wise, E. M., Trans. A.I.M.E. (1928), vol. 78, pp. 514-517.

3—Wise, E. M., and Eash, J. T., *"Strength and Aging Characteristics of the Nickel Bronzes,"* TRANS. A.I.M.E., Inst. of Met. Div., vol. 111, pp. 218, 1934.

4—Fetz, E., *"Hardenable Bronzes with Cu-Ni-Sn Basis, I—Alloys for Sand Castings,"* KORROSION U. METALLSCHUTZ, Vol. 11, No. 10, pp. 217-229. *"II—Workable Alloys,"* METALLWIRTSCHAFT 1936, Vol. 15, No. 7, pp. 167-170, No. 8, pp. 189-191. *"VI—Addition of work Hardening and Precipitation Hardening,"* Z. PHYSIK, Vol. 97, pp. 690-698, 1935. *"IV—Influence of Precipitation Hardening on Corrosion Resistance,"* KORROSION U. METALLSCHUTZ, Vol. 11, pp. 110-107, 1935.

5—Eash, J. T., and Upthegrove, C., *"The Copper Rich Alloys of the Copper Nickel Tin System,"* TRANS. A.I.M.E. (1933), Vol. 104, pp. 221-249.

6—Veszelka, J., *"Investigation of Equilibrium Relationships of Further Alloyed Bronzes,"* Com. Mining & Metallurgy Dept., Royal Hungarian Mining & Forestry College, Sopron, Hungary (1932), Vol. 4.

7—Kihlgren, T. E., *"Production of Pressure Tight Castings in 30 Per Cent Cupro Nickel,"* TRANS. A.F.A. 1937, Vol. 45, p. 225.

8—Hudson, F., *"Modern Non-Ferrous Castings and Their Engineering Interest,"* Institute of Engineers & Shipbuilders, Scotland, Paper No. 982 (Proof). See also METAL INDUSTRY (London), Vol. 52, No. 5.

DISCUSSION

Presiding: H. M. ST. JOHN, Detroit Lubricator Co., Detroit, Mich.

CHAIRMAN ST. JOHN: I am inclined to believe that the expert foundryman of a few years ago would have viewed with some amazement the physical properties given for the alloy described in this paper. It is another evidence of the remarkable development that was explained in the previous paper by Mr. Kempf.

WM. M. BALL[1]: The 88-5-5-2 alloy mentioned by Mr. Kihlgren is of interest to me. I would like to know what is a safe working temperature for this alloy and if a temperature of 450 to 500°F. would injure the aging characteristics?

MR. KIHLGREN: That is a good question to ask, and unfortunately, it is one we cannot give a very complete answer to right now. I think it is well to be cautious in attempting to use any of the age hardening alloys at temperatures closely approaching their optimum aging temperature. I hesitate to indicate the temperature to which it would be desirable to use this material but off-hand, I would think that up to around 450°F. might be all right. However, until we have some concrete data that will answer that question, I would rather not try to give any definite figure.

GEO. P. HALLIWELL[2] *(Submitted as Written Discussion):* This paper deals with a class of alloys, which because of their amenability

to heat-treatment, makes them available for a wide variety of applications. The data presented by Mr. Kihlgren constitute a welcome and necessary contribution to our technical information.

The effect of lead upon the age hardening properties is of interest since many foundrymen desire its presence not only to improve machinability, but to increase pressure tightness.

Recent results obtained in our laboratory, however, indicate that lead is not as detrimental to the age hardening properties of these alloys as indicated in this paper. In substantiation of this statement we submit the data shown in Table 7. Test bars were cut from a keel block casting, similar to that used by Roast, given a solution treatment at 1400°F. for 10 hours, water quenched, and subsequently aged at 600-650°F.

Table 7

Alloy	No. 1	No. 2	No. 3
Copper	82.04	82.46	82.65
Tin	8.35	7.33	7.26
Nickel	7.48	7.50	7.44
Lead	0.03	0.14	0.30

Aged 600°F. — 3 Hours

	No. 1	No. 2	No. 3
		83000 to	
Tensile Strength (lb. Per Sq. in.)		87000	75000
Elongation (Per Cent)		1	2
BHN		230	197

Aged 600°F. — 7 Hours

	No. 1	No. 2	No. 3
	90000 to	87000 to	
Tensile (lb. Per Sq. in.)	108000	93000	80000
Elongation (Per Cent)	1	1	1
BHN	280	246	215

Although these results show a decrease in properties with an increase in lead content, it is by no means as marked as indicated by Mr. Kihlgren's results. Could this be due to the higher nickel and tin contents of our alloys?

The statement that 0.05 per cent lead seriously retards age hardening requires a little clarification. Does Mr. Kihlgren mean that 0.05 per cent lead simply retards age hardening, and that the alloy will attain full hardness upon continued aging, or does he mean that the potential age hardening possibilities are permanently decreased? The curves showing the effect of lead are indicated as dotted lines between 0 and 0.25 per cent lead, indicating that no data was available between these two concentrations. In the light of our results it would be interesting if Mr. Kihlgren could show quantitative results at lower lead concentrations.

MR. KIHLGREN: Mr. Halliwell's test results on the effect of 0.03 to 0.30 per cent of lead on the 7.5 per cent nickel — 8 per cent tin mixture are interesting and indicate that a considerable degree of age hardening has occurred. Nevertheless, these data do demonstrate quite

effectively that lead is an undesirable element. Most of the results reported by Mr. Halliwell show elongations of one per cent, hardly a useful degree of ductility. Wise and Eash, in a previous paper before the American Institute of Mining and Metallurgical Engineers, to which reference is made in the text, develop the relationship between tensile strength and elongation for the 7½-8 per cent alloy (deoxidized with 0.05 to 0.010 per cent barium), and their data are shown in Fig. 9, the left hand graph being the one of immediate interest.

If the data shown by Mr. Halliwell are compared with this graph, it will be noted that lead has had a highly deleterious effect on the toughness of the alloy. When it is recognized that, in solidifying and cooling about a core, castings will be subjected to thermal stresses, any hot cracking tendencies will surely lead to trouble. Incidentally, the type of test bar we have used (Fig. 1) will tend to set up somewhat greater stresses in the cast metal during cooling in the mold than that used by Mr. Halliwell. While this might be considered a disadvantage,
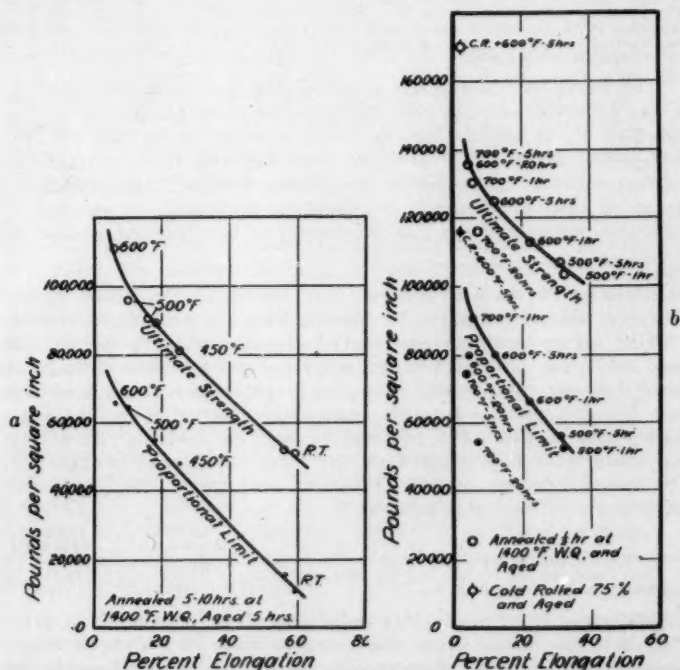

FIG. 9.—RELATION BETWEEN ULTIMATE STRENGTH, PROPORTIONAL LIMIT, AND PERCENTAGE ELONGATION OF ALLOY CONTAINING 7.5 PER CENT NICKEL AND 8 PER CENT TIN. (WISE AND EASH.)

a. Heat-treated sand castings. b. Heat-treated wrought strip.

the former type of test bar does, nevertheless, give some idea of the
behavior of the alloy in commercial castings.

Mr. Halliwell inquired as to what was meant by our statement
that as little as 0.05 per cent of lead "seriously retards" aging. By
that we mean that full hardness cannot be obtained within a reason-
able lengh of time with even this amount of lead, and under some con-
ditions involving stressing of the metal above the melting point of lead
(such as is accomplished by very rapid cooling from 1400°F.), prac-
tically no age hardening occurs. The hardness data of Table 8 were
obtained on chill cast 5 per cent nickel — 5 per cent tin bronze cold
rolled 50 per cent and then heat treated.

Table 8

PROPERTIES OF 88 PER CENT COPPER — 5 PER CENT NICKEL
5 PER CENT TIN — 2 PER CENT ZINC ALLOY

Heat Treatment	Brinell Hardness	
	0% Lead	0.05% Lead*
Annealed 1400°F. — 10 hr. — *water quenched*	72	72
Same treatment as above and then aged 600°F. — 5 hr.	170	88
Annealed 1400°F. — 10 hr. — *air cooled*	80	80
Same treatment as above and then aged 600°F. — 5 hr.	200	148

* Amount added to melt.

The *water quenched* specimen containing lead aged very slightly,
as may be noted, and hot cracking probably occurred. By air cooling
from 1400°F., appreciable age hardening occurred in the 0.05 per cent
lead sample, although the hardness level was still below that of the
lead-free material. It should be emphasized, however, that undesirable
stresses in castings often will be difficult to avoid and the presence of
impurities which increase the sensitivity to hot cracking should be
avoided.

Recently we made a 100 per cent remelt (30 lb.) heat (using
gates and risers from a 150 lb. foundry heat) in an unlined crucible
in which red brass (5 per cent lead) had been previously melted, and
found 0.009 per cent of lead on analysis in the tensile casting. A
second 100 per cent remelt heat (using gates and risers from the
same foundry heat) was made immediately thereafter in the same
crucible and showed 0.005 per cent of lead on analysis. Bars from
both heats were heat treated at the same time, using a 1400° F.
—10 hrs.—oil quench + 550° F.—5 hrs.—oil quench heat treatment
and tested with results as follows:

	1st heat 0.009% Pb	2nd heat 0.005% Pb
Tensile Strength, lb. per sq. in.	75,000	87,750
Yield Point, lb. per sq. in. (0.5 per cent ext.)	54,500	65,750
Elongation (per cent in 2 in.)	17.5	14.0

While the difference in lead content is slight, the change in prop-
erties is in the direction one might expect, since lead tends to retard
age hardening. It may, however, be noted that the relationship be-
tween tensile strength and elongation of the 0.009 per cent Pb melt
is quite satisfactory, and no indications of hot cracking were noted,
suggesting that this amount of lead may be tolerable. To obtain max-

imum properties on age hardening, however, lead should be apparently well under 0.01 per cent.

It likewise may be noted that the amounts of lead investigated by Mr. Halliwell are not of the order added to casting alloys to secure hydraulic tightness or free machining.

J. J. KANTER[3]: I should like to point out that further investigation of the effect of size of section on the age hardening of the 5 per cent nickel bronzes would be of considerable interest. Mr. Kihlgren indicated that in some of his experiments the slow-cooled test bars did have high strength, which he believes is evidence that large sections would respond to the age hardening treatments as well as the test bars have been shown to respond. But such evidence cannot be accepted as conclusive proof that sections of nickel-bronze of 2 in. or larger would age harden to a satisfactory degree by the treatments disclosed in Mr. Kihlgren's paper.

The reason I bring up this point is that when 70,000 lb. per sq. in. tensile strength is sought in a copper base alloy, one is quite helpless with the ordinary as-cast materials that are available. For thin sections, there are copper base alloys which will readily give 70,000 lb. per sq. in. tensile strength as cast. Now, if it can be demonstrated that the 5 per cent nickel bronze can be satisfactorily age hardened in large sections, there is a very definite field of application for it. For small sections, it is doubtful whether it will compete with copper base alloy which possesses 70,000 lb. per sq. in. tensile strength as cast.

MR. KIHLGREN: This question can be partially answered by drawing on the experience of one foundry making castings for an electrical application, the casting containing sections about 2 in. thick. We have been informed that no difficulty is encountered in obtaining age hardening through the heavy sections. Work is now in progress in an outside foundry and will supply a more complete answer to Mr. Kanter's question. We should not anticipate any difficulty in getting suitable age hardening in heavy sections as long as the casting is adequately fed.

Many of the copper base alloys developing 70,000 lb. per sq. in. tensile strength, such as aluminum and manganese bronze, are so called "high shrinkage" alloys and are difficult to cast, particularly in heavy sections. The aluminum bronzes also require special attention to avoid dressing effects. The 5 per cent nickel — 5 per cent tin bronze is relatively simple to cast, and is quite similar in its general characteristics to 88-10-2 (G-bronze).

It also has the considerable advantage of not introducing elements which would impair other bronzes being made in the same shop, if scrap should become mixed with them. This is a matter of considerable importance.

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[3] Research Metallurgist, Crane Co., Chicago, Ill.

The high proportional limit of the heat treated nickel bronze is also valuable — as a matter of fact, one of the factors leading to our work on this alloy was the request by a large foundry producing aluminum bronze for an alloy possessing a high proportional limit.

MEMBER: Will Mr. Kihlgren tell us what conductivity values were before and after age hardening?

MR. KIHLGREN: In the as cast condition, the resistivity is about 14 microhms per cm^3 and after quenching from 1400°F. about 15 microhms per cm^3. When age hardening is essentially complete, it will be about 10.5 to 11 microhms which correspond to approximately 15 per cent conductivity.

Effect of Aluminum on the Properties
of Medium Carbon Cast Steel

By C. E. Sims* and F. B. Dahle**, Columbus, Ohio

Introduction

1. Aluminum has been used in steel either as an alloy or deoxidizer practically ever since any of the metal has been available. As early as 1859, Rogers[1] reported the use of 0.8 per cent aluminum in steel. Aluminum at that time cost about $17 a pound. About 1885, the sodium reduction method brought the price to about $12 a pound and, in 1890, Charles Hall developed his electrolytic process and brought the price to about $2 a pound and in 1892 to $1 a pound. This naturally stimulated the use of the metal.

2. Although the steel casting industry was in its infancy in 1885, it already was recognized that aluminum is a valuable aid in preventing certain types of porosity in steel castings. This property of aluminum has been proved and well established in the intervening years until it is now an accepted fact, even though the basic mechanism of its action has not been established definitely [2, 3].

3. It is, however, just as well known that the use of aluminum to prevent porosity in steel castings has had several disadvantages. Aluminum has been castigated variously as a poisoner of steel and a destroyer of good physical properties. Some of the accusations are not without basis of fact, because numerous data are available to attest that small additions of aluminum to medium carbon cast steel may reduce the ductility to less than half of what it would be without the aluminum addition.

4. Efforts have been made to find a suitable substitute for aluminum or otherwise to eliminate its use. Such efforts have been

successful in varying degrees and many tons of steel castings have been made without the use of aluminum.

5.   Many foundrymen have been willing to suffer the adverse effects of aluminum to gain the advantages. These adverse effects have been by no means constant, but their variability and apparent unpredictability have fostered the hope that if conditions were properly controlled, good ductility might always be obtained.

6.   There has been much speculation and discussion but, despite the years that aluminum has been used, very little fundamental information has been advanced to account properly for its action. In an attempt to supply this lack, a study was undertaken at Battelle Memorial Institute and sponsored by the Aluminum Company of America which had as its objective the determination of the conditions under which the beneficial effects of aluminum might be obtained without suffering any harmful loss in ductility.

HISTORICAL REVIEW

4.   The cause of low ductility in aluminum treated cast steel in the past, often has been ascribed to the presence of aluminum oxide in the steel[2, 4]. This has been disproved substantially and it has been shown, on the other hand, that the cause is intimately tied up with the sulphides. The sulphide inclusions produce their effect partly by the actual quantity present but more particularly by the nature of their distribution.

5.   In 1924, Comstock[5] described the chain type of sulphide produced by aluminum additions to cast steel and showed their coincidence with low ductility. Sims and Lilliequist[6] gave considerable substantiating evidence that the type of sulphide inclusions in a steel directly affects the ductility and that additions of aluminum have a potent influence on the manner in which the sulphides precipitate. They gave evidence that a steel might have considerable alumina present and still have excellent ductility, providing it also had a favorable type of sulphide inclusions.

6.   Inasmuch as the present discussion concerns the effects of aluminum, this relation will be reviewed briefly. When a steel, deoxidized with manganese and silicon, freezes, it usually will be found to contain two principal kinds of inclusions, iron manganese silicates and iron manganese sulphides. Often these occur as duplex inclusions consisting of part silicate and part sulphide.

7. Such inclusions are globular in shape and have a wide range of sizes, determined by the section of the casting in which they formed. Their apparently random arrangement or distribution minimizes their ability to affect the physical properties of a steel, which will have the highest ductility commensurate with such factors as composition and heat treatment. A photomicrograph of a typical section of such a steel is shown in Fig. 5.

8. If, to such a steel, an amount of aluminum just sufficient to deoxidize it be added, there will be a profound change in the nature of the inclusions. The silicates vanish and in their place appear scattered flocs of alumina which may contain some spinel. The sulphides are very much smaller, fairly uniform in size, are mostly globular, but often appear as films or in a lamellar structure, the last named occurring where three or more primary crystals meet. What is of greatest importance, however, the sulphides are confined strictly to narrow grain boundaries of the primary crystals. In other words, they have precipitated as part of a sulphide-metal eutectic. Thus constricted and small, they are close together and form a network enveloping the primary crystals, constituting zones of weakness. In this arrangement, they have the maximum effect on the physical properties of the steel containing them and the steel has very low ductility. The structure of such a steel is shown in Fig. 7.

EXPERIMENTAL PROCEDURE

9. The laboratory heats of steel used in this study were made in a high-frequency induction furnace powered by a 150 KVA motor generator, designed to operate at a frequency of 1800 cycles. The furnace was lined with a pre-formed silica crucible having a normal capacity of 100 lb.

10. The charges used were 65 lb. in weight and were of low carbon steel plate clippings. No slag was purposely added, the slag which naturally formed from the silica of the crucible and iron oxide from the charge being considered sufficient. No slagging operations were attempted.

11. Temperature control was obtained as a function of power input and time. Temperature measurements were made with a calibrated optical pyrometer sighting either on a clean surface of the steel in the crucible or on the stream of steel as it poured from the spout. A correction factor of 0.4 for steel in the open was used.

12. Control of composition was obtained by analyzing the melting stock and introducing carefully weighed additions. Losses and recoveries are very constant in the induction furnace for a given type of operation and desired compositions can be obtained with considerable accuracy.

13. The bulk of the carbon was added as granular artificial graphite and usually was added with the cold charge. The other constituents, except aluminum, were added as standard ferro-alloys, after the charge was melted. Aluminum was added as commercially pure aluminum wire.

14. The time required for melting and finishing a heat seldom exceeded one-half hour, unless it was purposely prolonged. There is a motor action of the induced heating current which causes the molten steel to flow upward in the center of the crucible and down at the sides. The surface of the melt is slightly crowned. Because of the circulation of the metal, any addition is disseminated quickly throughout the melt and the time factor for any operation is reduced greatly as compared to some other types of melting furnaces.

15. When a heat was finished, the slag was held back and the steel poured directly into a mold. A core type mold of oil sand was used and the casting was a double keel test block having two coupons 1⅛ in. square and 14 in. long. These coupons were burned off the raw casting with an oxy-acetylene torch. A cross section of the test block is shown in Fig. 1.
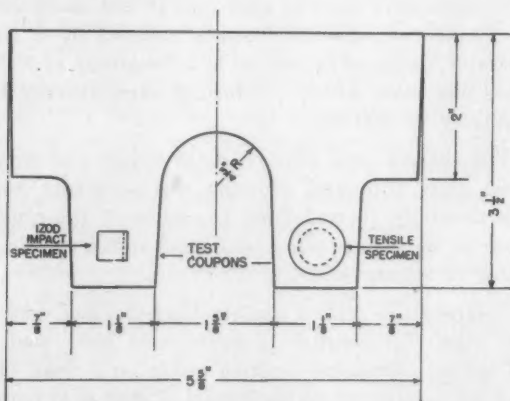

FIG. 1—CROSS-SECTION OF KEEL TEST BLOCK.

16. For simplicity and ease of comparison, all coupons were given a similar heat treatment. This was a normalizing treatment which consisted of heating to 1650°F., holding at temperature for 2 hr. and cooling in still air. The normalizing treatment was followed by a low temperature draw at 750°F. for 2 hr.

17. After heat treatment, the coupons were machined to standard 0.505 in. diameter tensile bars and standard v-notched Izod bars. Some of the tests were made in duplicate but most were in triplicate.

Experimental

18. The steel composition used throughout practically all of these tests is what is known as Grade "B" cast steel, a plain carbon steel free of alloys other than those used primarily for deoxidation. The specifications for this steel call for minimum physical properties as follows:

Ultimate strength, lb. per sq. in.........70,000
Yield strength, lb. per sq. in.............38,000
Elongation in 2 in., per cent............   24
Reduction of Area, per cent............   36

19. To eliminate variation in ductility due to composition, an effort was made to keep carbon in the range 0.26 to 0.30 per cent, manganese 0.60 to 0.80 per cent and silicon 0.30 to 0.40 per cent. The carbon content of a steel is known to be very potent in affecting both the strength and ductility of such a steel, whereas the manganese content affects the strength much more than it does the ductility. When variations in composition were used, they were referred to this base composition for comparison.

20. In the earlier tests, the aluminum addition was kept constant at 0.05 per cent or in the proportion of 1 lb. per ton. This quantity was selected as representing an average commercial addition and enough to produce typical effects. It is also slightly in excess of the quantity needed to combine with all the oxygen that could theoretically be present in a 0.30 per cent carbon steel at 2900°F. (1600°C.).

21. Every suggested variation in practice was tried on various heats. High finishing temperatures were compared with low finishing temperatures, highly oxidized steels were compared to others in which a high residual silicon content was maintained, and

various orders and times of adding deoxidizers were tried. In all of these variations, duplicate heats were run to which no aluminum was added, to evaluate the effects due to aluminum alone.

22. A partial list of these heats and the physical properties obtained are shown in Table 1. A perusal of the data in this table will disclose no steels with very low figures for ductility while some of those treated with aluminum show unexpectedly good ductility.

23. There are differences that may be due to the various

Table 1

EFFECT OF MELTING PRACTICE ON PHYSICAL PROPERTIES OF LABORATORY HEATS

Heat No.	Melting Practice No.*	C %	Mn %	Si %	P %	S %	Al Added %	Yield Strength, lb. per sq. in.	Ultimate Strength, lb. per sq. in.	Elong., %	R.A., %	Izod Impact Strength, ft. lb.
2327	1	0.28	0.86	0.37	0.021	0.035	......	52,500	84,500	28.5	52.0	33.5
2386	1	0.28	0.79	0.41	0.026	0.036	0.05	52,000	80,500	26.5	43.0	40.0
2393	1	0.26	0.78	0.34	0.023	0.035	0.05	48,500	75,500	28.0	41.5	39.0
2448	2	0.25	0.84	0.39	0.024	0.034	......	45,500	79,000	31.5	57.5	40.0
2449	2	0.27	0.84	0.36	0.025	0.033	0.05	47,000	77,500	32.5	53.0	47.0
2328	3	0.28	0.85	0.36	0.018	0.035	......	50,000	84,000	29.5	54.0	31.0
2387	3	0.25	0.70	0.36	0.023	0.036	0.05	51,500	79,500	27.5	44.5	39.0
2450	4	0.26	0.78	0.36	0.023	0.034	......	46,000	77,000	32.5	57.5	37.0
2451	4	0.29	0.77	0.35	0.022	0.037	0.05	47,500	76,000	29.0	47.0	39.0
2388	5	0.28	0.80	0.38	0.025	0.036	......	50,000	83,000	29.5	51.0	31.5
2389	5	0.27	0.79	0.36	0.022	0.036	0.05	52,500	80,000	28.5	46.5	40.5
2452	5	0.26	0.80	0.31	0.023	0.037	......	47,500	80,000	32.5	56.5	37.0
2453	5	0.28	0.81	0.34	0.025	0.037	0.05	51,000	78,000	29.0	53.5	45.5
2392	6	0.25	0.81	0.35	0.016	0.035	0.05	48,000	76,500	28.5	46.5	42.5
2454	6	0.30	0.84	0.39	0.021	0.034	0.05	51,000	78,500	30.0	50.5	46.0
2394	7	0.26	0.88	0.37	0.016	0.034	0.05	52,500	78,500	28.0	46.5	43.0
2395	7	0.26	0.86	0.37	0.018	0.034	......	48,500	82,000	29.5	49.0	42.0
2456	7	0.26	0.87	0.37	0.023	0.034	......	50,500	82,500	30.0	58.0	36.5
2457	7	0.28	0.58	0.38	0.023	0.034	0.05	55,000	83,000	28.0	50.5	34.5
2458	8	0.25	0.82	0.34	0.026	0.034	0.05	49,500	78,000	30.0	49.0	41.5
2459	8	0.23	0.84	0.35	0.023	0.034	0.05	49,500	77,000	28.0	47.0	33.0
2619	9	0.30	0.82	0.36	0.015	0.034	0.05	51,000	81,000	30.5	49.0	35.5
2670	9	0.35	0.79	0.43	0.015	0.035	......	47,500	84,500	23.5	36.5	26.5

*Melting Practice No.
1. Heat deoxidized with silicon and manganese at 2800°F., then superheated to 2950°F. before pouring.
2. Heat deoxidized with silicon and manganese at 2800°F., then superheated to 2850°F. before pouring.
3. Heat deoxidized with silicon and manganese at 3000°F., then cooled to 2950°F. before pouring.
4. Heat deoxidized with silicon and manganese at 2900°F., then cooled to 2850°F. before pouring.
5. Small amount of mill scale added to heat just before deoxidation was started.
6. Aluminum added to bath before deoxidation with silicon and manganese.
7. Final silicon content represents residual silicon carried through heat from cold charge.
8. Aluminum added to stream while filling mold.
9. Heat melted under reducing atmosphere.

melting practices but they do not seem to be consistent and no definite trend is shown. What seem to be diametrically opposite practices sometimes give similar results. For example practice no. 5, in which a charge of mill scale was added just before deoxidation and practice no. 7, in which the final silicon represents the residual silicon carried through the heat, when averaged, give very nearly the same results.

24. The preceding statement should not be interpreted as evidence that carrying a high residual silicon content is good practice. There is evidence to show that a high residual silicon content favors harmful absorption of hydrogen, which did not take place, however, in the short time the induction furnace heat was kept in a molten state.

25. In this connection, it is interesting to note that heat 2670, practice no. 9, which was made in an atmosphere of natural gas but had no aluminum addition, had relatively poor properties. The low ductility and impact resistance were due directly to porosity in the test bar. Regardless of the care in preparing the dry sand mold, porosity resulted from this practice in check tests, except when aluminum was used. Heat 2619 which was made the same way, except for the addition of 0.05 per cent aluminum, is sound and has a satisfactory ductility. In this case, aluminum improved the ductility by preventing porosity.

26. Of the steels shown in Table 1, all that were deoxidized with aluminum had the same addition, namely 0.05 per cent. It will be shown later that this particular amount of aluminum probably had as much to do with the erratic and high average results as the melting practices used. The general effect of this addition of aluminum was to slightly increase the yield strength, slightly lower the tensile strength, lower the ductility and increase the impact resistance. The averages show:

	Elongation %	Red. Area %	Izod ft.-lb.
Steels not deoxidized with aluminum	29.7	52.4	35.0
Steels deoxidized with 0.05 aluminum	29.0	45.2	40.6

27. Although the ductilities and impact values did not vary in any definite manner with the melting practice, there was a consistent relation between the microstructure and these properties. Those aluminum deoxidized heats which had the lowest ductility, also had sulphide inclusions which most nearly approached the

FIG. 2—TOP—HEAT 2393. BOTTOM—HEAT 2453. SHOWING DIFFERENT TYPES OF SULPHIDE
INCLUSIONS OBTAINED WITH AN ADDITION OF 0.05 PER CENT ALUMINUM. HEAT 2393
HAS FILM AND EUTECTIC TYPE, HEAT 2453, MORE MASSIVE AND WIDER SPACED. UNETCHED
250X.

chain-like, eutectic pattern while those with good ductility had more massive sulphide inclusions. Examples of these are given in Fig. 2, which shows micrographs of the inclusions of heats 2393 and 2453 of Table 1.

SULPHUR AND MANGANESE

28. Manganese, because of its great affinity for sulphur, is very potent in controlling the character of sulphides in steel and their effect on the properties of the steel. It was thought possible, therefore, that the manganese-sulphur ratio might be of great importance in controlling the effect of aluminum on the properties of steel.

29. To determine this factor, a series of heats was made in which sulphur contents were varied in steps of 0.020, 0.050 and 0.070 per cent while the manganese was varied in steps of 0.5, 1.0 and 1.5 per cent. These steels were made with the same melting practice and all were deoxidized with 0.05 per cent aluminum added after the manganese and silicon additions and about ½ min. before pouring. The compositions and physical properties are shown in Table 2.

30. The table shows that both sulphur and manganese have pronounced effects which are, however, quite independent of each other. Increasing the sulphur content produces a progressive lowering of the ductility and impact resistance without any appreciable change in the strength. Increase of the manganese content, on the other hand, causes a marked increase in y i e l d and ultimate strengths without an important effect on the ductility or impact resistance.

Table 2

EFFECT OF MANGANESE AND SULPHUR ON PHYSICAL PROPERTIES*

Heat No.	C %	Mn %	Si %	P %	S %	Yield Strength, lb. per sq. in.	Ultimate Strength, lb. per sq. in.	Elong., %	R.A., %	Izod Impact Strength, ft.-lb.
2414	0.27	0.49	0.37	0.018	0.021	44,000	75,000	32.5	55.0	64.0
2432	0.24	0.99	0.39	0.019	0.021	52,000	80,000	32.0	56.5	65.0
2433	0.26	1.51	0.41	0.017	0.020	60,500	90,000	30.0	62.5	77.0
2408	0.28	0.60	0.35	0.018	0.050	44,000	76,500	29.0	43.0	33.5
2410	0.25	1.07	0.37	0.022	0.051	57,500	83,000	28.0	47.0	35.0
2434	0.27	1.60	0.38	0.019	0.057	64,000	90,000	24.0	36.5	28.0
2409	0.29	0.63	0.38	0.019	0.071	48,000	77,500	20.5	22.5	25.0
2411	0.26	1.05	0.38	0.021	0.073	56,000	81,500	25.5	34.0	26.0
2413	0.26	1.60	0.41	0.020	0.073	65,500	90,000	22.5	27.5	22.5

*Results shown are average of duplicate tests. All coupons normalized 2 hr. at 1650°F. followed by a 2 hr. draw at 750°F.

VARIATION OF SULPHUR AND ALUMINUM

31. In the tests so far described, only sulphur content has been shown to have anything like a quantitative effect on ductility. The effect of aluminum variation was next determined. Test heats were made using four sulphur levels namely, 0.02, 0.03, 0.04, and 0.06 per cent and in which aluminum additions were varied from none to 0.25 per cent. An attempt was made to keep melting and finishing practices and temperatures constant.

32. There were some m i n o r differences in the materials charged. For the 0.02 per cent sulphur steels, ingot iron of low sulphur content was used as melting stock; for the 0.3 per cent sulphur a mixture of low carbon sheet clippings and ingot iron was used, and for the others the low carbon sheet only. Iron sulphide was used when it was necessary to raise the sulphur.

33. The log of a typical heat was as follows: The charge of 65 lb. of steel scrap was put into the crucible of the acid-lined, high-frequency induction furnace. Carbon in the form of granular graphite and iron sulphide, when needed, were added with the cold charge and placed near the bottom. After the power was turned on, the charge was melted in about 20 min. Three min. superheating brought the temperature to 2850°F. At this time, ferromanganese and ferrosilicon were added. One more minute of heating raised the temperature to 2900°F. At this time, the aluminum was added as ⅛-in. wire. The power input was then lowered to hold the temperature constant. With this reduced power, there is no turbulence in the bath, just a gentle roll. One minute was allowed in this condition for thorough mixing and the steel was then poured directly into the test block molds.

34. The coupons were given a single normalizing treatment at 1650°F. for 2 hr. and then drawn back at 750°F. All physical tests were made in triplicate. Such excellent agreement was found in the triplicate tests of any given heat that only the averages are recorded here. Some of these data together with the compositions of the steels are shown in Table 3.

35. In examining this table, it will be observed that there is great uniformity in the strength of the various steels. What small differences there are, can be accounted for in practically all cases by the carbon and manganese content. One exception to this is

Table 3

EFFECT OF ALUMINUM AND SULPHUR ON PHYSICAL PROPERTIES*

	Chemical Composition						Physical Properties				
Heat No.	C %	Mn %	Si %	P %	S %	Al Added %	Yield Strength, lb. per sq. in.	Ultimate Strength, lb. per sq. in.	Elong., %	R A %	Izod Impact Strength, ft.-lbs.
2792	0.28	0.68	0.32	0.016	0.019	....	45,500	75,500	34.0	61.5	51.5
2793	0.28	0.70	0.35	0.015	0.020	0.025	46,500	75,000	32.5	55.5	49.5
2794	0.29	0.69	0.36	0.018	0.019	0.050	46,000	75,000	33.0	58.5	56.5
2795	0.29	0.69	0.38	0.018	0.018	0.10	47,500	76,000	34.0	58.0	64.0
2796	0.31	0.72	0.39	0.018	0.020	0.15	47,000	77,500	31.0	57.5	68.0
3241	0.31	0.79	0.41	0.029	0.019	0.25	55,500	83,500	30.5	54.5	55.0
2797	0.29	0.66	0.27	0.011	0.031	....	44,500	73,500	34.0	56.5	51.0
2798	0.28	0.66	0.27	0.012	0.029	0.025	46,000	73,500	29.5	46.0	40.0
2791	0.29	0.79	0.40	0.012	0.035	0.05	49,500	77,000	31.0	49.0	41.5
2799	0.28	0.65	0.30	0.011	0.031	0.075	45,000	73,000	32.5	53.0	42.5
2800	0.26	0.64	0.28	0.010	0.030	0.10	45,000	71,000	33.0	53.5	52.0
3202	0.28	0.76	0.42	0.017	0.032	0.25	48,000	79,500	30.5	49.0	43.0
2802	0.28	0.62	0.31	0.011	0.041	....	46,500	74,000	32.0	50.5	35.0
2807	0.28	0.67	0.33	0.010	0.044	....	48,500	75,000	33.5	55.0	40.0
2803	0.28	0.65	0.31	0.011	0.037	0.015	46,000	74,500	26.5	35.5	27.5
2808	0.28	0.64	0.32	0.012	0.046	0.015	46,000	74,500	31.5	49.5	37.5
2804	0.29	0.64	0.32	0.012	0.038	0.025	47,000	74,000	26.0	37.0	34.5
2809	0.29	0.68	0.32	0.011	0.045	0.025	47,000	74,000	21.5	25.0	22.5
2805	0.28	0.67	0.32	0.011	0.042	0.05	47,000	74,000	30.0	50.5	36.5
2810	0.28	0.62	0.31	0.011	0.041	0.05	46,000	74,500	30.0	48.0	37.0
2806	0.29	0.63	0.33	0.012	0.040	0.075	48,500	75,000	30.5	47.5	38.5
2811	0.27	0.69	0.35	0.012	0.042	0.075	48,000	74,500	30.0	44.5	34.0
3203	0.28	0.75	0.42	0.020	0.037	0.15	47,000	79,500	31.5	47.5	34.0
3204	0.28	0.80	0.43	0.016	0.039	0.25	51,000	79,500	30.5	48.5	44.0
3205	0.28	0.77	0.38	0.019	0.061	....	46,000	78,500	29.5	46.0	27.0
3206	0.29	0.80	0.42	0.019	0.061	0.015	47,500	77,500	18.5	20.0	18.0
3207	0.29	0.78	0.41	0.019	0.060	0.025	49,000	79,000	26.0	35.5	22.0
3208	0.28	0.72	0.39	0.016	0.062	0.05	46,000	76,500	27.0	38.5	26.5
3209	0.29	0.76	0.42	0.021	0.061	0.10	49,500	78,000	27.0	39.5	29.0
3210	0.28	0.80	0.38	0.016	0.062	0.25	48,500	77,000	28.5	42.0	31.5

*All results average of triplicate tests. Heat treatment consisted of normalizing 2 hr. at 1650°F. followed by a draw at 750°F. for 2 hr.

heat 3241 which has a strength somewhat out of proportion to the composition.

36. Both sulphur and aluminum are seen to have positive and characteristic effects on the ductility and impact resistance. From a qualitative standpoint, the effects of these two elements are independent but quantitatively they partially are dependent on each other. This relation is best shown graphically as in Fig. 3 where the data of Table 3 are plotted.

37. Here is clearly shown an important and hitherto unreported relation between ductility and the amount of aluminum addition. With other conditions, such as compositions, constant, the addition of small increments of aluminum to a steel of this type at first produces a progressive drop in the ductility and impact resistance. Very soon, however, a minimum point is reached where further increments in the aluminum addition cause an upturn in the curves and a recovery of ductility and impact resistance.

38.   The recovery of ductility continues to a point only slightly below that of the aluminum-free steel and then further additions of aluminum up to 0.25 per cent or 5 lb. per ton have very little additional effect. It is significant, however, that such large additions of aluminum are definitely not detrimental to the ductility. If anything, they continue to produce a slight improvement.
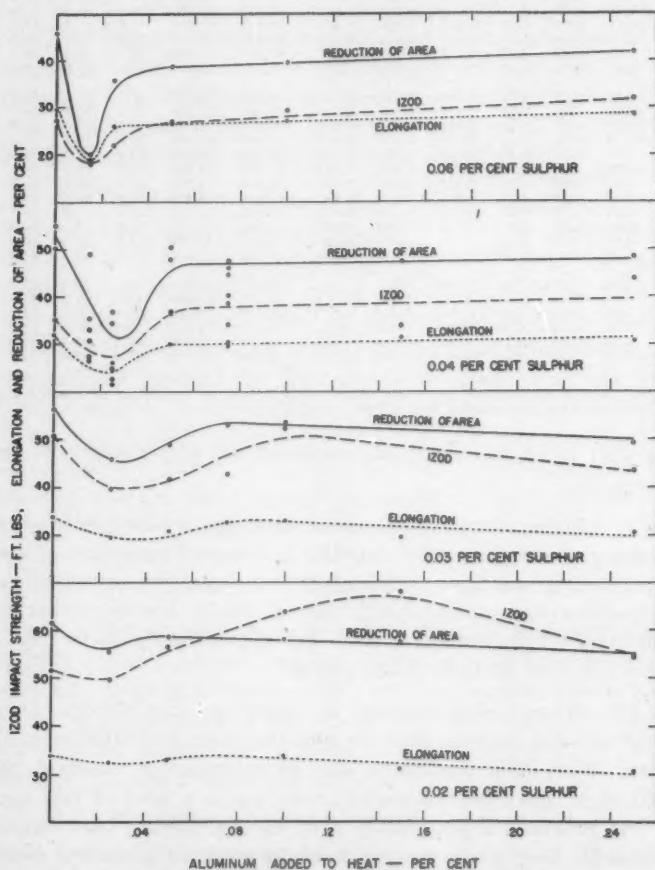

FIG. 8—QUANTITATIVE EFFECT OF ALUMINUM ON DUCTILITY AND IMPACT STRENGTH OF
GRADE "B" CAST STEELS SHOWN IN TABLE 8.

39. The recovery of the impact resistance continues to a value slightly higher than that of the aluminum-free steel. Otherwise, ductility and impact resistance parallel each other very closely.

40. The characteristic shape of the curves is the same for all sulphur contents but quantitatively they differ. In the high sulphur steels, the loss in ductility is very rapid and of great magnitude and the recovery is just as rapid. In the low sulphur steels, the loss in ductility is very slight at the lowest point in the curve. It is noteworthy that in steels containing as little as 0.02 per cent sulphur the ductility practically is unaffected by aluminum additions.

41. The minimum ductilities for any given sulphur content seem to be obtained with an aluminum addition of from 0.015 to 0.025 per cent, which corresponds to an addition of 5 to 8 oz. per ton. Because the curves are so steep in this region, however, it is doubtful whether the minimum possible ductility was reached in any of these steels. Where a very small difference in the aluminum causes such a relatively great difference in ductility, the results obtained are apt to be erratic. This is illustrated by the results obtained with additions of 0.015 and 0.025 per cent aluminum in the 0.04 per cent sulphur steels.

42. Recovery is almost complete with 0.05 per cent aluminum or 1 lb. per ton, although slight improvements are obtained with larger additions. It will be recalled that in most of the early tests, an addition of 0.05 per cent aluminum was used and it is now apparent why no very low results were obtained. It is a l s o probable that they were erratic because this addition is in a region (see curve for 0.04 per cent sulphur) where a slight variation in conditions might cause a large variation in ductility. Later data indicate that the size of the aluminum addition necessary to give recovery of ductility will vary somewhat with conditions of manufacture.

SULPHUR AND DUCTILITY

43. Even with no aluminum addition, the ductility and impact values of the steels vary inversely with the sulphur content. For any given aluminum addition, the ductility is lower the higher the sulphur content. The effect of increasing sulphur, of course, is greatest when the aluminum content is that which gives
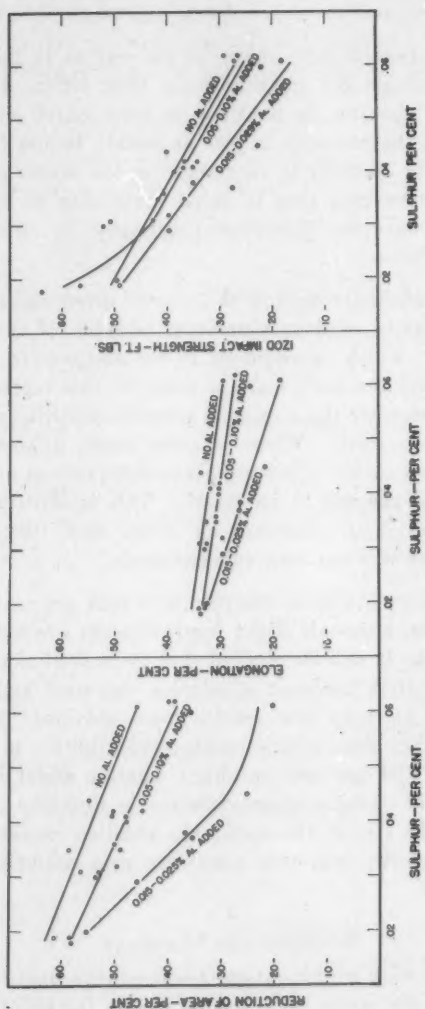
Fig. 4.—Effect of Sulphur, With and Without Aluminum, Additions on Ductility and Impact Resistance of Steels Shown in Table 3.
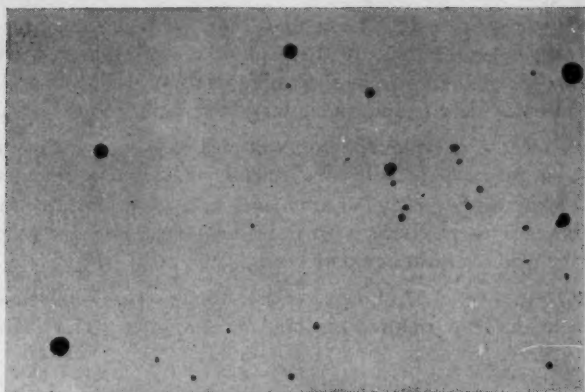
FIG. 5—HEAT 2807. NO ADDITION OF ALUMINUM. INCLUSIONS CONSIST OF GLOBULAR
SILICATES AND SULPHIDES. HIGH DUCTILITY. UNETCHED 250X.

lowest ductility. After recovery of ductility, the effect of quantity
of sulphur is practically the same as when no aluminum is used.
These relations are shown graphically in Fig. 4.

CORRELATION OF DUCTILITY AND MICROSTRUCTURE

44. In attempting to explain this action of aluminum, a
consistent and significant correlation was found to exist between
the ductility* and the character of the inclusions. This is illus-

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* Note: Inasmuch as the impact resistance was shown to vary with the ductility,
statements about the latter are qualitatively true about the impact resistance.
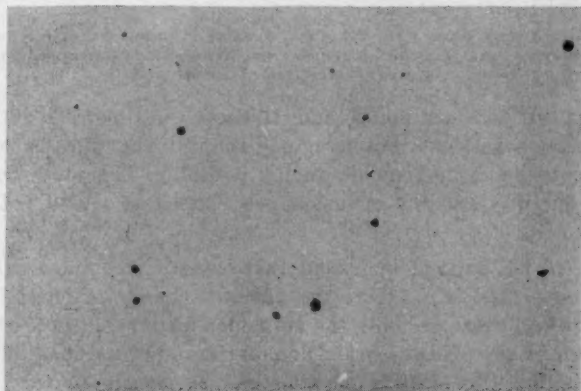
FIG. 6—HEAT 2808 WITH 0.015 PER CENT ALUMINUM ADDED. INCLUSIONS CONSIST OF
GLOBULAR SULPHIDES AND SLIGHTLY MODIFIED SILICATES. HIGH DUCTILITY. UNETCHED 250X.

trated by Figs. 5 to 10. The compositions and physical properties
of the steels represented by these photomicrographs are given in
Table 3. All of them are in the range of 0.04% sulphur and were
selected as typical examples showing the influence of aluminum.

· 45.   Fig. 5 is a photomicrograph of a steel in which no alumi-
num was used. The inclusions consist of globular silicates and
sulphides of a wide range of sizes and with an apparently random
dispersion. This steel has high ductility. For reference purposes,
these inclusions are designated no. I type.

46.   In Fig. 6 is shown a type of structure sometimes ob-
tained with very small additions of aluminum. The inclusions
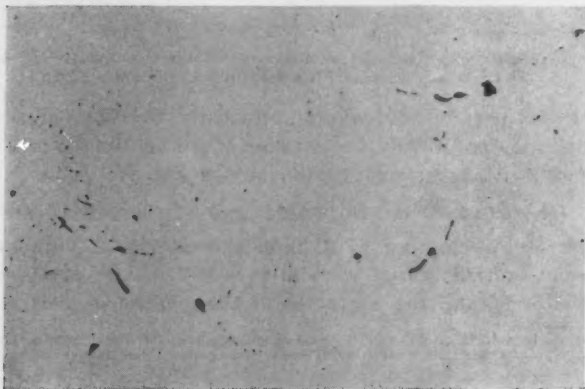
FIG. 7—HEAT 2809 WITH 0.025 PER CENT ALUMINUM ADDED. INCLUSIONS SHOW NO
SILICATES BUT ONLY EUTECTIC SULPHIDES AND OCCASIONAL CLUSTERS OF ALUMINA.
DUCTILITY VERY LOW. UNETCHED 250X.

consist of both sulphides and silicates but, instead of being
globular, have somewhat irregular shapes. The silicates have a
strong tendency to crumble during polishing. This steel also has
high ductility and the inclusions are classed as a modified no. I
type.

47.   The type of inclusions that cause extreme low ductility
are shown in Fig. 7. Here the silicates have completely disap-
peared and there are only the sulphides distributed in a eutectic
pattern in the primary grain boundaries, and occasional clusters
of alumina. No clusters of alumina are shown on this figure
because they are usually isolated but a typical group is shown in
Fig. 11. These inclusions are designated the no. II type.

FIG. 8—HEAT 2805 WITH 0.05 PER CENT ALUMINUM. INCLUSIONS CONSIST OF DUPLEX SULPHIDES, LARGE AND IRREGULAR DARK CONSTITUENT Al₂S₃. SOME ALUMINA PRESENT. DUCTILITY GOOD. UNETCHED 250X.

48. When more aluminum is added, as in heat 2805, Fig. 8, the inclusions undergo another marked change coincident to the recovery of ductility. There are still only sulphides and alumina but the sulphides are larger and consequently farther apart, but still in the grain boundaries. The shapes are very irregular and there is obviously more than one constituent present. Clusters of alumina are seen occasionally in such a steel but they are so rare that it seems some of the alumina must be in-
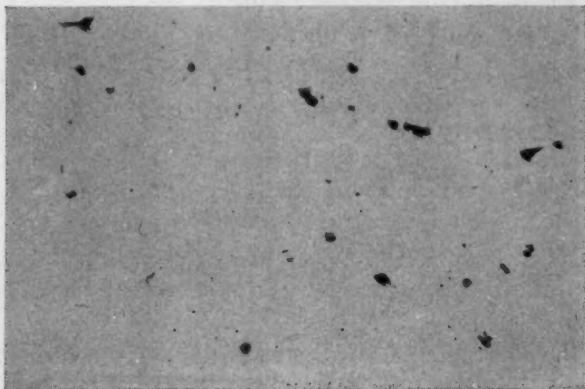
FIG. 9—HEAT 2811 WITH 0.075 PER CENT ALUMINUM. INCLUSIONS SIMILAR TO THOSE OF FIG. 8. UNETCHED 250X.

corporated in the sulphides. These inclusions are designated the
no. III type.

49.   The recovery of ductility is due to the fact that the
sulphides, being larger, are now farther apart and do not form
continuous zones of weakness where the steel can start tearing.

50.   When still more aluminum is used, as in heats 2811 and
3204, Figs. 9 and 10, there is a further progressive change in the
sulphides. The light-colored constituent decreases and the dark-
colored constituent predominates. At the same time, they get
somewhat smaller and rounder. There is no significant difference
in ductility with this change and the inclusions are regarded as
modified no. III type.

EXPLANATION OF ALUMINUM EFFECT

51.   The basic cause for this variable behavior of the sul-
phide inclusions is quite obviously a variation in their solubility.
All sulphides apparently have a considerable solubility in molten
steel and are completely soluble in quantities usually present in
commercial steels. According to Chipman[7], experimental evidence
favors the belief that sulphur in liquid iron exists as a sulphide
whose molecule contains one atom of sulphur and whose formula
may be written FeS.

52.   In the process of freezing, the sulphides in a steel are
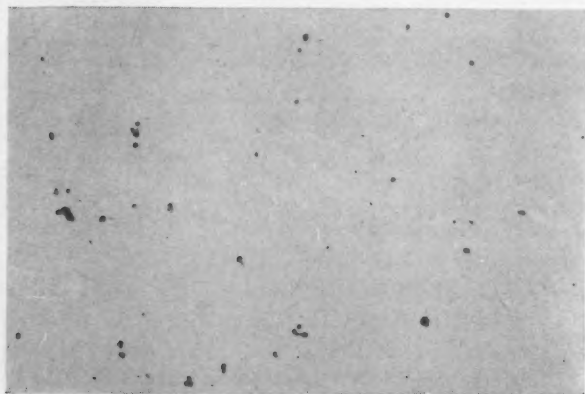concentrated in the mother liquid until the saturation point is

FIG. 10—HEAT 3204 WITH 0.25 PER CENT ALUMINUM. SULPHIDE INCLUSIONS DARKER
AND ROUNDER. BECAUSE OF GREATER EXCESS OF ALUMINUM, THEY CONTAIN MORE Al_2S_3.
DUCTILITY GOOD. UNETCHED 250X.

reached, at which time they start precipitating to form sulphide inclusions. With varying degrees of solubility, the precipitation will start at different stages of the freezing.

53. Iron sulphide is nearly completely miscible with liquid iron. It is true that at high concentrations FeS and Fe crystallize as a eutectic with 31 per cent sulphur but at low concentrations the eutectic is not in evidence and the sulphide forms as a film in the grain boundaries. The mechanism of this formation is that, during freezing, iron crystallizes out of solution leaving the liquid ever richer in sulphide until only sulphide is left in the crystal interstices.

54. In referring to grain boundaries and crystal interstices, it should be made clear that these are not the boundaries of grains found after the steel has undergone one or more allotropic modifications, which would occur in simple cooling to room temperature. These are the primary crystals, formed during freezing or the interstices of dendritic branches of the same crystal as shown by Wohrman[8].

55. Manganese has a high heat of formation with sulphur and when manganese is present in steel, it takes the lion's share of the sulphur. When there is an appreciable excess of manganese over the sulphur equivalent, the sulphides that form will be primarily manganese sulphide, although it generally is believed that some FeS will always be present. In other words the reaction.

$$FeS + Mn \longrightarrow Fe + MnS$$

goes almost but not entirely to completion.

56. The solubility of manganese sulphide in liquid steel is a variable which is dependent greatly on the composition. For example, there is reason to believe MnS has a limited solubility in cast iron which decreases rapidly with fall in temperature. In low carbon steel, the solubility is influenced by the oxygen content, and is greater the lower the oxygen content.

57. When a steel contains active oxygen (FeO), oxides and sulphides often precipitate together during cooling and freezing. The oxides and sulphides are mutually soluble in each other in the liquid state but separate at the time of solidification. Thus, duplex inclusions of sulphide and oxide are often found. Wentrup[9] has shown the type of inclusions obtained in steels containing sulphur, manganese and 0.05 per cent oxygen. In this

same article, and another by Löfquist[10], has been summarized much of the available information regarding the formation of inclusions.

58. All the steels of the present paper contained silicon, usually in the range from 0.3 to 0.4 per cent. According to Chipman[7], a steel containing 0.35 per cent silicon also will contain, at 1600°C. (2900°F.), 0.02 per cent FeO in equilibrium with the silicon. During freezing, this FeO reacts with manganese and silicon in the steel to form an iron manganese silicate with a composition that varies with the proportions of the constituents.

59. Such silicates are shown in Fig. 5. It is not intended to imply that all silicates found have this source, because an analysis of extracted inclusions usually will show a quantity of oxides (including silicates) in excess of what could be accounted for by the equilibrium quantity of FeO.

60. There is sufficient oxygen, however, to so lower the solubility of the manganese-iron sulphide that it precipates fairly
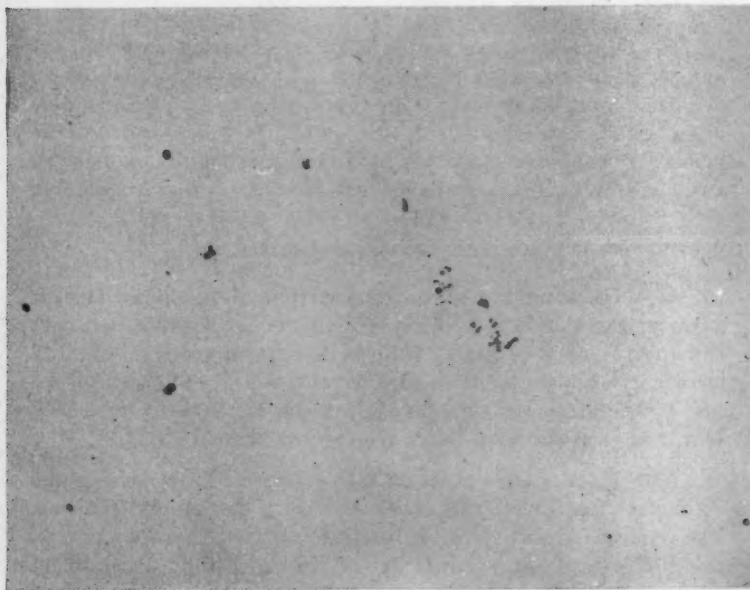

FIG. 11—ALUMINA INCLUSIONS IN ALUMINUM TREATED STEEL. THESE CLUSTERS OF Al_2O_3 ARE USUALLY ISOLATED FROM THE SULPHIDES AND MUST BE SHOWN SEPARATELY. HEAT No. 2796. UNETCHED 250X.
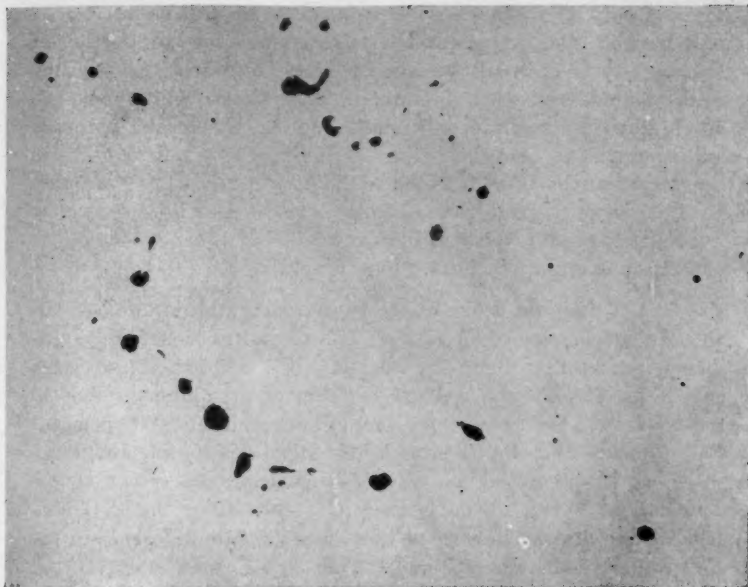
FIG. 12—HEAT 3205 ALUMINUM FREE STEEL WITH 0.06 PER CENT SULPHUR. NOTE HOW
THE SULPHIDES OUTLINE A SIX-SIDED AREA. OTHER ROWS GO OFF ON 120° ANGLES FROM
THE CORNERS. UNETCHED 250X.

early in the process of freezing and appears as relatively large
globular inclusions such as are shown in Fig. 5. These sulphides
are so few in number and so far apart that they appear to have
a purely random distribution. The fact that they are formed after
freezing has progressed far enough to saturate the liquid portion,
necessitates that they be located in crystal interstices. If, how-
ever, they start to form early enough, the interstices are wide and
they will have considerable freedom of location. When the sul-
phur content is high enough, as in heat 3205, Fig. 12, the position
of the sulphides in the crystal boundaries sometimes is shown
clearly.

61. In the absence of oxygen, the solubility of MnS is in-
creased to the extent that it precipitates as a eutectic with a
small portion of the iron. This formation is shown very clearly
when a large quantity of sulphur is present, as in heat 3649,
Fig. 13. The iron portion of the eutectic merges with the ground
mass and only the sulphide portion is distinguishable. When

there is only a small content of sulphur, there will be, of course, only a small quantity of eutectic which, being the last portion of the steel to freeze, will necessarily be deposited in the primary grain boundaries. Such a eutectic most often will appear as single rows of small evenly sized and evenly spaced sulphide inclusions, as in heat 2809, Fig. 7. Sometimes it takes the form of discontinuous films, as in heat 2393, Fig. 2. At the junction of three or more grains, the eutectic often will be represented by a group of sulphide dots, as in heat 2667, Fig. 17 or by a lamellar eutectic pattern, as in heat F1608, Fig. 15.

62.   Aluminum is one of the most powerful deoxidizers available for steel. According to Chipman's[11] calculations, the equilibrium constant at 1600°C. (2900°F.) is 7×10^{-13}. This means that if there is as little as 0.01 per cent of aluminum present in the unoxidized condition, there cannot be more than 0.002 per cent FeO present. But unless there is this small residue of aluminum, there can be absolutely no way of calculating the extent of the deoxidation. When silicon is used as a deoxidizer, the FeO content varies with the silicon but there is very little difference in the FeO content as between 0.3 and 0.4 per cent silicon.
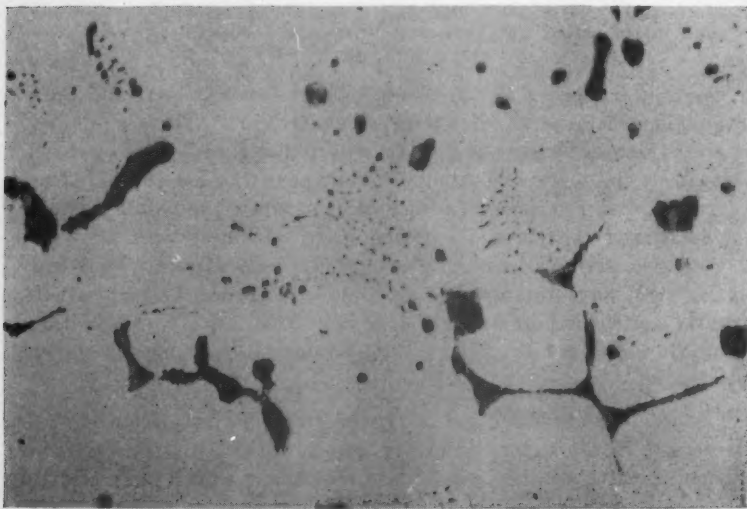

FIG. 13—HEAT 3649 CONTAINING 0.91 PER CENT SULPHUR, 0.75 PER CENT ALUMINUM ADDED. SMALL DOTS ARE MnS EUTECTIC, LARGE SULPHIDES ARE DUPLEX FeS AND Al_2S_3 WITH SOME MnS. UNETCHED 250X.
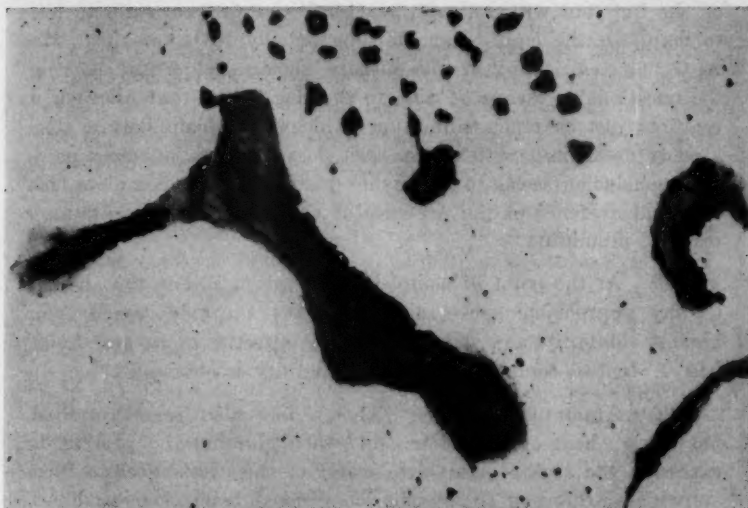
FIG. 14—HEAT 3649. SAME AS FIG. 13. BLACK AREAS IN LARGE SULPHIDES ARE Al₂S₃. LIGHT AREAS FeS. UNETCHED 1000X.

63. Being such a powerful deoxidizer, aluminum will be completely oxidized, unless added in excess of the total oxygen of the steel and, if added in such excess, will reduce all other oxides likely to be present. The first action of aluminum, when added in small quantities, less than enough for complete deoxidization, is to modify the character of the silicate inclusions. They appear to contain less FeO and MnO, probably with some substitution of Al_2O_3. These silicates tend to crumble in polishing, as shown in heat 2808, Fig. 6. In this condition, the solubility of the sulphides is not changed and they precipitate as globules with minimum effect on the ductility.

64. The reason the solubility of the sulphide is not changed is because the concentration of FeO is not altered by such small additions of aluminum. Before the silicon addition to a steel, the residual silicon may be about 0.05 per cent and the FeO in equilibrium with this will be approximately 0.055 per cent. Upon the addition of, say 0.35 per cent silicon, the reaction

$$2Si + 2FeO \rightleftarrows SiO_2 + 2Fe$$
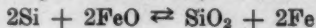
proceeds from left to right until an equilibrium at 0.02 per cent FeO is reached. Subsequent small additions of aluminum will act

on the FeO but will cause the reaction to move from right to left to maintain the equilibrium as long as any SiO_2 remains. The Al_2O_3 formed is out of the picture chemically. Thus, barring the questionable escape of SiO_2 to the slag in the time allowed, it requires just as much aluminum to deoxidize completely a steel already deoxidized with silicon as it does when no silicon is used. This conclusion seems to be further justified by the complete lack of visual evidence of the presence of silicates in steels containing residual aluminum.

65. At the point of complete deoxidation and in the absence of any appreciable excess aluminum, the sulphides reach their highest solubility and precipitate as a eutectic, as in heat 2809, Fig. 7. In this condition the lowest ductility is obtained.

66. Aluminum sulphide (Al_2S_3), like manganese sulphide, has a high heat of formation and, when aluminum is present in excess of the oxygen equivalent, some of the excess goes to form sulphide. According to information gleaned from chemical hand-
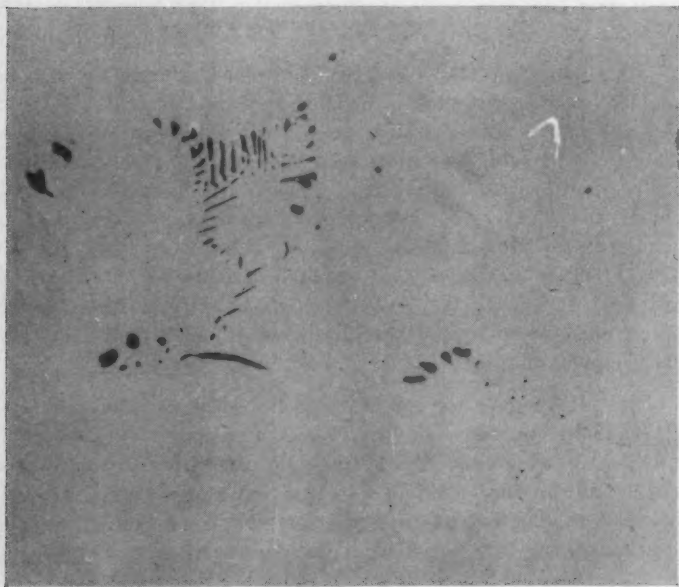

FIG. 15—HEAT F-1608 SHOWING LAMELLAR TYPE MnS EUTECTIC OCCURRING AT JUNCTION OF THREE OR MORE PRIMARY CRYSTALS. UNETCHED 500X.

books, Al_2S_3, if pure, as when sublimed, may occur as colorless or yellow asbestos-like crystals. But when slightly impure, it usually is obtained as a black sintered or vitreous mass which, when polished, has a black metallic luster. Al_2S_3 reacts slowly with water or even moist air to form aluminum hydroxide and hydrogen sulphide.

67. Aluminum sulphide forms a double sulphide with FeS. Aluminum will react with fluid FeS to produce Al_2S_3 and Fe, but it requires a 50 per cent excess of aluminum over the equivalent proportions to carry the reaction almost to completion. In a steel, the large excess of iron and the manganese will allow only a portion of the aluminum to form Al_2S_3.

68. This was demonstrated in special high sulphur heat (no. 3649) containing 0.91 per cent sulphur and 0.67 per cent manganese which had an addition of 0.75 per cent aluminum. A photomicrograph of some of the sulphide inclusions is shown in Fig. 14. Aluminum was added in excess of the oxygen and sulphur equivalent combined but there was only enough manganese to combine with less than half the sulphur.

69. The cluster of small sulphides in this steel is typical of MnS in an oxygen free steel and these inclusions are undoubtedly MnS. The large sulphides are principally FeS and Al_2S_3. There are gradations of color from a very light gray yellow to black in the constituents of these sulphides. The lightest portions are recognized as being mainly FeS and the black portions as Al_2S_3. Other portions are probably mixed sulphides. Obviously FeS, MnS and Al_2S_3 are soluble in each other in the liquid state but tend to separate during solidification.

70. The black aluminum sulphide is soft and crumbly and it is almost impossible to keep all of it in place during polishing. It is not attacked appreciably by the water in wet polishing but is readily attacked by boiling distilled water.

71. Aluminum sulphide is less metallic [17] than either MnS or FeS and should, therefore, be less soluble in steel. The effect of the Al_2S_3 is to make the sulphides less soluble, and being less soluble, they precipitate earlier in the freezing. Thus, in spite of the absence of oxygen, Al_2S_3 tends to prevent the formation of a eutectic and causes the sulphides to form large, irregular-shaped sulphides which are too far apart to have much effect on the ductility. These

sulphides plainly were deposited in crystal interstices but were formed while these interstices were wide and they are not restricted to such a narrow zone.

72.   The dark Al_2S_3 constituent readily is apparent in those steels to which were added 0.05 per cent or more of aluminum. This is shown in Figs. 8 to 10. It also will be observed that, as the aluminum addition is increased, the proportion of Al_2S_3 increases. With increased Al_2S_3, the sulphides tend to become smaller and rounder but the ductility is relatively unaffected.

73.   Alumina is quite soluble in molten Al_2S_3, as shown by the Haglund process for purifying bauxite. Steels containing Al_2S_3 tend to show fewer clusters of alumina inclusions. Apparently the alumina is incorporated in the sulphides.

GRAIN SIZE CONTROL

74.   Most of the steels of Table 3 were examined for grain size and inhibition to grain growth. An interesting relation was observed between sulphur content, aluminum addition and grain size, which indicates that sulphur may play an important part in grain size control. In those steels which contained about 0.020 per cent sulphur, fine grain was obtained only in a very narrow range of aluminum additions. Either more or less aluminum gave coarser grain size. With sulphur contents above 0.030 per cent, almost any addition of aluminum gave a fine grain structure. It may be that aluminum sulphide rather than aluminum oxide is the inhibiting material. Much more work must be done, however, before any definite conclusions can be drawn.

EFFECT OF OTHER DEOXIDIZERS

75.   The question naturally arises, "If it is the absence of oxygen which causes the MnS to form as a fine eutectic, should not other strong deoxidizers have a similar effect?" Fig. 16 is a photomicrograph of heat 2616 to which was added 0.05 per cent zirconium as a zirconium-silicon alloy. Likewise in Fig. 17 is shown the structure of heat 2669 which had an addition of 0.05 per cent titanium as low carbon ferrotitanium. It readily is apparent that both of these steels have the typical eutectic precipitation of sulphides and both had correspondingly low ductility. Young[13] reported low ductility as a result of titanium additions.

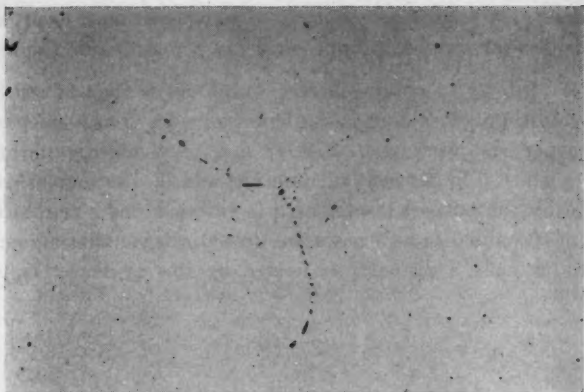76.   These isolated examples are cited merely as additional

FIG. 16—HEAT 2616. DEOXIDIZED WITH 0.05 PER CENT ZIRCONIUM. SHOWS TYPICAL EUTECTIC MnS INCLUSIONS. UNETCHED 250X.

proof that any means of depleting the oxygen in the steel will have a similar effect on the sulphides. Perhaps they also show the need of avoiding critical additions of zirconium and titanium as is true with aluminum. Urban and Chipman[12] described sulphides in a zirconium deoxidized steel that are very similar to the No. III type obtained with excess aluminum, which further emphasizes the analogy. Titanium does not cause low ductility in all concentrations, as is well known, but has shown itself to be an efficient and useful
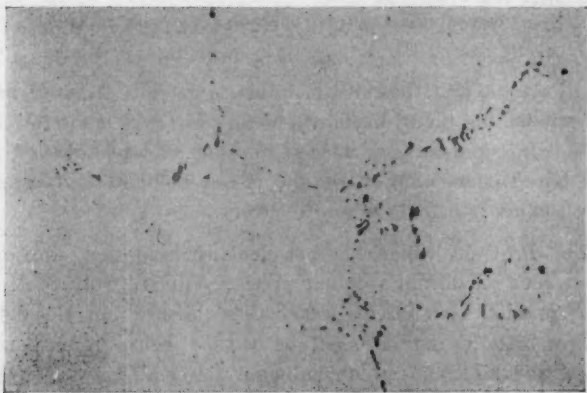

FIG. 17—HEAT 2669. DEOXIDIZED WITH 0.05 PER CENT TITANIUM. SHOWS TYPICAL EUTECTIC MnS INCLUSIONS. UNETCHED 250X.

deoxidizer. Especially in intermediate manganese steel has it shown
to excellent advantage in improving the properties of the steel.

77. Efforts to obtain eutectic sulphides with calcium failed,
when additions of calcium, as calcium silicide, from 0.05 per cent
to 0.15 per cent were used. Fig. 18 is a photomicrograph of heat
3656 to which 0.15 per cent calcium was added. The inclusions have
practically the same character as in a silicon killed steel. Although
calcium is reputed to be a powerful deoxidizer, ranking higher than
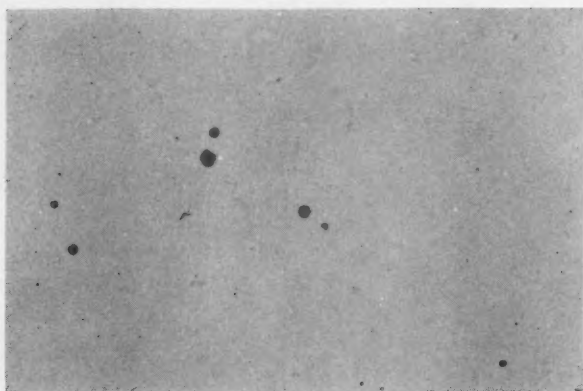aluminum, such a property is denied by this evidence. It is pos-

FIG. 18—HEAT 3656. DEOXIDIZED WITH 0.15 PER CENT CALCIUM. SHOWS NO EFFECT
BEYOND THAT PRODUCED BY SILICON. UNETCHED 250X.

sible, of course, that the calcium failed to get into the steel because
of its low boiling point, inasmuch as no effort was made to forcibly
hold the alloy under the surface of the steel. It was, however, added
to the bare surface of the steel and it was thought the large excess
might compensate for the low recovery.

78. Previous experience has demonstrated that additions of
0.1 per cent vanadium will not produce eutectic sulphides. What
larger quantities would do was not established but the evidence
indicates that vanadium must be ranked below aluminum, zir-
conium and titanium in deoxidizing power. These observations,
naturally, refer only to the relative deoxidizing powers of the
elements mentioned, and have no bearing on other properties.

Commercial Steels

79. The paradoxical effect of aluminum, which first reduces ductility through depletion of FeO in the steel and then raises ductility again by changing the solubility of the sulphides, was demonstrated very positively on experimental heats of steel. In these heats, extraneous variables could be reduced to a greater extent than in most commercial practice which made the intentional variable of aluminum addition more pronounced in its effect.

80. Utilization of this principle commercially gave promise of a means of eliminating the very low ductility sometimes conferred by aluminum treatment. As a part of the study of aluminum deoxidation, a request had been sent to a number of foundries asking for samples of test coupons together with certain pertinent operating data. The response to this request was very gratifying and a large number of bars were received.

81. These steels had been studied in an effort to detect any consistent trend or relation between ductility on one hand and aluminum additions and melting practice on the other hand. No such trend could be detected. They were now studied in regard to the effect of the size of the aluminum additions on the type of inclusions and ductility.

82. A cross section of these heats, which is fairly representative of electric furnace practice, is given in Table 4. A few heats had no aluminum added and these serve roughly as a base line from which to judge the effect of the aluminum additions. Aluminum was added to bottom-pour ladles, bull ladles, hand shanks and, in a few cases, to the metal as it entered the mold. Sometimes there were several additions of aluminum. Most of these steels were examined microscopically and the inclusions classified according to the type designations described earlier. In many cases, two types of inclusions were apparent and both types were listed. It will be observed that, in general, the ductility coincides very closely with the type of inclusions present, the no. I type giving the best ductility, no. III next best and the no. II giving the lowest ductility.

83. There are a few exceptions which might very well be due to factors other than those being considered. There was no case found where a steel with no. II type had good ductility except in

Table 4

COMMERCIAL HEATS OF ALUMINUM KILLED STEELS

Heat No.	Types of Furnace [a]	Heat Treatment [b]	C %	Mn %	Si %	P %	S %	Al Added [c]	Yield Strength lb. per sq. in.	Ultimate Strength lb. per sq. in.	Elong. %	R.A. %	Type of Inclusions
F1608*	A.E.	N 1650°F.	0.26	0.77	0.40	0.030	0.037	0.05 B.P.	47,000	78,500	22.5	28.0	II
F1594*	A.E.	N 1650°F.	0.24	0.74	0.39	0.027	0.034	0.02 B.P.	48,500	76,000	32.0	54.5	I
F1638*	A.E.	N 1750°F.	0.26	0.76	0.41	0.028	0.038	0.02 B.P.	41,500	76,000	32.0	54.0	I
F1641*	A.E.	N 1650°F.	0.25	0.69	0.40	0.027	0.037	0.02 B.P.	41,500	78,500	28.5	49.0	I
F1644*	A.E.	N 1650°F.	0.26	0.72	0.39	0.027	0.037	0.02 B.P.	43,500	80,000	30.0	49.0	I
J18596-1	A.E.	A 1650°F.	0.26	0.60	0.37	......	......	0.08 H.S.	53,000	77,500	16.5	34.0	II & III
J18596-2	A.E.	A 1650°F.	0.26	0.60	0.37	......	......	0.02 B.P.	41,500	76,000	27.0	44.0	II & III
J18604-1	A.E.	A 1650°F.	0.28	0.84	0.44	......	......	0.08 H.S.	57,500	82,000	21.0	46.0	II & III
P-1	A.E.	A 1650°F.	0.24	0.67	0.34	0.031	0.028	0.035 H.S.	52,000	78,500	19.5	27.0	II & III
P-2	A.E.	A 1650°F.	0.24	0.67	0.34	0.021	0.028	0.028 B.L.	48,500	75,000	28.0	46.5	I
W4082	A.E.	N 1650°F.	0.27	0.70	0.39	0.032	0.034	0.016 B.L. +0.04 H.S.	48,500	80,000	29.5	55.5	I
W4997	A.E.	N 1650°F.	0.30	0.64	0.34	0.025	0.029	0.016 B.L.	46,500	81,500	28.0	48.5	I & III
W4997-2	A.E.	N 1650°F.	0.30	0.64	0.34	0.025	0.029	0.016 B.L.	51,500	79,500	26.0	35.5	I
W5018	A.E.	N 1650°F.	0.25	0.75	0.42	0.024	0.030	0.016 B.L.	48,500	78,500	31.5	53.5	I
X1-10	A.E.	N	0.26	0.75	0.35	......	......	0.0125 H.S.	44,500	76,000	29.5	48.0	No specimen received
X1-10	A.E.	N	0.26	0.75	0.35	......	......	0.0250 H.S.	45,000	75,500	24.9	33.5	
X1-10	A.E.	N	0.26	0.75	0.35	......	......	0.0875 H.S.	46,000	77,000	20.0	21.0	
X1-10	A.E.	N	0.26	0.75	0.35	......	......	0.0875 H.S.	47,500	75,000	24.5	29.5	
U484B	A.E.	N 1650°F.	0.27	0.86	0.35	......	......	0.075 H.S.	48,000	80,500	20.5	29.5	III
U500B	A.E.	N 1650°F.	0.25	0.81	0.37	......	......	0.075 H.S.	44,500	75,000	28.0	35.5	III
U512B	A.E.	N 1650°F.	0.22	0.81	0.38	......	......	0.075 H.S.	47,500	76,000	27.5	40.0	III
U581B	A.E.	N 1650°F.	0.27	0.85	0.37	......	......	0.075 H.S.	49,500	82,500	27.5	42.0	III
ZC145	A.E.	A 1350°F.	0.30	0.54	0.40	0.031	0.037	0.030 H.S.	49,000	77,500	27.0	36.5	I
ZC149	A.E.	A 1575°F.	0.27	0.58	0.36	0.036	0.030	0.080 H.S.	48,000	76,500	27.5	43.0	I
ZC152	A.E.	A 1575°F.	0.29	0.48	0.38	0.032	0.035	0.080 H.S.	40,500	76,500	20.5	28.0	II & III
ZC153	A.E.	A 1575°F.	0.29	0.51	0.43	0.031	0.035	0.080 H.S.	45,000	78,000	25.5	39.0	II & III
O-5667	B.E.	A 1650°F.	0.32	0.77	0.33	0.019	0.032	0.05 B.P.	39,500	76,500	26.0	41.5	III
O-7801	B.E.	N 1750°F. 1550°F. 1200°F.	0.35	0.80	0.26	0.006	0.025	0.05 B.P. 0.03 Mold	48,500	74,000	30.0	46.0	III
O-7001	B.E.	N 1750°F. 1550°F. 1200°F.	0.33	0.70	0.41	0.026	0.022	0.05 B.P. 0.03 Mold	43,000	77,500	29.0	55.0	III

*Note: Aluminum added as alsifer in F Series of heats and heat P-2.

a A.E. = Acid Electric; B.E. = Basic Electric.
b N = Normalized; A = Annealed.
c B.P. = Bottom Pour Ladle; H.S. = Shank Ladle; B.L. = Bull Ladle.

very low sulphur steels. When both nos. II and III types of inclusion are present the effect of the no. II type seems to predominate, although the ductilities are often intermediate.

84. A similar relation exists between the type of inclusions and the aluminum addition. In steels having no. I type sulphides, the highest aluminum addition was 0.03 per cent but in most, the additions were not over 0.02 per cent which is equivalent to 5 oz. per ton. The no. III type begins to appear with additions of about .06 % or 18-20 oz./ton.

85. Hand shank and mold additions of aluminum tend to produce mixed types of inclusions while the ladle additions give purer types. The obvious reason for this is the time factor which allows for more uniform distribution with ladle additions. It is true, however, that border-line additions of aluminum to the ladle sometimes gave mixed types of inclusions.

86. There was a consistent difference noted between the laboratory and the commercial heats; namely, greater additions of aluminum were required to reach the same endpoint in the commercial heats. This is thought to be due primarily to lower efficiency in recovery which, in turn, may be due to such factors as longer time of holding in contact with slag and ladle lining, and less uniform mixing.

87. A number of selected steels from both laboratory and commercial heats were analyzed for soluble, insoluble and total aluminum content, on the assumption that the soluble aluminum represented the unoxidized portion (including Al as Al_2S_3) and the insoluble represented the aluminum present as Al_2O_3. The results of these analyses are given in Table 5 together with the size and manner of aluminum addition and the type of inclusions formed.

88. The Al_2O_3 content seems to bear little relation to any other factor in the commercial heats. It merely represents what was left after part of it was lost to the slag. There is, however, a very interesting relation in the case of eight laboratory heats at the bottom of Table 5. These heats were made under closely controlled and very similar conditions. When the aluminum was added, it was mixed quickly and uniformly, and the steel was poured directly into the molds almost immediately after mixing. Under these

Table 5
CHEMICAL ANALYSIS OF SELECTED COMMERCIAL HEATS
AND SOME LABORATORY HEATS

| | Aluminum* | | Al₂O₃, | Al, | Total Al, | Type of |
Heat No.	Added—Per Cent		%	%	%	Inclusions
F1608	0.05	B.P.	0.009	0.001	0.006	II
J18595-1	0.08	H.S.	0.04	0.054	0.075	II & III
J18595-2	0.02	B.P.	0.007	0.0042	0.008	I & II
J18596-1	0.08	H.S.	0.030	0.061	0.077	II & III
J18596-2	0.02	B.P.	0.011	0.0044	0.010	I & II
P-1	0.028 B.L.-0.035 H.S.		0.038	0.017	0.037	II & III
P-2	0.028	B.L.	0.011	0.009	0,015	I
U484-B	0.075	H.S.	0.050	0.034	0.060	III
U500-B	0.075	H.S.	0.030	0.032	0.048	III
U512-B	0.075	H.S.	0.040	0.046	0.067	III
U581-B	0.075	H.S.	0.020	0.053	0.064	III
W4982	0.016	B.L.	0.001	0.002	0.003	I
W4997	0.016	B.L.	0.014	0.000	0.007	I
W4997-2	0.016 B.L.-0.041 H.S.		0.043	0.010	0.033	II & III
W5013	0.016	B.L.	0.007	0.000	0.004	I
2791	0.05	F.	0.026	0.034	0.048	II & III
2793	0.025	F.	0.023	0.006	0.018	II
2798	0.025	F.	0.028	0.010	0.025	II
2800	0.10	F.	0.035	0.045	0.065	III
2803	0.015	F.	0.006	0.002	0.005	II
2804	0.025	F.	0.023	0.005	0.017	II
2808	0.015	F.	0.009	0.000	0.005	I
2809	0.025	F.	0.018	0.008	0.018	II

*B.P. = Aluminum added to Bottom Pour Ladle; H.S. = Aluminum added to Hand Shank; B.L. = Aluminum added to Bull Ladle; F. = Aluminum added to Furnace.

conditions, the aluminum had little chance to react with any oxygen not actually contained in the steel and there was little time for the escape of any alumina formed by the reaction.

89. It is not unexpected, therefore, to find a fairly uniform content of alumina in those steels to which an excess (more than 0.015 per cent) of aluminum was added. The average Al₂O₃ content of these six heats is 0.0255 per cent. This is equivalent to 0.0135 per cent aluminum. The oxidation of this amount of aluminum requires 0.055 per cent FeO, which is the amount of FeO in equil-

ibrium with 0.05 per cent silicon. Even though 0.35 per cent silicon was introduced and allowed to react before the aluminum was added, the latter assumed the entire burden of deoxidation. The exactness of these figures is regarded as fortuitous but the principle is illustrated.

90.   The soluble or residual aluminum contents show a direct relationship to the type of sulphide inclusions. This applies to both the laboratory heats and the commercial heats, although the latter show more irregularities because of the various practices. This relation is shown graphically in Fig. 19.

91.   Although the range of residual aluminum contents in which both no. II and III types was found is quite large, due
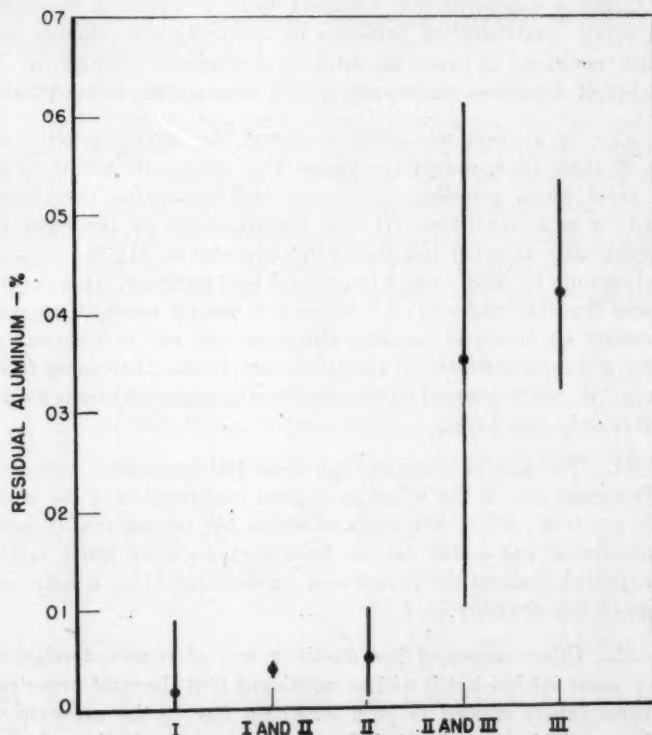
Fig. 19—Relation Between Type of Inclusion Found and Residual Aluminum Content. Showing Range and Average for Heats Tested.

mainly to hand shank additions, the average analyses allow rough conclusions to be drawn as follows:

- (a) The no. I type of inclusions exist where there is practically no residual aluminum but some FeO.
- (b) The no. II type appear with the virtual elimination of FeO and the first traces of residual aluminum.
- (c) The no. III type appear and occur with the no. II when the residual aluminum is from 0.01 to 0.03 per cent.
- (d) A residual aluminum content of about 0.04 per cent practically assures the no. III type of inclusions.

COMMERCIAL TEST HEATS

92. It is believed that the value of residual aluminum in producing a comparatively harmless type of sulphide inclusion was firmly established in principle by the foregoing evidence but it still remained to prove its value in commercial application. A number of foundries generously agreed to cooperate in such tests.

93. No attempt was made to change the current practice in any of these tests, except to suggest that aluminum additions to the ladle would probably give more uniform results than hand shank or mold additions. It was recommended on the basis of previous data that not less than 0.075 per cent or $1\frac{1}{2}$ lb. per ton, of aluminum be used, except for comparison purposes. It was considered that 0.05 per cent or 1 lb. per ton was too small an amount to ensure an adequate residual aluminum content in commercial heats, although sufficient in the laboratory heats. Increasing from $1\frac{1}{2}$ to 2 lb. per ton would be of some benefit, larger additions would neither help nor harm.

94. The results obtained from these test heats were varied in their nature but on the whole gave good confirmation of the principle involved. All of the steels to which $1\frac{1}{2}$ lb. per ton or more of aluminum was added did not have high ductility which might be expected, because the presence of eutectic sulphides is only one cause of low ductility.

95. Other causes of low ductility are, of course, foreign to the present subject but it will be mentioned that the most prevalent of these causes seemed to be a condition due to the presence of hydrogen. This type of embrittlement appears erratically and may be present whether or not aluminum is used. It can be detected in

the fractured surface by the presence of small spots which appear to be finely crystalline. This trouble usually can be remedied by a simple aging treatment or a draw treatment at a temperature too low to affect the strength of a normalized steel. Several hours at 400 to 700°F. is sufficient.

96. Whenever a residual aluminum content was obtained, the no. III type of inclusions were present but in a few cases there was a persistence of the no. II eutectic type. A comparison of two heats made in the same foundry, using ostensibly the same practice and with the same aluminum addition, is shown in Fig. 20. Heat J19490 has no. III type of inclusions but they are small and resemble the no. II type, whereas heat J19600 has the large distinctive no. III type of sulphides. The latter also has higher ductility. The reason for variations of this kind in two such similar heats has not been determined fully but probably is due to some such variable as temperature.

97. The proper way to evaluate the effect of aluminum additions, is to compare the same or similar steels with and without the addition of aluminum or with the addition of varying quantities of aluminum. This is done in Table 6 where steels from four different types of furnace are compared. The high frequency induction furnace heats at the top of the table were made in a commercial laboratory but the rest were made in regular production units in plants where the melting practice is known to be under close control.

Table 6
EFFECT OF ALUMINUM IN REPRESENTATIVE COMMERCIAL
TEST HEATS MADE IN FOUR TYPES OF MELTING UNITS

Heat No.	Type of Furnace	C. %	Mn. %	Si. %	P. %	S. %	Al.* Added %	Yield Strength, lb. per sq. in.	Ultimate Strength, lb. per sq. in.	Elong., %	R.A. %	Izod Impact Str'gth, ft.-lbs.
1482	100 lb. Acid	0.24	0.63	0.36	0.015	0.032	......	42,500	71,500	36.0	58.5	42.0
1469	Induction	0.24	0.65	0.38	0.014	0.030	0.10	49,000	73,500	36.0	56.5	46.5
2284	6 Ton	0.26	0.75	0.40	0.027	0.032	......	49,000	81,500	32.0	55.0	29.5
2282	Acid	0.26	0.63	0.36	0.027	0.032	0.03**	48,500	78,500	32.5	55.5	30.0
1608	Electric	0.26	0.77	0.40	0.030	0.037	0.05	47,000	78,500	22.5	26.0	......
2283		0.26	0.79	0.40	0.027	0.032	0.10	51,000	80,000	28.0	41.0	34.0
300-1	Acid Open	0.27	0.73	0.38	0.034	0.038	......	49,000	80,500	32.5	53.5	25.0
300-2	Hearth	0.27	0.73	0.38	0.034	0.038	0.10	50,000	78,500	28.5	42.0	32.0
3832	Basic	0.29	0.72	0.36	0.022	0.032	......	47,000	82,500	32.0	55.5	29.5
3821	Open Hearth	0.26	0.72	0.37	0.020	0.034	0.10	52,000	76,500	31.5	44.0	37.0

* Aluminum added as ladle additions except for the induction furnace where it was added in the furnace.
** Aluminum added as Alsifer.

98. In the case of the induction furnace heats, the ductility is just as good in the aluminum-treated steel as in the control heat. Low compositions favor high ductility in both of these steels.

99. In the acid-electric arc furnace, the addition of 0.02 per
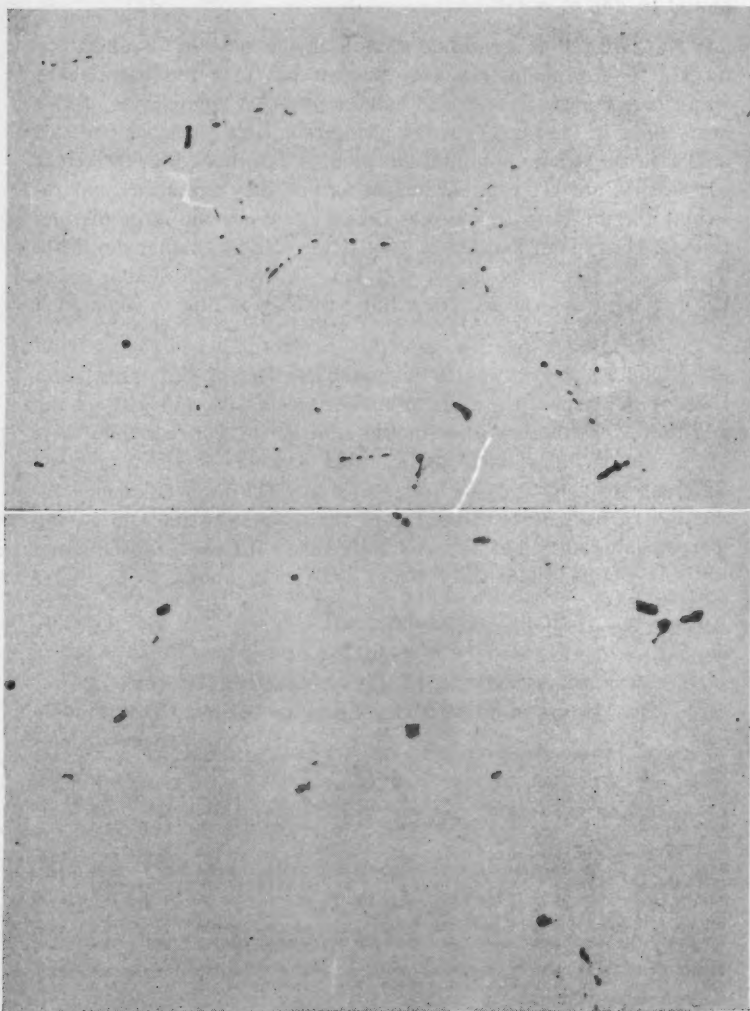

FIG. 20—TOP—HEAT 19490. BOTTOM—HEAT 19600. BOTH HEATS DEOXIDIZED WITH 0.10 PER CENT ALUMINUM. UNETCHED 250X.

cent (5 oz. per ton) aluminum produced no change in the ductility and as shown in Fig. 21 did not materially alter the inclusions. An addition of 0.05 per cent (1 lb. per ton) aluminum, caused a sharp drop in the ductility and a corresponding change to the
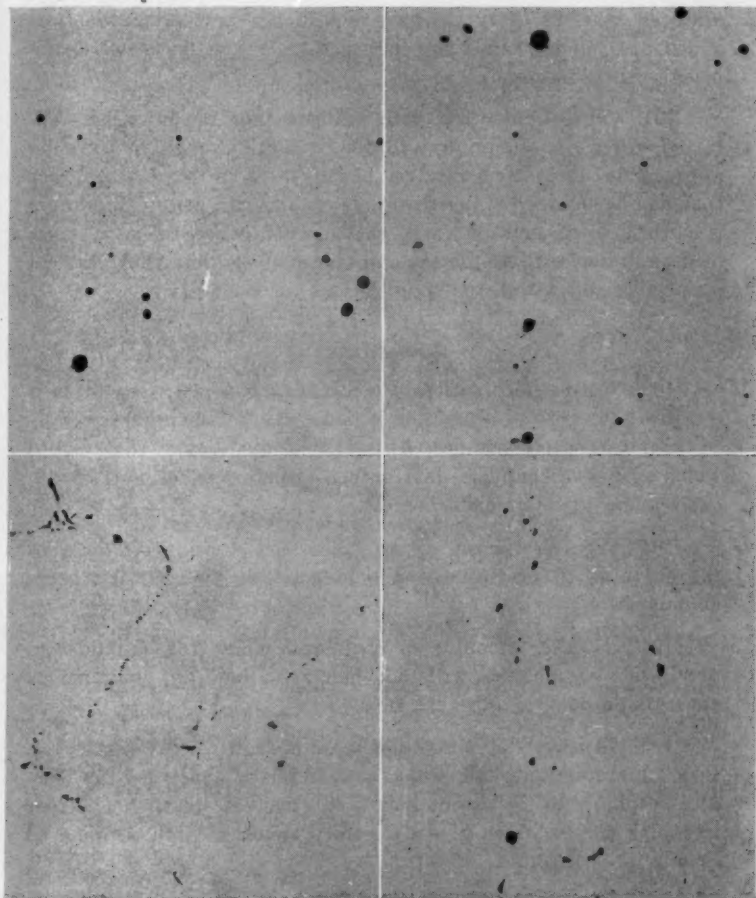
Fig. 21—Acid Electric Furnace Heats. Left Top—Heat 2284—No Aluminum Added. Type I Inclusions. Right Top—Heat 2282—0.02 Per Cent Aluminum Added as Alsifer. Type I Inclusions. Left Bottom—Heat F1608—0.05 Per Cent Aluminum Added. Type II Inclusions. Right Bottom—Heat 2283—0.16 Per Cent Aluminum Added. Type III Inclusions. Unetched 200X.

no. II type of inclusions. An addition of 0.10 per cent (2 lb. per ton) aluminum partially restored the ductility and further changed the inclusions to the no. III type.

100.   The acid and basic open-hearth heats were almost exact duplications of the first and last electric furnace heats. It will be observed that, in every case where 0.10 per cent aluminum was added, the impact values are higher than in the steels not deoxidized with aluminum.

101.   These and other data indicate that the ductilities obtained, when an aluminum addition is large enough to leave an appreciable unoxidized residue, usually will be somewhat lower than can be obtained under favorable conditions when no aluminum is used. They do show, however, that one can obtain the advantages of deoxidation with aluminum and obtain steels that have physical properties well within the specifications for their grade.

SUMMARY

102.   In the deoxidation of medium carbon cast steel, there is a critical quantity of aluminum which will produce minimum ductility and impact resistance. This critical quantity is the amount which will give complete deoxidation (elimination of FeO) without leaving an appreciable excess.

103.   The mechanism of producing low ductility is to cause the sulphide to precipitate as a eutectic in the primary grain boundaries.

104.   Smaller amounts of aluminum allow the sulphides to precipitate as globules with no change in ductility, but s u c h amounts do not deoxidize the steel.

105.   Amounts of aluminum large enough to leave an excess, form aluminum sulphide which lowers the solubility of the complex sulphides to the extent that they precipitate earlier as larger irregular masses which give a ductility almost as high as when no aluminum is used.

106.   Sulphur quantitatively reduces ductility, regardless of whether or not aluminum is used.

107.   In low sulphur cast steels (below 0.025 per cent sulphur), aluminum additions of any amount are practically without effect on the ductility but may increase the impact resistance.

108. Manganese, up to 1.50 per cent, affects strength but has almost no effect on ductility.

109. The effect of size of aluminum addition outweighs the effect of variations in furnace practice, although the latter can affect ductility directly.

110. Under some controlled conditions, even with high sulphur content, aluminum additions in slight excess are practically without deleterious effect on ductility.

111. The effect of the aluminum often is masked partially by other factors which affect ductility, the most prevalent of which seems to be a type of embrittlement due to the presence of hydrogen.

Acknowledgements

112. The authors wish to express grateful acknowledgment to Dr. F. C. Frary for helpful suggestions and for the aluminum determinations and to those foundries who went out of their way to furnish specimens and data and to make special tests.

Bibliography

1—Rogers: Moniteur Industrial, 1859, p. 2379.

2—McCrae, J. V. and Dowdell, R. L.: *"Deoxidation and Mold Conditions on the Tensile Properties of Carbon Steel Castings,"* Trans. A.S.S.T., vol. 18, 1930, p. 159.

3—Melmoth, F. A.: *"Variables in Steel Foundry Practice,"* Trans. A.F.A., vol. 36, 1928, p. 323.

4—Batty, George:*"The Interrelationship of Pin Hole Trouble and the Low Ductility Problem,"* Trans. A.F.A., vol. 39, 1931, p. 861.

5—Comstock, G. F.: *"Aluminum and Titanium as Deoxidizers,"* Iron Age, Dec. 4, 1924, p. 1477.

6—Sims, C. E. and Lillieqvist, G. A.: *"Inclusions—Their Effect, Solubility, and Control in Cast Steel,"* Trans. A.I.M.M.E., 1932.

7—Chipman, John and La Li: *"Equilibrium in the Reaction of Hydrogen with Iron Sulphide in Liquid Iron and the Thermodynamics of Desulphurization,"* Trans. A.S.M., 1936.

8—Wohrman, C. R.: *"Inclusions in Iron,"* Trans. A.S.S.T., vol. XIV, 1928, chapter V.

9—Wentrup, Hans: *"Die Bildung von Einschlüssen im Stahl,"* Technische Mitteilungen Krupp, 5 (1937), pp. 131-152.

10—Löfquist, Helge: *"Sulphide Inclusions in Iron and Steel and*

Their Equilibria," TEK. TID., vol. 67 May 8, 1937, pp. 37-44. Abstract, METALS & ALLOYS, Sept. 1937, MA544.

11—Chipman, John: "*Application of Thermodynamics to the Deoxidation of Liquid Steel,*" TRANS. A.S.S.T., 1933.

12—Urban, Stephan F. and Chipman, John: "*Nonmetallic Inclusions in Steel, Part II Sulphides,*" TRANS. A.S.M., 1935, pp. 645-671.

13—Young, E. R.: "*Making Steel Castings to Specifications,*" TRANS. A.F.A., vol. 29, 1921, p. 245.

14—Melmoth, F. A.: "*Some Metallurgical Points on Acid Open-Hearth and Electric Steel for Castings,*" TRANS. A.F.A., vol. 39, 1931, p. 458.

15—Portevin, Albert and Bastien, Paul: French Exchange paper of 1936, Inst. Brit. Foundrymen.

16—JERNKONTORETS ANNALER, No. 7, 1936, p. 311.

17—Portevin, Albert: METAL PROGRESS, Dec. 1936, p. 50.

18—Guedras, M.: "*The Cause of Pin Holes in Cast Steel,*" REVUE FONDERIE MODERNE, vol. 31, Mar. 10, 1937, pp. 74-75. Abst. METALS AND ALLOYS, Sept. 1937, MA544.

DISCUSSION

J. B. CAINE[1] *(Written discussion):* Mr. Sims and Mr. Dahle are to be congratulated on their important contribution to the complete understanding of the variables effecting cast steel. The importance of this paper cannot be overemphasized. It is the writer's conviction that the inclusions in cast steel have a greater relative effect on the physical properties than the analysis or heat treatment. If the inclusions are right, it is possible to meet any of the physical specifications now in use with practically any analysis or heat treatment that will give the hardness required, and with aluminum killed steels. There are, however, cases when just an examination of the steel for strings is misleading. This is most evident when investigating larger sections, as it is possible to have widely spaced inclusions, in strings that are too long for the area examined under the microscope at the higher magnifications necessary.

It seems that all sulphide inclusions tend to form strings, regardless of the deoxidation practice, so it is not a difference of kind with good and bad inclusions, but one of degree. Therefore, a quantitative study is necessary.

This can be done by counting the inclusions in a number of fields, when the diameter of the field under the microscope is known. Then the lower the count in comparable sections the less danger there is of having planes of weakness in the steel.

The charts of Figs. 21, 22 and 23 show the relation between the inclusions and the elongation and reduction of area for a series of acid electric heats all deoxidized with aluminum. The number of inclusions

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[1] Metallurgist, Sawbrook Steel Casting Co., Lockland, O.

were varied by melting an deoxidation practice. The inclusions were counted in an average of 15 fields of known diameter and area and the count converted into number of inclusions per square millimeter. The quantitative relation between the ductility and the sulphide inclusions is self evident.

Tests with an inclusion count of more than 120 per square millimeter always show strings, those with less than 75 per square millimeter seldom show strings. The intermediate zone from 75 to 120 per square millimeter sometimes show strings, sometimes not, and it is here where a quantitative study becomes particularly important.

The two heats shown in Fig. 2, page 72, are very interesting. The
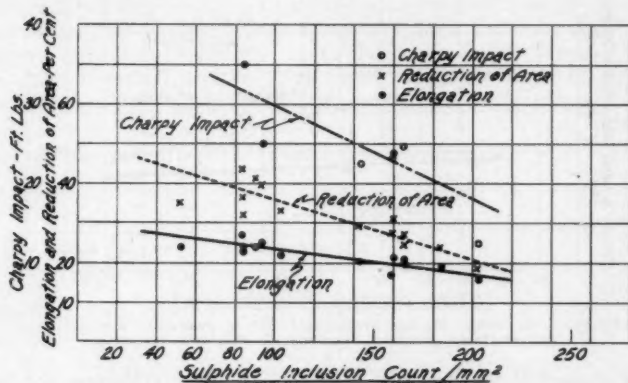
FIG. 21—1035 STEEL—TENSILE STRENGTH 80,000 TO 90,000 LB. PER SQ. IN.—BOTTOM POUR HEATS—POURING TEMPERATURE 2900°F.—ANNEALED.
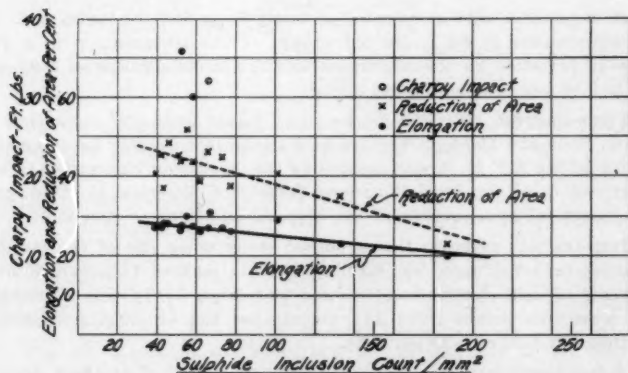
FIG. 22.—1025 STEEL—TENSILE STRENGTH 70,000 TO 80,000 LB. PER SQ. IN.—BOTTOM POUR HEATS—POURING TEMPERATURE 2900°F.—ANNEALED.

difference in the inclusions with the same aluminum addition is thought by the authors to be due to some other variable such as temperature. Temperature is one variable as can be seen by comparing charts, Figs. 22 and 23, of this discussion. However, the major variable is the oxide content of the steel. The tests in this paper prove this conclusively.

The iron oxide content of the steel after deoxidation with aluminum depends not only on the aluminum addition, but also on how much iron oxide was present before the aluminum addition. With the same metal analysis and temperature the controlling factor then becomes the iron
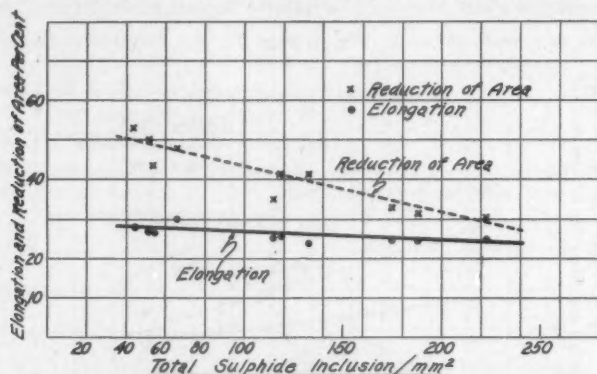
FIG. 23—1025 STEEL—TENSILE STRENGTH 70,000 TO 80,000 LB. PER SQ. IN.—
SHANK HEATS—POURING TEMPERATURE 3050° F.—ANNEALED.

oxide content of the slag. The amount of residual aluminum and iron oxide present in the steel after say a 0.10 per cent aluminum addition will be entirely different if the slag contains a 10 per cent iron oxide or 20 per cent iron oxide.

It is possible with constant slag analysis as well as metal analysis and temperature to see a decided change of the inclusions with a 0.01 per cent variation in the aluminum addition in the range of 0.01 per cent to 0.10 per cent aluminum added.

FRED GROTTS[2] (*Written discussion*): Based upon the researches of Messrs. Sims and Dahle, presented at Battelle last October to a regional meeting of the A.F.A., it was decided by the Operating Executives Group of Division 1, of the Steel Founders' Society of America, to investigate the commercial applicability of the high Al additions to cast steel.

Impetus was given to this proposed study when one of the member foundries reported upon the excellent results secured in a run of over 100 heats of acid electric steel treated with high Al. In this discussion, high aluminum means from 1½ pounds per ton or 0.075 per cent to 4 pounds per ton, or 0.20 per cent.

A technical committee was appointed and a plan of research decided

[2]Chairman, Technical Committee, Division No. 1, Steel Founders' Society of America and Vice President, Lebanon Steel Castings Co., Lebanon, Pa.

on which consisted of specific detail to be covered by each of the partici-
pating foundries. This work was summarized and gave such consistently
good results that it was decided to present the data as a discussion on
this paper as presented here at this Cleveland meeting.

In this investigation it was brought out that good steel making
practice is essential. A majority of the foundries agreed that it was
necessary to boil the charge to a carbon content of about 10 points. A
suitable ore being used in this reaction, followed by pig iron for re-
carburization. One research showed boiling to about 14 points of carbon,
followed by ferro silicon addition to stop the reaction.

It was also reported by all investigators that normal fluidity was not
impaired, tendency to hot tearing was not increased, porosity was not
encountered, no trouble in welding or machining had been found and
that an observance of castings, plus some rough comparative tests, does
not show excessive shrinkage. One foundry does report, however, that
Al reacts on the ladle linings and spout, thus increasing the slag.

The analysis range for this research was as follows:

C%	Mn%	Si%	S%	P%
0.23-0.28	0.65-0.75	0.28-0.33	0.04	0.04

Dodge Steel Company, Philadelphia, Pa.
Average Properties

Tensile Strength per sq. in.	Yield Point per sq. in.	Elongation per cent	Reduction of Area per cent	Al Added per cent
72,500	41,750	30.5	46.5	.125 or .10

The Al is placed on bottom of the ladle and heat tapped on it. The
castings and bars are given a normalize or a fast anneal, depending on
carbon content and physicals desired.

Malleable Iron Fittings Co., Branford, Conn.
Average Properties

Tensile Strength per sq. in.	Yield Point per sq. in.	Elongation per cent	Reduction of Area per cent	Al Added per cent
71,000	41,000	30.5	49.0	0.10

The Al is wired on a rod and added to the furnace before tapping.
Castings and bars are given a normalize and temper.

Lebanon Steel Foundry, Lebanon, Pa.
Average Properties

Tensile Strength per sq. in.		Yield Point per sq. in.	Elongation per cent	Reduction of Area per cent	Al Added per cent
Dry Sand................71,500		42,500	28.5	50.0	0.20 and CaSi
Green Sand........72,500		45,000	27.5	46.5	0.20 and CaSi
Dry Sand................72,000		44,000	32.0	55.0	0.0125 and CaSi

The Al is pushed to the bottom of the ladle on a rod after the
ladle is full. Castings and bars are given a normalize and temper.

It appears from this investigation that test bars in green sand molds

are not quite as good as those in dry sand molds, but are still very satisfactory.

Reading Steel Casting Co., Reading, Penna.
Average Properties

	Tensile Strength per sq. in.	Yield Point per sq. in.	Elongation per cent	Reduction of Area per cent	Al Added per cent
No CaSi	72,000	48,000	29.2	46.2	0.125 to 0.175
With CaSi	74,000	45,350	31.2	54.2	0.125 to 0.175

The Al is added to a full ladle of metal by means of wiring to a rod. Castings and bars are given a normalize and temper.

From these results, which are checked by Lebanon Steel Foundry, it appears that CaSi in addition to the high Al has a very beneficial effect.

It would be interesting to get the author's reaction to this effect of CaSi.

Treadwell Engineering Co., Easton, Penna.
Average Properties

Tensile Strength per sq. in.	Yield Point per sq. in.	Elongation per cent	Reduction of Area per cent	Al Added per cent
72,000	40,300	30.0	48.5	0.125

The Al is plunged to the bottom of the ladle when nearly all the metal is in and the slag begins to flow. Castings and bars are given a normalize and temper.

In observing these physical properties, it is rather remarkable how all heats from five foundries are in such excellent agreement, giving such good results after adding quantities of Al varying from two to four pounds per ton. Past experience has always indicated poor physicals and woody fractures.

In summarizing these data, there seems to be some real value in connection with this use of high aluminum as researched by Sims and Dahle. It has been demonstrated conclusively that good physical properties can be obtained if good melting practice is observed. The group of foundries presenting this discussion surely congratulate the authors on their important contribution to steel foundry practice.

G. A. LILLIEQVIST[3] (*Written discussion*): I have read with a great deal of interest this paper and wish to congratulate the authors for their splendid contribution to our knowledge of steel-making. They have given us a clear picture showing how different amounts of aluminum and other special deoxidizers will affect the physical properties of medium carbon cast steel and how these properties can be correlated with the types of inclusions. A number of questions which have previously bothered me in regard to the addition of special deoxidizers have been cleared up by this paper. To gain some additional information, however, I would like to ask the authors several questions which in my opinion are of great importance.

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[3] American Steel Foundries, East Chicago, Ind.

(1) The heats made in the induction furnace were drawn for two hours at 750°F. after the normalizing treatment; and I would like to ask the authors if the commercial heats, made in the different types of furnaces, were also drawn at 750°F. for two hours. In our laboratory it is a regular practice to age all of our bars, after machining, for 15 hours at a temperature of approximately 400°F. This practice is followed to eliminate a variable which might result in misleading conclusions. I know from many commercial heats that an aging treatment after machining will improve the ductility in many cases to a great extent. It has been my belief for some time that poor ductility obtained from bars without the aging treatment was due to some hydrogen absorption, but never have had positive evidence. Have the authors some more substantiating proof besides the observation of the small spots in the fractured surface of a test bar.

(2) I was very much interested in melting practice No. 7 where the residual silicon was carried through the heat from the cold charge. I would like to ask the authors for some more details regarding the means which were taken to maintain the high silicon content throughout the melt.

(3) I agree with the authors that the type of furnaces and the method of finishing heats will affect to a great degree the physical properties as influenced by the addition of different amounts of aluminum. In my experience the temperature of the steel, before the deoxidizers including aluminum are added, is of great importance. The higher the temperature, the better are the results, especially in regard to Izod impact.

A few years ago we made experimental heats in our induction furnace, the carbon being from 0.50 to 0.65 per cent manganese, 0.70 per cent silicon 0.40 per cent and sulphur 0.030 per cent. Some of the heats were deoxidized with ferro-silicon and ferro-manganese when the bath was on the cold side and then heated to the proper pouring temperature; while other heats were deoxidized in the same manner when the bath was rather hot. No other deoxidizers were used. In all cases we found that the steel had the eutectic type of inclusions and correspondingly low ductility when the deoxidation was made to the cold bath. Do the authors believe that the low ductility of the cold heats was due to the absence of oxygen as iron oxide, or do they have some other explanation?

(4) In regard to Fig. 13—heat No. 3649—I would appreciate very much knowing the physical properties. I would expect from looking at this photomicrograph a very low ductility, not only due to the high sulphur content, but mainly due to the arrangement of the inclusions.

(5) In Fig. 17, heat No. 2669, the authors show typical eutectic manganese sulphide inclusions due to the addition of 0.05 per cent titaniumas low carbon ferro-titanium. We have made a number of heats of grade "B" steel and certain alloy steels, and have found that the addition of the same amount of titanium, added as high carbon ferro-

titanium, will give the round type of inclusions and, therefore, good ductility. In similar steels, low carbon ferro-titanium and the so-called foundry ferro-titanium gave decidedly poorer ductility and showed the presence of the eutectic inclusions.

(6) I was very much interested in paragraph 89, in which the authors say: "Even though 0.35 per cent silicon was introduced and allowed to react before the aluminum was added, the latter assumed the entire burden of deoxidation." Does this mean that is might be possible to kill completely a bath of steel with approximately ½ lb. of aluminum per ton without the addition of ferrosilicon?

Again, I wish to congratulate the authors for their splendid work and hope that some additional papers will be published to further our knowledge in regard to the behavior of these special deoxidizers.

R. H. FRANK[4] *(Written discussion)*: The writers are certainly to be congratulated on their precise manner in handling this moot question and also their extensive laboratory research work. While I do not believe their findings are a panacea from a practical standpoint, they have gone a long way in clearing up many of the disputed questions in relation to low the ductility problems encountered by the use of aluminum.

Their explanations of the phenomena in the recovery of ductility by the use of larger amounts of aluminum are clear, concise and easy to follow. It gives us the general idea to follow or what to expect, yet on the other hand there seems to still be a doubt in my mind as to the consistency of the results. In Fig. 3 the curves for 0.04 per cent S show a considerable scattering of the points. The authors mention their erratic results up to the addition of 0.025 per cent Al but it seems as if there is more scattering on all the curves, especially at approximately 0.075 per cent Al than on any of the charts for higher or lower sulphur. There were more experimental heats made in this range of approximately 0.04 S probably due to the fact that it is closer to commercial practice. Would it not be logical then to assume that, if more heats were made in the other range of sulphur, there would also be a scattering or erratic results on these also? If not, are we then to assume that approximately 0.04 per cent S is a critical range which is to be avoided, if we expect to get consistent results in respect to ductility in aluminum treated steels? In other words, while we know that higher sulphurs give lower ductility, we should keep our sulphurs either above or below approximately 0.04 per cent S, if we are to add aluminum in any percentage and expect to get consistent results in our ductility figures. I would like to see the authors extend their experiments especially in the 0.03 per cent S and 0.06 per cent S and probably add another set in the range of 0.05 per cent S. This may go a long way in alleviating the doubt in so many of our minds as to the consistency of the results as usually obtained in commercial practice.

The authors make no mention as to the type of fractures on their tensile bars. It has always been observed that on aluminum treated steels, whether they show either high or low ductility, the fractures

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show a woody structure with numerous tears on the sides of the bar not necessarily near the break. On a few occasions this has also been observed on steels that have not been treated with aluminum. In every case however, the inclusions were of the chain type, designated as No. II, or the cluster type. It has been my contention that these tears observed on the sides of the bars are the direct result of these chain type inclusions situated on the surface of the bar and, since these areas are weaker, they cause a premature yielding with a resulting lower ducitlity.

In paragraph 64, the statement is made that—"it requires just as much aluminum to completely deoxidize a steel already deoxidized with silicon, as it does when no silicon is used." Complete deoxidation then means, as the authors state, the removal of *all* oxides. We have all read, in the past about over-reduced steels and that they should be avoided because they produce chain type inclusions with the resulting low ductility. Also they can be avoided by adding ore or rusty scrap late in the heat. This so-called over-reduced condition is observed by sluggish metal, low film count, more shrinkage and piping in the risers, and by stringers forming on the nozzle brick during bottom pouring. By the use of sufficient aluminum to cause an excess over what is needed for deoxidation, this steel would then be over-reduced and we should expect some of the deleterious effects as mentioned above. From my own observations and from talking to others, this is just what happens when one to two pounds of Al per ton are added in bottom pouring practice. But, on the other hand these effects are not consistent, sometimes they will not be observed at all which again brings up the doubt that, while the authors have increased our knowledge to a considerable extent, there is still something that could explain this inconsistency. While I have not observed this, I would like to ask the authors if the following hypothesis could be correct in explaining some of this inconsistency. If the heat was over-oxidized, that is, in some manner the equilibrium constant for FeO was disturbed so that more than the theoretical amount was in solution, we would expect to have more SIO_2. Therefore, it would take more aluminum to deoxidize this heat and there might not be enough excess to prevent the sulphides from precipitating as a eutectic. In other words, even by using sufficient aluminum, as recommended by the charts, we would still be in the area of low ductility.

Also from this statement in paragraph 64, why should we use silicon at all. If it takes just as much aluminum whether silicon is used or not, we are just throwing the silicon away. As far as we know at the present time, silicon in the usual commercial ranges, has no appreciable effect on the physical properties, so are we to assume the use of silicon is superfluous if we use aluminum?

To further substantiate my oft repeated reference to inconsistency in commercial practice, let us look at heat* designated as X1-10 in Table 4. No sulphurs are given but it can be eliminated as a variable

* EDITORS NOTE: Since this discussion was submitted the authors have corrected typographical errors appearing in Table 4 of the preprint of their paper. The sulphur content for the four heats of X1-10 of Table 4 originally shown as 0.0 per cent, 0.125 per cent, 0.250 per cent and 0.075 respectively have been corrected to read 0.0 per cent, 0.0125 per cent, 0.0250 per cent and 0.0375 per cent respectively.

Table A

Heat No.	C %	Mn %	S %	Si %	Al %		El %	R.A. %	Y.P. lb. per sq. in.	T.S. lb. per sq. in.	Fracture
20079	0.31	1.20	0.035	0.36	0.00	B.P.	23.0	60.8	76650	102040	Silky—full cup
					0.090	H.S.	22.0	52.5	88350	102450	Woody—tears on side
20087	0.29	0.79	0.041	0.39	0.04	B.P.	23.5	50.8	52000	87250	Silky—half-cup
					0.09	H.S.	22.5	37.5	51500	80050	Woody—tears on side
20095	0.28	0.84	0.041	0.88	0.00	B.P.	27.0	51.1	50050	88150	Silky—full up
					0.000	H.S.	28.5	38.5	45600	80050	Woody—tears on side
20108	0.33	1.20	0.035	0.39	0.00	B.P.	22.5	59.6	82750	100850	Silky—full up
					0.000	H.S.	20.0	47.2	87000	105150	Tears on side
20111	0.33	1.22	0.035	0.38	0.00	B.P.	20.0	54.1	88900	105750	Silky—half cup
					0.000	H.S.	20.0	47.8	76850	100040	Tears on side
20130	0.31	1.18	0.036	0.34	0.00	B.P.	22.5	60.6	87500	105000	Silky—full cup
					0.000	H.S.	19.0	43.7	78250	100950	Woody—tears on side
20185	0.26	0.71	0.039	0.30	0.00	B.P.	30.0	52.2	40800	78450	Silky—partial cup
					0.045	H.S.	25.0	38.5	46950	76650	Woody—tears on side
20191	0.27	0.62	0.039	0.26	0.00	B.P.	30.0	46.8	45000	76700	Silky—full up
					0.045	H.S.	20.0	27.2	46650	76550	Woody—tears on side

since all results are apparently from the same heat. The results* with no aluminum added show very good ductility; with 0.075 per cent Al (1½ lbs. per ton) the ductility is very poor; with 0.125 per cent Al (2½ lbs. per ton) the ductility has increased slightly but cannot be called good; with 0.250 per cent Al (5 lbs. per ton) the ductility is the lowest in the group. While it is true that these aluminum additions were made in a hand shank and the authors state this is not the best practice, with which most everyone will agree, yet it seems as if the large addition of aluminum would counteract the time factor involved. Furthermore, the loss in ductility could not very well be from any other cause than the aluminum addition, since they were all from the same heat and with no aluminum it showed good ductility.

A few of our own experimental results with the use of aluminum are given in Table A and, in a general way they conform to the authors' contentions. All heats to which aluminum was added show a reduction in ductility, some quite considerably, others not so much, with a few on the very edge of minimum specifications for certain types of steels. The last heat shown is definitely out of all specifications as is a good example of the inconsistency in commercial applications. While it does not contain the amount of aluminum recommended by the authors, the heat above has the same amount and yet it would pass. All the specimens that had aluminum added showed the typical woody structure on the fracture with numerous tears on the sides.

I wish again to say that the authors should be congratulated on their excellent work in contributing so much knowledge to our much discussed question of low ductility with aluminum additions and hope they will continue their experiments on a commercial basis in the hope that the inconsistency, so often mentioned above, may be eradicated. Furthermore, I hope that the authors will view the remarks in this discussion as constructive and not destructive.

H. D. PHILLIPS[5] (*Written discussion*): Since the preliminary presentation of this work at the A. F. A. meeting in Columbus last Fall, considerable experimentation and discussion thereof has taken place amongst steel founders in the Eastern states, particularly those operating electric arc furnaces, and it has generally been conceded that it is a subject the true worth of which can hardly be over-estimated.

In 1933-34 experiments were made with higher aluminum additions to acid electric steel of medium carbon grade poured with a bottom pour ladle. With the Al added to the ladle at the time of tapping, it was found that an addition of 0.075 per cent Al gave very good ductility values while an addition of 0.05 per cent Al caused a dropping off of elongation and reduction of area.

C %	Mn %	Si %	S %	P %	Al (Added) %	T.S. lb. per sq. in.	Y.P. lb. per sq. in.	El. %	R.A. %
0.31	0.69	0.32	0.041	0.040	0.075	76,000	40,000	32.0	53.0
0.30	0.76	0.33	0.040	0.039	0.050	76,500	48,000	27.0	40.5

These steels were all given a normalize and temper. The analysis and physicals given are the average of 30 heats with 0.075 per cent Al

and 27 heats with 0.05 per cent Al. The higher Al addition proved to be very beneficial, also in the casting of 5 per cent Cr, 0.50 per cent steels. It was thought that in the chromium steels, the possible formation of a chromium sulphide of low solubility might possibly interfere with the formation of the eutectic normally encountered in medium carbon steels treated with Al. This was after the presentation of Sims' and Lillieqvists' paper to the A. I. M. E. in 1932.

While this practice of adding 0.075 per cent Al gave good results, even though it could not be explained at that time, when later it was tried with the extra hot metal used in the pouring of light castings in green sand with hand ladles, an occasional heat would show porosity and others would show low ductility, it being difficult at times to pass "B" specifications.

As a result of this, the practice was discarded with the reservation that it would be investigated again later on, and the conventional practice of adding about 0.03 per cent Al to the heat and an extra addition of 0.01 or 0.02 per cent to the hand ladles was resumed.

There were many experiments which suggested the practice of adding 0.075 per cent Al. One was the combined use of zirconium and Al in fairly large amounts. The first two heats were very successful but later heats would not check. Again, it was found by another investigation that when 5 lb. of Al were added to a 5000 lb. heat of steel while it was still in the furnace, poor ductility and a tendency to porosity resulted, but if the furnace addition were followed with another 5 lb. addition to the ladle, good ductility and casting soundness was the result.

Since last Fall, however, the preliminary presentation of this research has served to crystallize and explain earlier results and since then it has been the practice of the Dodge Steel Company to add 0.10 per cent to 0.15 per cent Al to all heats in the ladle at tap and make no further additions to ladles or molds, and the results have been very gratifying.

Due to less efficient addition, extra transfer and a desire to be commercially safe, it has been found that a minimum of 0.10 per cent Al is required.

Some foundries report a difference between bars poured in green sand and those poured in dry sand when given a simple normalizing treatment. In the opinion of the authors, would this be due to hydrogen embrittlement in the green sand bar? Others report that where test bars are attached at the far end of a green sand casting the bars are not so good as those nearer the gate. Could this be due to a combination of reasons, hydrogen pick-up and oxidation of Al while the metal is traversing the mold? These and a multitude of other questions are suggested.

With the furnace practice followed by the Dodge Steel Company, the test bars poured with 0.10 per cent Al steel, after being given a

simple normalize show 90 to 95 per cent of the ductility values secured from similar steels with no Al added.

The true effect of the variation of sulphur and Al has been clearly demonstrated in actual practice. Basic electric steels of low sulphur content, 0.02 per cent or below, do not have their ductility impaired by high Al additions.

The relation found between sulphur content, Al addition and grain size is certainly worth further investigation. It throws an entirely new light on this subject of grain growth inhibition.

In the casting of the high Al steels, no tendency to low fluid life has been observed. In the macro-examination of casting sections and test bars, however, a coarser dendritic pattern and also deeper etching characteristics have been noted.

The results of the commercial tests are very interesting. They bear out the experience of many that a higher Al addition is necessary in commercial practice than is indicated by the authors' experiments in the induction furnace.

The statement that it requires just as much Al to deoxidize completely a steel already deoxidized with silicon as it does where no silicon is used is rather startling, but it is certainly borne out by the results secured by the authors as well as those secured by McCrae, Dowdell and Jordan in "A Study of the So-called Over-reduced Condition in Molten Steel," published in 1930 by the Bureau of Standards.

I would like finally to congratulate the authors on the excellent work that has been done on this subject which is of such great importance to the manufacturer of steel castings in green sand.

H. H. BLOSJO[6] (*Written discussion*): This paper represents an immense amount of work well done. I believe that quite a large number of metallurgists have recognized the fact that aluminum additions of more than one pound per ton of melt are beneficial when properly added and when ladle and mold conditions are right, but the correlation of the properties with microstructure and especially the cause of these microstructures have never been clearly defined before.

It has been the practice of our foundry for some time to add two pounds aluminum per ton of melt to heats going into definite physical specification work. We came to this practice because we noticed that the physical properties were more consistently good than when one pound of aluminum per ton of melt was added. According to the authors paper one pound aluminum per ton of melt should be sufficient, but as they also state, the efficiency of recovery commercially is usually lower than in laboratory additions. This latter factor is unquestionably responsible for relatively low ductility in some of our heats when one pound aluminum per ton of melt was used.

Last year our company had the occasion to make 16 dredge castings to comply with government specification QQ-S-681a, Class 2. This

specification is rather easily met with the exception perhaps of yield
point which is rather high in proportion to the tensile strength for a
casting that must be given a full anneal. These castings were of uni-
form wall thickness of 2½ in. and weighed 2500 lb. cleaned. Each cast-
ing was poured from a different heat made in a 1½ ton, acid lined, direct
arc electric furnace. The test bars were cast attached to the casting and
were cut to within ½ in. of removal with a gas torch before annealing
and then broken off by the government inspector after annealing. The
heat treatment consisted of heating to 1700°F. and holding furnace at
that temperature for five hours, and then furnace cooling to about
500°F. before removing from furnace. To these heats (5,000 lb.) was
added five pounds of aluminum to the tea pot ladle or two pounds per
ton of melt. It will be noted that the sulphur content was quite con-
sistent in the 0.04 per cent range. The heats were all plain carbon ex-
cept heat No. 813 which had the carbon reduced and a small amount
of copper added. These heats were all produced without any silicon addi-
tions. The silicon being mostly reduced from the slag late in the heat, al-
though the residual silicon at the start of the boil is usually in the
range 0.10 to 0.15 per cent. Table B gives the analysis and physical
properties of these 16 heats.

It will be noted from Fig. 3 of the authors paper that the elonga-
tion and reduction of area with 0.04 per cent sulphur and 0.10 per cent

Table B
ANALYSIS

Heat No.	C. %	Mn. %	Si. %	P. %	S. %	Cu. %
802	0.28	0.72	0.34	0.030	0.035	------
807	0.32	0.76	0.37	0.034	0.036	------
813	0.21	0.67	0.31	0.029	0.038	0.38
822	0.29	0.68	0.39	0.033	0.036	------
827	0.29	0.75	0.44	0.034	0.040	------
832	0.30	0.69	0.34	0.031	0.032	------
842	0.28	0.69	0.30	0.036	0.034	------
852	0.28	0.73	0.41	0.031	0.040	------
859	0.25	0.66	0.43	0.033	0.043	------
864	0.25	0.76	0.35	0.034	0.040	------
869	0.26	0.76	0.45	0.034	0.041	------
874	0.32	0.73	0.38	0.040	0.041	------
879	0.29	0.68	0.34	0.035	0.040	------
884	0.29	0.75	0.40	0.038	0.036	------
893	0.33	0.72	0.39	0.036	0.042	------
898	0.27	0.77	0.43	0.035	0.041	------

PHYSICAL PROPERTIES

Heat No.	Yield Point lb. per sq. in.	Tensile Strength lb. per sq. in.	Elong. %	R.A. %	Bend Test degrees
802	41,100	73,350	31.5	45.2	180 +
807	42,900	77,300	29.5	47.7	Not tested
813	41,950	70,500	33.5	54.1	180 +
822	42,750	76,100	28.0	40.6	180 +
827	46,950	78,950	30.5	45.7	180 +
832	41,850	77,500	29.0	45.4	180 +
842	42,500	71,500	31.0	46.8	180 +
852	41,100	74,350	31.0	48.4	180 +
859	40,000	74,100	30.0	44.9	180 +
864	40,650	72,050	32.5	47.2	Not tested
869	44,250	74,900	31.0	47.2	Not tested
874	45,400	79,850	29.0	43.9	180 +
879	40,500	74,650	28.0	41.8	Not tested
884	43,400	79,850	30.5	47.7	Not tested
893	49,550	85,100	27.5	39.1	150
898	42,900	75,750	30.5	47.7	Not tested
Average	43,000	75,900	30.2	45.8	
Required	40,000	70,000	24.0	30.0	120 d

aluminum addition, averaged about 30 per cent and 47 per cent respectively. This checks very closely with the results we obtained on these 16 heats, which was an averege elongation of 30.2 per cent and an average reduction of area of 45.8 per cent. There was a difference in the heat treatment of the authors heats and ours, but a normalize followed by a 750°F. draw should give very nearly the same ductility as a full anneal although the yield point especially and the tensile strength to some extent will be increased with the normalize and draw.

The authors should be highly commended for their excellent contribution on a relatively misunderstood subject.

GEO. F. COMSTOCK[7] AND W. G. WELLING[7] (*Written discussion*): This paper is an interesting and valuable record of quite a large number of tests of aluminum-treated steels, and the importance of the sulphur content of the steel in its relation to the effect of aluminum on the ductility is well brought out. Although the authors should be congratulated on their success in showing that an excess of aluminum over the amount generally used for deoxidation did not progressively lower the ductility but caused the ductility to rise above a minimum value, it might perhaps be emphasized, somewhat more than is done in the paper, that the ductility was not brought back completely, with higher aluminum additions, to the value found for the steel without aluminum.

Foundrymen should probably proceed with caution in attempting to apply the teachings of this paper to regular practice, since the effects of the large aluminum additions may not work out the same in steel made in various ways at different plants. The results from commercial heats in Table 4, for instance, especially those from the X series of heats, do not seem to support very well the theory that excessively large aluminum additions are not detrimental. It is noteworthy also that in this table results from only two untreated steels are given, and one of these shows the best ductility values of its series. These commercial results in fact seem to prove that if the aluminum addition is sufficiently limited in amount it may be safe as regards ductility, rather than that larger aluminum additions may safely be used.

The authors' very reasonable remarks regarding titanium are appreciated, but we should like to point out that the inferior type of sulphide inclusions, as illustrated in Fig. 17, was found in steel treated with low-carbon ferro-titanium, which undoubtedly contained a significant amount of aluminum, whereas the ferro-titanium generally recommended and used for deoxidation of cast steel is a high-carbon, low-aluminum alloy marketed by our firm which is the alloy which, as the authors state, has "shown itself to be an efficient and useful deoxidizer," and it is unfortunate that the distinction between the two kinds of ferro-titanium is not made in the paper.

As an interesting discussion of inclusion types and their effects on the properties of steel this paper is highly commendable and worthy of careful study.

[7] Metallurgist and Development Engineer respectively, Titanium Alloy Manufacturing Co., Niagara Falls, N. Y.

JOHN CHIPMAN[8] (*Submitted as written discussion*): The authors have made a very important contribution to our understanding of the function of aluminum in steel making and have cleared up a good deal of the mystery that has formerly attended the use of this element. The interaction of oxides and sulphides has been shown to be a controlling factor in determining the type of inclusions formed in the steel and the ductility of the casting.

The explanation of the effect of oxygen in apparently decreasing the solubility of sulphides in steel probably lies in the fact that the non-metallic phase is not a simple sulphide when any appreciable amount of FeO or MnO is present, but a complex solution of FeO—MnO—FeS —MnS. It may be considered that the sulphides are partially extracted from the liquid steel by this complex solution.

The authors' explanation of the observed effects of aluminum as set forth in paragraphs 51 to 73 appears to be entirely sound. It is noteworthy in this connection that recent work on the deoxidizing power of silicon has shown it to be a weaker deoxidizer at high temperatures than was heretofore supposed. These new results give added support to the contention of Sims and Lilleqvist that silicates are appreciably soluble in liquid steel. They emphasize again the great difference between silicon and aluminum and contribute to the certainty with which we may accept the explanation given in paragraph 64.

The suggestion that aluminum sulphide may have an influence upon grain size is new and contains interesting possibilities. It is to be hoped that the authors will continue their work along this line.

S. F. URBAN[9] (*Written discussion*): The authors are to be highly congratulated for presenting in so precise a manner the effect of aluminum on the ductility of cast carbon steels. This is particularly worthy since there has been considerable controversy on the effect of this element. With the data presented by the authors it is very evident that all the parties in the controversy on this subject were probably correct namely, that the additions of aluminum yield steels of poor ductility or excellent ductility. In the light of the data presented by the authors it is evident that these effects are obtained by variations in the amount of aluminum added. In this connection it is hoped that these data will counteract the statement that aluminum killed steels are "doped" and therefore are of inferior characteristics.

It is the writer's opinion that the authors have demonstrated the possibility of showing the difference in the solubility of various non-metallic inclusions by indirect methods. Although this is of no immediate commercial importance, the approach is very worthy of further consideration. For example it is well established that ferrous sulphide is considerably soluble in liquid iron, however, its solubility in the solid

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[8] Associate Professor of metallurgy, Massachusetts Institute of Technology, Cambridge, Mass.
[9] Carnegie Illinois Steel Corporation.

state, even at elevated temperatures, is considered to be small. The only direct evidence on this subject is that by Whitely,* poured molten iron of various sulphur contents into water and examined the resultant pellets. By such a series of tests Whitely was able to show that sulphide inclusions were evident when the sulphur content exceeded 0.03 per cent. These data are of course on ferrous sulphide which is believed to be considerably more soluble than other sulphides. It is therefore apparent that it would be an enormous task to measure the difference in solubility of various sulphides in iron by direct methods. The data presented by the authors show conclusively that a solubility of the sulphides in iron can be varied to a considerable degree by the addition of other elements such as aluminum.

In paragraph 64, the authors discuss a reaction $Si + 2FeO = SiO_2 + 2Fe$, and state that for a given residual Si and FeO a given amount of SiO_2 will co-exist, and that the subsequent addition of aluminum will cause above reaction to go from the right to the left to the extent that aluminum has combined with FeO. Since aluminum is a much more powerful deoxidizer than silicon, it appears that the statement by the authors that "it requires just as much aluminum to deoxidize completely a steel already deoxidized with silicon as it does when no silicon is used, "is correct. If this is to be accepted without any reservations it appears to the writer that it is not necessary to add the nominal amount of silicon to steels that are subsequently dead killed with aluminum. Unless there are specific advantages to be obtained from silicon which are not evident at this time.

T. N. Armstrong[10] *(Written discussion):* It seems superfluous to compliment the authors as the high quality of the work is so obvious, and the subject is one of paramount interest to the steel foundry industry. This paper and an earlier one by Sims and Lillieqvist[1] would be valuable if for no other reason than to direct attention to the fact that steel suitable for ingots may not be suitable for castings. With this in mind, it appears expedient for the foundry industry to proceed cautiously before adopting melting technique developed specifically for steel that is to be finished by hot working.

The authors have presented their conclusions in such a manner as to leave little room for criticism. It is believed, however, that slight clarification of several points may prevent wrong interpretation of the role aluminum plays in cast steel.

In their explanation of the aluminum effect, the authors state in substance that small additions of aluminum deprive the metal of oxygen by directly or indirectly reducing FeO, MnO and SiO_2. Depriving the steel of iron and manganese oxides, therefore, causes the eutectic type of sulphide inclusion to form, as reduction in oxygen content increases the solubility of iron-manganese sulphide. After combining with all of

the available oxygen, part of the aluminum, in excess of that required for deoxidation, will form complex sulphides with iron and manganese, lowering the solubility of the sulphides, thereby decreasing the tendency toward sulphide network.

This explanation of ductility recovery, based upon aluminum addition in excess of the quantity required for deoxidation, infers that oxygen content of the bath prior to the aluminum addition is relatively unimportant, providing sufficient aluminum is subsequently added. The authors state, "Thus, in spite of the absence of oxygen, Al_2S_3 tends to prevent the formation of a eutectic and causes the sulphides to form large, irregular-shaped sulphides which are too far apart to have much effect on the ductility." If such be the case, the question is raised as to the advantages of the widely used practice of melting under oxidizing conditions, other than to reduce the solubility of gasses by reducing the residual silicon. Also, can good ductility in cast steel be assured simply by the addition of 0.07 per cent or more of aluminum?

The photomicrographs shown in the paper are unusually good. Because of the wide interest in this subject, it is believed that it would be helpful if the authors would give some description of the polishing technique. Many foundries encounter the low ductility problem but when specimens are examined under the microscope, there is no indication of the sulphide network. The fact that it is there is proven when the same specimens are prepared by individuals experienced in this type of work. A few details of polishing procedure used by the authors would be useful to many foundry metallurgists.

C. W. BRIGGS[11]: I was very much interested in the fact that No. 9 steel of Table 1 was melted under gas and yet there was porosity in the test bar. I should like to know more concerning the reasons for this condition.

Also, I noticed from Mr. Sims' graph on fluidity that there might be a tendency for steel with aluminum added to have greater penetration into sand. However, there is also another point that is opposite to that view in that oxides of polyvalent metals, such as alumina, possess atomic valency bonds and are insoluble in the metal and form on the surface a film which has great influence on the surface tension of a metal. This film opposes wetting of refractory molds and tends to prevent the penetration of the metal into the walls of molds or cores. I should like to ask Mr. Sims in this regard if in his opinion steels with aluminum additions will show less tendency to penetrate capillary openings than a steel with no aluminum additions.

I certainly want to thank the authors for their excellent paper. It is indeed gratifying that a paper of this type has been presented to this meeting. I think it is one of the finest that has been presented before this society.

H. W. MAACK[12]: Under the heading EXPERIMENTAL PROCEDURE, the authors describe their method of temperature control, but make no

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[11] Naval Research Laboratory, Anacostia Station, Washington, D. C.
[12] Crane Company, Chicago, Ill.

mention of temperature effects in their summary. In his discussion, Mr. Lillieqvist also raises this question—the variable of temperature of the steel at the time the aluminum addition is made.

Sometimes it is very desirable to have the steel at a relatively low temperature when complicated castings with thick sections are poured. Our experience has been that we are most likely to get inferior physical properties under these conditions. We are inclined to attribute these to the lack of ability of such heats to assimilate the aluminum addition.

I would like to hear some discussion on the effect of aluminum additions to steel baths at various temperatures.

F. A. MELMOTH[13] *(Written discussion):* The question of aluminum, and its action on steel for castings, has become of almost annual prominence. It is perfectly natural that this should be so, as most of us know that the insistence upon physical tests of steel castings is constantly growing. We also know that one admits with almost a flush of guilt that one uses aluminum, as though some offense against good steelmaking morals were involved. This attitude could continue indefinitely, were it not for such organized attempts as presented in the present paper to define the mechanism of the reactions involved, and so enable us to find or evolve means of appreciating the benefits of aluminum and avoiding any incidental possible ill effect.

Many years ago I stated that aluminum was of use in the steel foundry as an adjunct to good, well-made steel, but that its use was indefensible if intended to supplant the necessity for good furnace practice. At the same time, it was reported that aluminum additions to basic electric mild steels, carefully manufactured, had little or no effect on physical properties as manifested by tensile tests.

In this country, acid electric steel manufacture is almost universal for the lighter types of mild steel castings, and I quickly found that in such steels the aluminum effect was a real source of trouble to many people, and that quite generally aluminum was classified as poisonous in its effect on steel. In practically all cases it was noticeable, on microscopic examination, that the globular normal form of non-metallic inclusions was changed as a result of the aluminum addition to a form in which they occurred as more or less broken films at the crystal junctions.

The work of the present authors, added to that of Comstock, has built up a much more complete knowledge of the mechanism of the reactions of aluminum in molten steel; but while yielding to none in my appreciation of their work there are points in my own experience which do not check completely with their explanations.

In acid electric practice, it is positively consistent with us that aluminum, or a compound material containing appreciable aluminum percentage, can be added to the large ladle during pouring from the furnace, in either small or large quantities, with little or no effect on the physical properties of bars poured at any time during the disposal of

the charge. Should, however, a further addition be made to a hand shank, an immediate reduction in ductility results. If the effect is purely on the sulphide arrangement, is it logical to believe that the time period which elapses from tapping into the ladle to actual pouring into the molds permits a reversal to the globular form? Is it not true to say, also, that an excess of aluminum is obviously present in the case of further additions to shank, and that therefore the test should be of even higher standard?

The curves shown in Fig. 3 were evolved from the results obtained by the treatment in an organized way of a series of induction furnace charges. They suggest that with sulphur content of approximately 0.03 per cent little or no influence of a harmful type is exerted by aluminum additions. Yet, in acid electric practice, the results of many times repeated tests demonstrate in my experience that a late addition of aluminum has a marked ill effect on properties, even when the sulphur is in the neighborhood of 0.03 per cent.

I quote a typical case, and our records show very many more of equal significance. Analysis: Carbon 0.24 per cent; manganese 0.67 per cent; silicon 0.34 per cent; sulphur 0.028 per cent; phosphorus 0.021 per cent. During pouring from the furnace into the large ladle, 0.6 of a pound per ton of aluminum was added in the form of alsifer. Bars were cast with no further addition (a) and others with a further 11 ounces per ton of 99 per cent pure aluminum added to a 100-lb. hand shank, before casting (b). The total, therefore, was 20.6 ounces per ton of aluminum, in the later bars cast. Heat treated together at 1650 deg., these test bars gave the following results:

	(a)	(b)
Maximum Stress (lb. per sq. in.)	75,000	73,500
Yield Point (lb. per sq. in.)	48,750	52,000
Elongation on 2 in. (per cent)	28	19.5
Reduction of Area (per cent)	46.32	26.81

The fall in ductility is very marked, and is of a magnitude sufficient to cause rejection, if specification work were involved. It is realized, of course, that the pound and a quarter per ton, while higher than is often used, is not so high as is recommended by the authors as a means of partially restoring ductility. The point I am mainly emphasizing is that this sulphur is under 0.03 per cent, and yet the effect is most marked.

While I would assure the authors that it is not my intention to throw doubt upon the conclusions they have drawn, which appear incontrovertible for the steelmaking method adopted, I would appreciate their comments or explanations, and their opinions as to whether or no process has a bearing. The expression of other members' opinions, from their own experiences of acid electric steelmaking, would also be of considerable value.

I am doubtful whether one can safely postulate the idea that what occurs in one steelmaking process is representative necessarily of others, and I am inclined to the idea, also, that steelmaking procedure, even in one process, exerts an effect on aluminum action.

In acid electric charges of the type frequently described as over-reduced (or under-oxidized) the effect is very marked, and the inclusions, even in the absence of aluminum, tend toward the deleterious arrangement. Low ductility is a characteristic.

As a matter of historical interest, it would seem that in the 1890's there is little doubt that Dr. Arnold, of Sheffield University, knew something of this type of trouble in steel castings. He referred to the micro-constituent as probably sulpho-silicide membranes, and also referred to their effect upon ductility. Is it not possible that there existed real truth in even this early attempt at definition?

WALTER CRAFTS[14] *(Written discussion)*: Messrs. Sims and Dahle have demonstrated in a remarkably clean-cut manner the effects of aluminum and sulphur on the ductility of cast steel. While sulphur has long been considered undesirable, it is believed that this is the first quantitative and wholly rational study of its harmful effect and the authors are to be highly commended for their contribution.

The tendency to No. III type inclusions has been recognized but this less undesirable type of inclusion has seldom been found to be unassociated with No. II type in our experience. The reason for this is indicated by the authors to be due to lack of uniformity and it is considered that this phase of the problem can not be unduly emphasized. The galaxies or large groups of "alumina" inclusions that tend to form in the transition from type II to type III inclusions with higher aluminum additions was not encountered by the authors to a serious degree but has been observed to a greater extent in heavy sections and can usually be controlled by making aluminum additions on the high side of the recommended range.

The observations regarding the effects of other deoxidizers on inclusions have been confirmed in our studies. However, the conclusions regarding deoxidizing power are believed to be misleading as they are based on form rather than analysis of the inclusions. Equilibrium conditions are seldom, if ever, reached in steel making, and silica is found in the presence of up to at least one per cent aluminum, although it is soluble in the sulphides and does not appear as a separate phase. In the case of calcium, however, sulphides have very little solubility in calcium silicate, so that the silicate and sulphide phases are not changed in shape or manner of occurrence, although strong deoxidation has taken place and that glassy character of the silicates has been destroyed. Similarly, vanadium has a powerful effect on the size of sulphides but does not greatly change their manner of occurrence. Addition of a small amount of aluminum to either a calcium or vanadium steel will increase the mutual solubility of sulphides and silicates and produce an amount of type II inclusions that is out of all proportion to the deoxidizing effect that such an aluminum addition alone might have.

As mentioned above, neither calcium nor vanadium produce type II inclusions when used singly. Although either, used with any other de-

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[14] Research Metallurgist, Union Carbide and Carbon Research Laboratories, Inc.

oxidizer, will intensify the formation of type II inclusions, they are unique in that when used together they form only type I inclusions. The tensile properties of a typical calcium-vanadium treated 0.30 per cent carbon, 1.50 per cent manganese normalized steel are: yield point, 77,000 lb. per sq. in.; tensile strength, 97,250 lb. per sq. in.; elongation, 26.0 per cent; and reduction of area, 52.8 per cent.

The purpose of a deoxidizing addition in our opinion is to reduce porosity by combining with, increasing the solubility of, or otherwise suppressing the evolution of gases, primarily hydrogen and nitrogen, that are not related directly to oxygen and sulphur. This is substantiated by the fact that sound castings can be obtained by means of the addition of many elements that do not greatly change the character of the inclusions. Furthermore, although a reduction in porosity is associated with the disappearance of silicates in aluminum-treated steel, this relation has not been found to be true for any other deoxidizer. For this reason it is believed that too much emphasis should not be placed on deoxidizing power when it is desired to degasify rather than deoxidize.

It is felt that while the authors have made a thorough and commendable investigation of the effect of aluminum, the intermediate degree of ductility obtained with high aluminum additions does not appear to us to be a practical method of developing high quality steel castings. Their recognition of the effect of sulphur and their excellent study of its harmful effect is an achievement of a very high order for which the authors are to be congratulated.

R. B. RAMSDELL, JR.[15]: We have had a little experience with aluminum. We make, normally, a comparatively low sulphur steel. About a year and a half ago we increased the amount of silicon from half a pound per ton to one pound per ton and we got a decided increase in ductility. Since then we have done a little experimenting with increasing the aluminum to further extent—up to one and a half pound, and did not get any marked increase.

In our practice it must be that this pound per ton gives us a way from the valley in this ductility. We find some of the type II inclusion in this steel but with the low sulphur it evidently does not have much effect on it.

We wish to thank Mr. Sims for explaining the causes of this ductility.

C. E. SIMS AND F. B. DAHLE (*Authors' written closure*): In replying, the authors wish first of all, to express a deep appreciation of the contributions that have been made in these discussions. The data given have a value which is certainly no less than that of the original paper. They constitute, in fact, the "proof of the pudding" from a utilitarian standpoint. Much of the discussion is in such close agreement with the authors' conclusions, that no comment is necessary other than to thank those who cooperated in supplying these data.

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[15] Metallurgist, American Rolling Mill Co., Middletown, O.

Mr. Caine has done a great deal of work in correlating inclusion counts with the physical properties of cast steel. He shows that the ductility and impact resistance vary inversely with the inclusion count. This agrees very well with the data of Fig. 4, because with higher sulphur contents there will be more sulphide inclusions. On the other hand, with the same sulphur content the eutectic sulphides, being smaller, are greater in number. Mr. Caine's studies apparently furnish a quantitative relation.

Mr. Caine suggests that the FeO content of the steel may have been responsible for the difference between the steels of Fig. 20. That is quite possible. The authors not being sure of the reason, mentioned temperature as a potent variable. Under equilibrium conditions temperature is often the most important factor in determining FeO content of both metal and slag.

Mr. Phillips brings up the question of reoxidation and hydrogen absorption in a metal while traversing a green sand mold. It seems a reasonable assumption that this would take place, especially if there is any turbulence or surging. The reserve of excess aluminum would then necessarily be lowered. If the oxidation is due to water vapor, hydrogen might easily be absorbed at the same time.

Hydrogen embrittlement already has a well-recognized meaning and we probably are not justified in using it to describe the type of low ductility which occurs in cast steel, and which we think is due to hydrogen. It is a type of hydrogen embrittlement but we need a distinguishing name.

Mr. Blosjo has been using as much as 2 lb. of aluminum per ton for some time and his corroboration of results is especially interesting.

We wish to take this opportunity to commend the group of foundrymen who have been meeting to discuss their common problems and pool their data. It represents an entirely modern spirit of cooperation. The uniformly excellent results reported from all five foundries in the group discussion attest to the value of such an interchange of ideas.

The comments of this group on such aspects as fluidity, cracking, etc., make a valuable addition to the data.

Mr. Comstock and Mr. Welling have pointed to the fact that heat 2669 Fig. 17 was made with low carbon ferro-titanium and that the eutectic sulphides probably are due to significant quantities of aluminum contained in this alloy. The alloy used was the low-aluminum grade affirmed by the manufacturers to contain less than 3 per cent Al. Thus the maximum possible addition of aluminum was 0.007 per cent, which is hardly enough to be significant.

The authors merely desired to illustrate that the eutectic sulphides are due to the absence of FeO, not to the presence of something. Wentrup obtained them without any alloy additions.

In reply to Mr. Lillieqvist's question as to whether all of the commercial bars were given an aging treatment after normalizing, the

answer is—No. In this he brings up a very good point, for there is little doubt that if all had been given the accelerated aging treatment there would have been fewer erratic results. In one case where a comparison was made, the reduction of area was raised from 34 per cent to 48 per cent by a draw at 700°F.

In regard to melting practice No. 7 where a high residual silicon was carried all through the heat: This was done by using clean scrap and allowing for some loss in the addition. The loss was not very large because of the short time the heat was held molten.

Mr. Lillieqvist cites some high-carbon experimental heats in which eutectic sulphides were obtained with no stronger deoxidizer than ferrosilicon. No other explanation is offered at this time, than that there was apparently a dearth of FeO.

He is quite right in his surmise that heat No. 3649, Fig. 13, would have low ductility. The elongation was 3.5 per cent, the reduction of area 4 per cent, and the impact resistance 3 ft./lb.

He asks whether a bath of steel can be killed completely with ½ lb. of Al per ton when no silicon is present. This is true, but such a steel would not be sufficiently buffered with excess of reserve aluminum for handling.

We agree with Mr. Frank that this paper does not give a panacea for low ductility. Panaceas, like the philosopher's stone, have a highly hypothetical existence.

He questions the significance of the scattering of data in 0.04 per cent sulphur range, especially as to the curves in Fig. 3. While an effort was made to have only one variable at a time, it will be noticed in Table 3 that the sulphur contents of these heats varied from 0.037 per cent to 0.046 per cent. Moreover, the heats were made on different days and some slight changes in practice crept in. The 0.06 per cent sulphur heats, on the other hand, were made consecutively and all in one afternoon. Here the variables were reduced to a minimum and all points fall on the curves.

Regardless of this, however, no claims for quantitative values are made for these curves. It is reiterated, on the other hand, that they have characteristic shapes. There is nothing critical about a sulphur content of 0.04 per cent except that it is the quantity with which most acid shops have to deal, and is, therefore, the most interesting range. Mr. Frank is referred to Fig. 4 for the effects of sulphur content.

The eutectic grouping of sulphides causes tearing in the steel which results in early fracture and this tearing can take place on the surface as well as any other part of the test bar. The brittle spots due to residual hydrogen, however, cause a similar type of tearing and these may occur regardless of the type of sulphide inclusion.

Complete deoxidation does not contemplate "removal of all oxides" as Mr. Frank states, but only the reduction of all FeO by the formation of some other oxide. Theoretically there is no such thing as complete deoxidation since all reactions have an equilibrium value. Aluminum

comes close to it, however, and with this reservation, it can be said that to get complete deoxidation the aluminum must reduce not only FeO, but also reduce all other oxides likely to be present.

The term "over-reduced" as applied to steel has never been clearly defined. It is sometimes applied to a steel in which just enough aluminum has been added to produce low ductility. It has been shown that very small additions of aluminum do not harm the ductility whereas slightly larger additions may lower it. Thus the conclusion that the latter is "over-reduced." Some of the properties attributed to "over-reduction" are due to insufficient boil in working the heat and can be blamed on "under-oxidation" rather than "over-reduction."

On the other hand, it is not believed possible to "over-oxidize" steel to its detriment. But if conditions are such as to give a higher FeO content to the steel, it would, naturally, require a larger aluminum addition to leave an unoxidized residue.

As to whether silicon is necessary or advantageous when aluminum is used, no complete answer is available at this time. Because of its quantity silicon may serve as a local buffer against oxidizing influences during pouring. On the other hand, steels with very excellent physical properties have been made without silicon and with only manganese and aluminum as deoxidizers.

Mr. Crafts has given some very interesting conclusions in regard to the effect of various deoxidizers on inclusion structure. The authors were particularly interested in the statement that silicates are present in steels deoxidized with Al even when as much as one per cent is used. He states that the silica is soluble in the sulphides and does not appear as a separate phase. The authors would be interested in knowing how the presence of silica was shown under such conditions.

In regard to the occurrence of the No. II type of inclusions attention is again called to the evidence presented that they are caused by the absence of FeO. It is quite true, as Mr. Crafts states, that the primary purpose of a deoxidizing addition is to reduce porosity but there seems to be some difference of opinion as to the mechanism by which a deoxidizer prevents porosity. Mr. Crafts evidently believes it is due to a direct combination with the gas-forming element. There is some evidence to support this contention, as, for example, the formation of metal nitrides. We believe, however, that hydrogen is the principal offender in producing porosity and that the prevention is directly due to deoxidation. Evolution of hydrogen apparently is initiated in many cases by a small bubble of water vapor and the formation of the latter requires the presence of some FeO.

In reply to Mr. Melmoth it should first be said that there was no intention to infer that sulphur contents below 0.03 per cent guarantee freedom from low ductility. The curves of Fig. 4 show a nearly straight-line relation between sulphur content and ductility. At 0.02 per cent S the capacity for damage is very slight, but let it be said again that the data of Figs. 3 and 4 are quantitative only for the particular conditions under which they were obtained.

The two samples of steel to which Mr. Melmoth refers were examined and are reported as P-1 and P-2 in Table 4. The A sample (P-2) had very characteristic type I inclusions showing that the aluminum had been all oxidized. The B or shank samples (P-1) had mixed II and III type of inclusions with the type II predominating. All things considered these steels are found to be characteristic rather than exceptional.

Dr. Chipman's reference to the great difference in the deoxidizing powers of silicon and aluminum is most interesting. He has made notable contributions to our knowledge of deoxidizers which have also served to emphasize the need for further detailed study.

The authors emphatically agree with Mr. Urban that aluminum should be regarded as a legitimate and useful alloying agent for steel and not as a "dope." His suggestion about the possibility of determining the difference in solubility of non-metallic inclusions, especially sulphides, by indirect methods is certainly worthy of further consideration.

Mr. Ramsdell is in the enviable position of being able to produce acid electric steel with a sulphur content that hovers around 0.02 per cent. This would naturally minimize his troubles with low ductility and it is possible that one lb. of aluminum per ton is enough for his conditions.

Mr. Armstrong brings up the question of whether, in view of the explanation for recovered ductility advanced by the authors, the widely used practice of melting under oxidizing conditions has any value other than to reduce the solubility of gases. Also, whether good ductility in cast steel can be assured simply by adding 0.07 per cent or more of aluminum.

These are pertinent questions but cannot be given an unqualified answer. The removal of gases is an important function of the "boil" and in this manner it more or less directly influences the ductility. That an oxidation operation is not essential to good ductility, however, is shown by melting practice No. 7 in Table 1. Mr. Blosjo reports in his discussion that he carries a fairly high residual silicon, and reduces all his silicon from the slag and still gets good ductility. It is true, of course, that if the temperature is high enough severe oxidation and a moderate residual silicon may be coincidental.

Mr. Grotts reports that a consistently good ductility was not obtained until a consistent melting practice was adopted and this checks with the reports of other foundries. The practice used includes a vigorous boil. It is believed, therefore, that the correct answer to these questions is that the steel melting practices previously found to be best are necessary for optimum results when 0.07 per cent or more aluminum is used.

At Mr. Armstrong's request a brief description of the method used for polishing specimens for micro-examination has been written and attached as APPENDIX A. This method is recommended because of its triple merit in being simple, rapid and especially satisfactory for inclusions.

The porosity obtained in the No. 6 practice (Table 1) and referred to by Mr. Briggs, is considered to be due to absorption of hydrogen produced by decomposition of the natural gas that was used to create a reducing atmosphere. Absorption of hydrogen is apparently much faster after deoxidation.

It is somewhat difficult to draw conclusions regarding penetration from test bar castings, but it was observed that the aluminum deoxidized steels conformed closely to the contour of the mold surface even to the individual sand grains. This would indicate a tendency to penetrate, but no burned on sand was encountered.

Mr. Maack's questions about the effect of the temperature of the steel at the time the aluminum addition is made brings up an interesting and important point. It was mentioned earlier in the discussion that temperature is considered a potent variable and a number of attempts were made in the experimental work to establish a definite relation between temperature of the steel at the time of the additions and the physical properties obtained. This work was not reported because the results were not sufficiently clear cut to make a good case, although the evidence favors the belief that higher temperature tends to give better ductility.

In the presentation of this paper some data on the effect of aluminum on the "flowability" was presented. Inasmuch as these data were not in the preprint they are presented in APPENDIX B.

Appendix A

METHOD OF POLISHING STEEL SPECIMENS FOR OBSERVATION OF INCLUSIONS

The method used for polishing the specimens for microscopic examination was developed by G. A. Lillieqvist of the American Steel Foundries, and has been used since 1927. In this method maintainence of equipment is simple, polishing is rapid, and a high quality finish is obtained.

The ideal size specimen is about $\frac{1}{4} \times \frac{5}{8} \times \frac{3}{4}$ in. polished on a $\frac{5}{8} \times \frac{3}{4}$ in. face. After rough preparation the specimens are ground dry with paper clamped to a rotating disc. The dry paper is Behr-Manning, with grit No. 1 or 2, No. 0 and No. 000 used in succession, rotating the specimen 90° between papers. The No. 000 scratches should run lengthwise.

Two wet wheels are used in succession. The cloth used is a duvetyn type such as "selvyte." New cloth is used on the second wheel and when somewhat worn moved up to the first. The polishing medium is levigated alumina.

In preparing the alumina for use 60 grams of commercial levigated alumina is shaken thoroughly in 2 to 2½ liters of freshly distilled water. It is then allowed to stand for from 8 to 15 hours and then about one liter decanted off. The first suspension is then diluted to two liters and used for the first wet polish while the decanted material is used for the final polish.

The first wheel is kept fairly wet with the alumina suspension and the specimen is polished at 90° to the No. 000 scratches, occasionally changing the position 180°. With the wheel running counter clockwise at about 650 r.p.m. the specimen is rotated in a clockwise direction using fairly heavy pressure. The No. 000 scratches should be removed in about one minute, or at most two minutes. If polished too long relief polishing results.

On the second wheel only a small quantity of the suspension, uniformly distributed, is used. The specimen is wiped clean and polished at 90° to the first wheel and very slight pressure is used. Polishing is continued until the specimen is practically dry. The critical part of the procedure is to remove the specimen at just the right time. If removed too wet it will stain, and if polished too long relief and burnishing will result. The proper time can be told in several ways. A slightly greater pull from the cloth will be felt and if the specimen is quickly turned over a thin film of moisture will be observed to disappear in one to two seconds. The specimen is taken directly from the wheel to the microscope or to a desiccator.

Trouble will sometimes be encountered from coarse grit in the No. 000 paper, and some levigated alumina is deficient in fine material.

Appendix B

FLOWABILITY TESTS ON MEDIUM CARBON CAST STEEL

Tests on the flowability of aluminum killed medium carbon cast steels were completed after the paper had been preprinted. The data showing the effect of aluminum additions on the flowability of the steel are presented here so that more complete information would be available.

Fig. 1 of this appendix shows the type of test mold used in the tests. This mold was only recently developed but seems to work satisfactorily and it is possible to obtain check results with fair consistency. Data on the tests are shown in Table 1 and Fig. 2 of this appendix. The tests were made in a high frequency induction furnace using an optical pyrometer to read temperatures. In pouring it is important that some metal be poured off first to heat the lip of the crucible. The first metal to come over the lip is chilled somewhat. It is also important to pour rapidly in order to keep some metal in the pouring cup until the

Table 1
FLOWABILITY DATA ON MEDIUM CARBON CAST STEEL

Heat No.	Aluminum Added Per cent	Flowability—Inches		
		2850°F.	2900°F.	2950°F.
4021	....	5½	9	12
4028	0.025	6	8½	11¼
4029	0.05	6	10	13¾
4032	0.075	6½	9¾	12¾
4033	0.15	6¾	9½	12¼
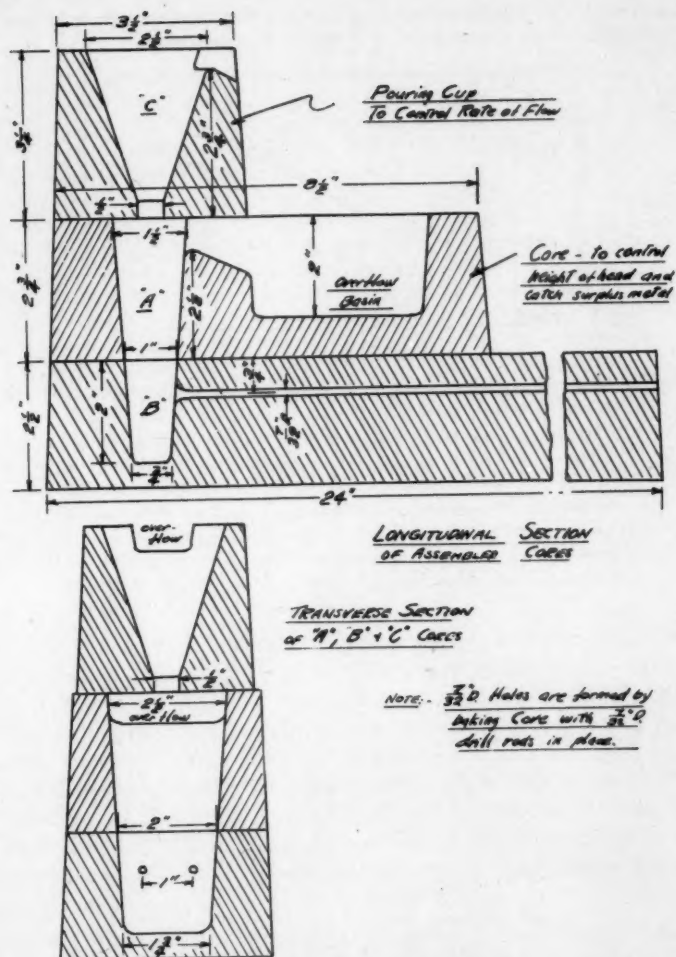
FIG. 1—DETAILS OF FLOWABILITY TEST MOLD.

mold is filled. The lower overflow automatically regulates the bead of metal if pouring is fast enough.

Aluminum in amounts up to 5 pounds per ton or 0.25 per cent has not changed the flowability of the medium carbon cast steel to any marked extent. The data merely illustrates, however, that large amounts of aluminum are detrimental to flowability in cast steel.
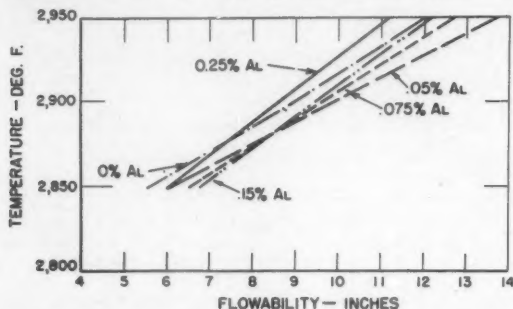
FIG. 2—EFFECT OF ALUMINUM ADDITIONS ON FLOWABILITY OF GRADE "B" CAST STEEL

The Effect of Deoxidation Treatments on the Ductility of Cast Steels

By Albert P. Gagnebin*, Bayonne, N. Y.

Abstract

The ductility of heat treated cast steel is directly related to the distribution of the sulphide inclusions. Steels which have chain-like sulphides, arranged in a network around the primary grains, have low ductility because stress concentrations cause failure through the sulphide links, whereas steels with globular sulphides randomly distributed have high ductility. The arrangement of the sulphides depends largely on the oxygen content of the melt, a high oxygen content causing random distribution and a low oxygen content resulting in the network formation. A rather critical amount of oxygen is required for sound, ductile steel and it is difficult to control this important variable; porosity may develop because of insufficient deoxidants, or low ductility may result because of an insufficient oxygen content promoted either by reducing conditions during melting or by the use of a strong deoxidizer, such as aluminum. A melting practice has been developed involving oxidation of the melt, then deoxidation with silico-maganese, calcium and aluminum, which thoroughly kills the steel, produces random sulphide distribution and consistently results in steels having high ductility, high impact resistance and high yield ratios.

1. Wide variation in the ductility of cast steels, similar in composition and heat treatment, is common knowledge among those familiar with cast steel. This phenomenon naturally has claimed the attention of foundrymen interested in producing cast steel meeting definite ductility requirements and has been discussed in the literature at various times. Sims and Lillieqvist,[1]** MacCrae, Jordan, et al.,[2] and Comstock[3] are among the authors discussing the low ductility of moderately hard, properly heat-treated steels which are sometimes referred to as "overreduced" steels.

2. This study is largely the outgrowth of the consideration of this problem and shows that its cause is the root of considerable

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evil for cast steel. The direct relation between the iron oxide content of cast steel at the time of pouring and its ductility, already pointed out by Sims and Lillieqvist, is a fundamental factor involving the effect of different melting processes, various deoxidation procedures and finally porosity. All these variables are discussed, and suggest that the consistent production of sound, ductile cast steel with present methods is rather difficult, since steels are quite likely to contain porosity or to have low ductility. An improved method of deoxidation is proposed which shows good promise of consistently producing absolutely sound cast steel with maximum ductility.

THE RELATION BETWEEN SULPHIDE INCLUSION DISTRIBUTION, IRON OXIDE CONTENT, AND DUCTILITY IN CAST STEEL

3.   An early attempt to produce a few high frequency induction furnace heats of pearlitic cast steels similar in composition, cast into well fed one-in. sections, and normalized and drawn together, yielded rather surprising differences in ductility, from 25 to 55 per cent reduction in area. There was nothing unusual in the production of these heats. Armco iron was simply melted, carburized with wash metal and deoxidized with silico-manganese. A critical examination of the melting record indicated that the

Table 1

EFFECT OF INCREASED OXYGEN ON DUCTILITY OF CAST STEEL

Composition: C., 0.30 to 0.35 per cent; Mn., 1.25 to 1.50 per cent; Ni., 0.50 per cent; Si., 0.40 per cent; S., 0.035 per cent—Normalized and Drawn.

Special Treatment	Yield Point lb. per sq. in.	Tensile Strength, lb. per sq. in.	Elong., Per Cent	Reduction of Area, Per Cent
None	56,000	88,600	19.5	24.1
Atmospheric Oxidation				
Held 2½ min.	58,650	88,650	24.0	36.9
Held 6 min.	55,600	87,900	27.0	44.0
Effect of Lower Silicon				
0.12 per cent Si.	46,350	85,850	27.0	40.3
0.25 per cent Si.	52,200	86,450	26.5	44.9
Effect of Added Oxygen				
0.06 per cent O₂ as NiO................	50,800	85,000	24.0	32.8
0.10 per cent O₂ as NiO................	49,100	82,200	25.5	41.0
0.05 per cent O₂ as Fe₂O₃............	50,800	82,500	20.5	28.9
0.10 per cent O₂ as Fe₂O₃............	54,700	89,600	22.5	33.5
0.11 per cent O₂ as Fe₂O₃............	52,500	86,100	25.0	38.8
0.15 per cent O₂ as Fe₂O₃............	49,700	83,500	29.0	56.3
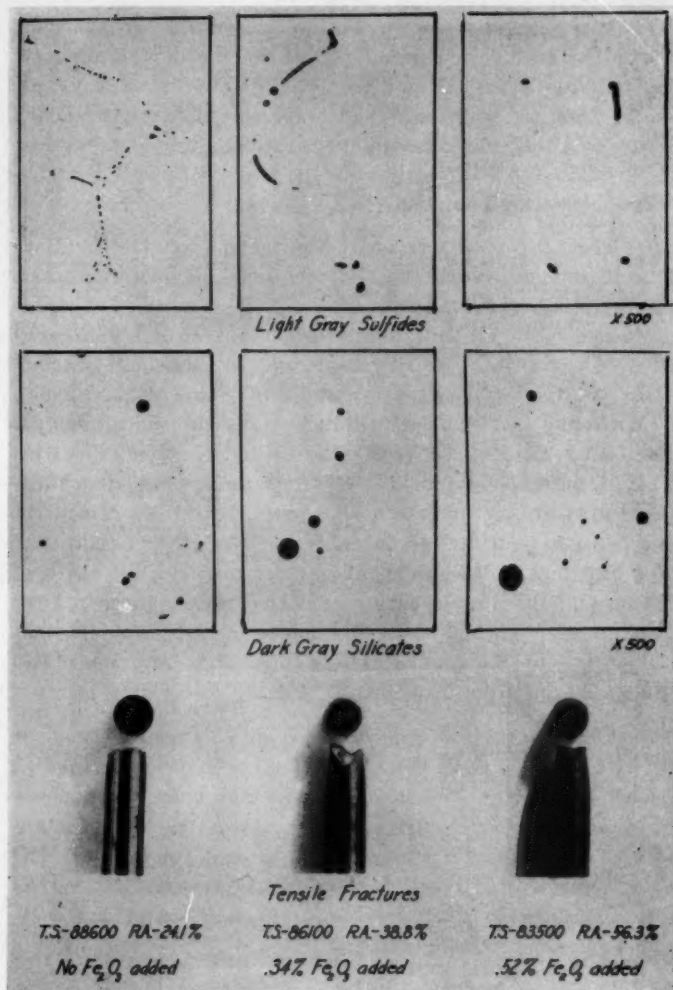
FIG. 1—EFFECT OF IRON OXIDE CONTENT ON THE DUCTILITY AND INCLUSION DISTRIBUTION OF CAST STEEL—COMPOSITION: C., 0.30 PER CENT; NI., 0.5 PER CENT; MN., 1.5 PER CENT; SI., 0.85 PER CENT—NORMALIZED AND DRAWN.

melts with better ductility had been subject to greater oxidation during the melting cycle.

4. The relation between the oxide content of the melt and ductility was demonstrated readily in a group of melts in which the iron oxide content was increased in three ways: by oxidation from the atmosphere, by lower silicon contents, and by additions of nickel and iron oxide. The tensile properties of these heats, given in Table 1, clearly show the improved ductility resulting from increasing oxidation.

5. The excellent paper on inclusions in cast steel by Sims and Lillieqvist[1] shows the effect of inclusions on ductility, which was substantiated by an examination of these steels. Two types of inclusions were evident, light gray sulphides and dark gray silicates. The sulphides varied considerably in shape and distribution. In low ductility steels, they were thin, elongated, and chain-like, encircling the primary grains, while in well oxidized, high-ductility melts, they were globular and randomly distributed. With intermediate ductility, the sulphides were partially agglomerated and of an intermediate size, but still outlined the primary grain boundaries. Silicates, on the other hand, were always globular and well dispersed regardless of the state of oxidation and ductility; heavier oxidation simply increased the size and number of inclusions. Examples of the sulphides, silicates and tensile fractures of steels representing low, intermediate and high ductility are shown in Fig. 1.

6. The designations oxide content and oxygen content, as used in this paper, refer to the amount of ferric oxide applied in contact with the molten surface of the high frequency furnace melts. Consequently, greater or less oxidation means the addition of larger or smaller amounts of iron oxide to the melts. It does not necessarily follow that the oxygen content of the resulting metal, as determined analytically, bears a direct and proportional relation to this quantity.

7. The identification of inclusions was accomplished by the series of etching reagents recommended by Comstock and Campbell.[4] The inclusions reacted sharply to the various reagents making their identification rather simple. The final step in the procedure, exposing the identity of sulphides in the first specimen

of Fig. 1, is shown in Fig. 2. Partial attack and blackening by chromic acid (not shown) and more severe attack by sodium picrate indicates the presence of sulphides, presumably in the form of complex iron-manganese-oxy-sulphides. Silicates were identified by the fact that they were hardly affected by any of the reagents except hydrofluoric acid.

Mechanism of Failure

8. A steel containing intergranular sulphides can be recognized during the tensile test. After the yield point is exceeded, a number of checks appear on the surface of the tensile bar which gradually open to form small cracks; after very little deformation, fracture starts in one of these, progresses slowly across the specimen, and final rupture takes place with a weak, dull thud. With well dispersed sulphides and high ductility, no surface checks appear, and after necking down a goodly amount, rupture occurs with a sharp crack which indicates that most of the section has failed simultaneously.

9. The intergranular sulphides are directly responsible for the surface checks and low ductility. A transverse section through a check (Fig. 3) shows its origin directly in a string of
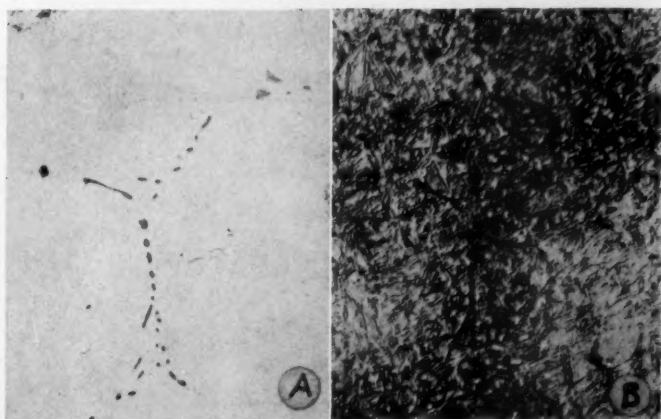
Fig. 2—Identification of Sulphide Inclusions in Steel 9717—(A) Unetched, 500x; (B) Etched, Ten sec. in 10 per cent Nitric Acid, 5 min. in 10 per cent Chromic Acid, then 10 min. in Boiling Sodium Picrate, 500x.
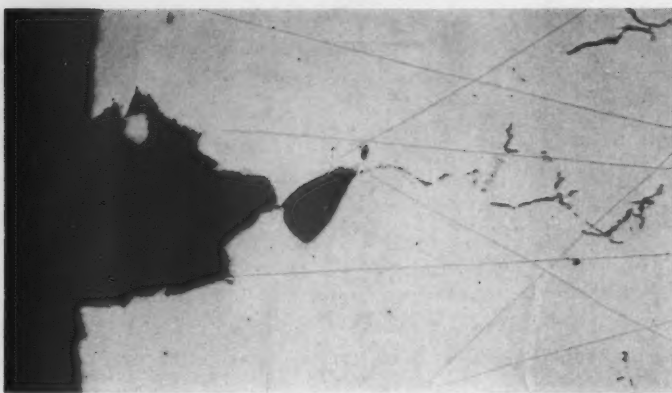
FIG. 3—CROSS SECTION THROUGH A SURFACE CHECK SHOWING ORIGIN IN CHAIN TYPE SUL-
PHIDE INCLUSION, 250X.
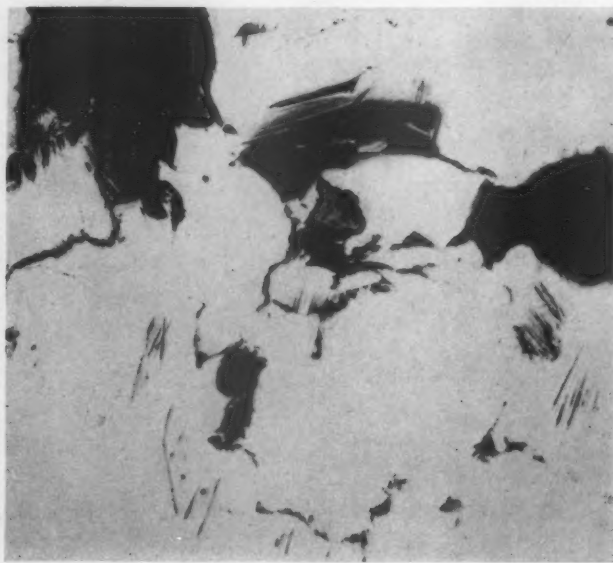
FIG. 4—CROSS SECTION OF FRACTURED TENSILE BAR SHOWING INTERNAL VOIDS AND CHAIN-
TYPE SULPHIDE INCLUSIONS, 500X.

sulphide inclusions. Fig. 4 shows local internal ruptures origi-
nating in intergranular sulphides in a zone of the cross section
near the center of the bar. Apparently, the chain-like sulphides
prevent a uniform distribution of strain, causing local cracks
which induce fracture with very little deformation.

FIG. 5—INTERGRANULAR SULPHIDES IN ACID ELECTRIC STEEL—S., 0.032 PER CENT—NORMAL-
IZED AND DRAWN—TENSILE STRENGTH, 84,000 LB. PER SQ. IN.; REDUCTION OF AREA,
22.5 PER CENT, 200X.
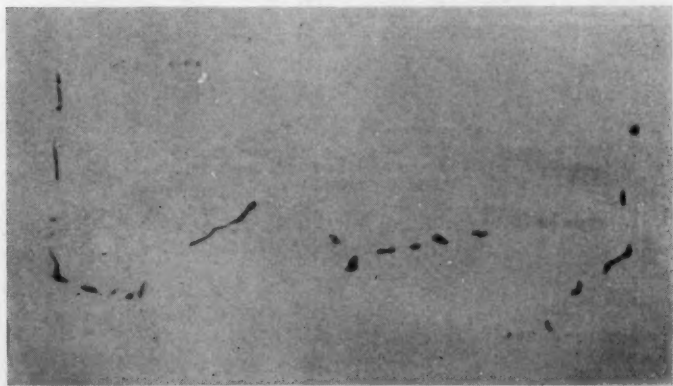
FIG. 6—INTERGRANULAR SULPHIDES IN BASIC ELECTRIC STEEL—S., 0.024 PER CENT—NOR-
MALIZED AND DRAWN—TENSILE STRENGTH, 95,000 LB. PER SQ. IN.; REDUCTION OF AREA,
37.0 PER CENT—500X.

10. It may seem surprising to find sulphides encircling grains in steels with normal sulphur content and more than 1.25 per cent manganese, since this is the time honored sulphur fixing element. Manganese seems to be a poor desulphurizer in the absence of sufficient iron oxide, since it forms sulphides which are rejected near the end of the iron solidification and are caught and strung out in the grain boundaries. With proper oxidation, manganese plays its expected role; sulphides are rejected from solution early and thus have sufficient opportunity to coalesce and disperse themselves in a random manner.

INTERGRANULAR SULPHIDES AND POROSITY IN STEELS FROM VARIOUS COMMERCIAL PROCESSES; ALUMINUM DEOXIDATION

11. Samples of regularly produced commercial heats of cast steel, received at Bayonne over a period of years, were examined for sulphide inclusion distribution. The ductile heats have invariably had random sulphide distribution while the brittle heats
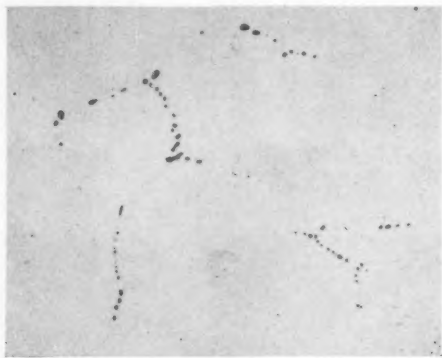
FIG. 7—INTERGRANULAR SULPHIDES IN BASIC OPEN HEARTH STEEL—S., 0.025 PER CENT— NORMALIZED AND DRAWN—TENSILE STRENGTH, 93,000 LB. PER. SQ. IN.; REDUCTION OF AREA, 37.0 PER CENT—500X.

have contained intergranular sulphides. The sulphide distribution of three low ductility heats from different foundries are represented in Figs. 5, 6 and 7 showing that intergranular sulphides can occur with any commercial process and with any normal sulphur content, even as low as 0.025 per cent.

12. With normal operation, these processes differ in the degree of oxidation developed and, consequently, should have greater or less tendency to produce non-ductile heats. MacCrae,[2]

et al. and Sims and Lillieqvist[1] point out that steels which will
have poor ductility are sluggish, tend to stick in nozzles, and
have poor fluidity. Melmoth[5] has said that "the following would
represent the order of steel making processes based on their com-
parative ability to produce steel with a degree of fluidity high
in proportion to its temperature: (1) Converter, (2) electric
furnace (Single Slag), (3) Siemens open-hearth, (4) electric
furnace (Double Slag)." This order of decreasing fluidity is
also, with normal operation, the order of decreasing oxidation of
these practices. The old converter process is the most highly
oxidizing, blowing air directly into or on the surface of the melt.
The acid electric is probably more oxidizing than the basic open
hearth, although both provide heavy oxidation. The basic elec-
tric is different, since it utilizes two slags, the second of which
is quite reducing and consequently produces steel low in oxygen
content. This last fact is substantiated by the experience of one
foundry which failed to meet high ductility specification with
several basic electric heats, and then turned to acid steel and
easily met the requirements.

Porosity

13. While it is desirable, from a standpoint of ductility, to
utilize a process which produces a high degree of oxidation, with
ordinary production, extreme care has to be taken to avoid po-
rosity, a condition that can occur with any process because of
variable conditions. The criticism often heard of cast steel is
that it is likely to contain porosity; in fact, exacting service
sometimes requires a radiograph of every casting. The desire
for assured soundness, a fundamental requisite for any material,
has led to the use of strong deoxidizers, such as aluminum, espe-
cially in converter and acid electric steel.

Aluminum Deoxidation

14. In quantities of 0.03 to 0.08 per cent, aluminum is an
excellent deoxidizer which assures soundness in cast steel, but
may seriously lower ductility unless it is used under controlled
conditions. A few data in Table 2, quoted from the literature,
confirm the low ductility normally induced by this element. In-
clusion examinations of aluminum finished steels with low duc-
tility, revealed intergranular sulphides (Fig. 8), which is to be
expected, since aluminum so effectively reduces the iron oxide

content. Another indication of low oxide content is that aluminum killed steels contain very few, small silicates. Its normal use is precluded in alloy steels with high ductility requirements. However, it was necessary to study its behavior in more detail since occasional reports claim its use with no impairment of ductility.

Table 2
EFFECT OF ALUMINUM ON DUCTILITY OF CAST STEEL

Source	C. %	Mn. %	Si. %	Ni. %	Heat Treatment	Ladle Treatment	Yield Point lb. per sq. in.	Tensile Strength, lb. per sq. in.	Elong. Per Cent	Reduction of Area Per Cent
MacCrae[2]	0.10 to 0.25	0.70 to 0.73	0.33 to 0.40	......	Normalized	None	45,000	71,000	34.0	53.0
						1 lb. Al. per ton	46,000	71,000	27.5	26.5
Sims and Lillieqvist[3]	0.28	0.79	0.37	......	Normalized	None	47,700	79,200	31.5	56.0
	0.28	0.79	0.37	......	Normalized	0.05 per cent Al.	50,450	80,000	25.0	35.7
See Note*	0.28	1.37	0.48	1.20	Normalized	None	60,500	101,000	26.5	52.5
	0.28	1.37	0.48	1.20	Normalized and Drawn	0.03 per cent Al.	63,000	102,000	20.2	32.8

* Private information from a commercial foundry.

FIG. 8—INTERGRANULAR SULPHIDES IN ALUMINUM KILLED ACID ELECTRIC STEEL—NORMAL S. —NORMALIZED AND DRAWN—TENSILE STRENGTH, 101,000 LB. PER. SQ. IN.; REDUCTION OF AREA, 23.0 PER CENT—500X.

15. The first question considered was the length of time required for the sulphides to readjust themselves to the intergranular type if aluminum were added to well oxidized steel. A few melts in Table 3, finished with 0.06 per cent aluminum but with different time intervals between aluminum addition and pouring, indicate that ductility is impaired even when the melt is poured immediately, and becomes steadily worse as the time interval increases to 5 minutes.

16. Fig. 9 records the ductility of heats finished with various amounts of aluminum and held 2 to 3 min. before pouring. Notice that ductility drops sharply with small amounts, of 0.02 to 0.03 per cent, then recovers somewhat with larger amounts but is never high enough for quality ductility specifications. Small amounts of aluminum do not completely refine the grain but sometimes promote a segregate structure and fine grains in the vicinity of the primary grain boundaries encircling coarser grains. This condition apparently causes greater loss in ductility. With larger amounts of 0.05-0.25 per cent, the grains are uniform in size.

17. There are but three conditions where aluminum does not have a deleterious effect on ductility. The first is when small amounts are added to steels very high in oxygen so that the aluminum is completely oxidized and sufficient iron oxide remains after its addition to maintain random, globular sulphides. Such steels have been produced commercially with 0.025 per cent aluminum. This method has little to make it attractive because control of iron oxide is still required, since the same possibilities of porosity and intergranular sulphides exist as in regular manganese and silicon deoxidation. The second method of aluminum addition with a minimum effect is its introduction immediately before pouring, where ductility will be lowered somewhat but will sometimes meet high ductility requirements. This method is very difficult to perform and might easily result in large variations in properties, even in adjacent sections. The third method is the use of aluminum in steels virtually free from sulphur. One commercial heat, tested at Bayonne, containing 0.06 per cent aluminum and 0.01 per cent sulphur had very high ductility.

Silicon Deoxidation

18. Silicon is another strong deoxidizer which causes low ductility if the iron oxide content is reduced sufficiently, but duc-

Table 3

EFFECT OF TIME INTERVAL BETWEEN ALUMINUM ADDITION AND POURING ON DUCTILITY OF CAST STEEL

Composition: C., 0.30 per cent; Ni., 1.50 per cent; Mn., 1.50 per cent; Si., 0.40 per cent; S., 0.035 per cent—Normalized and Drawn 1600 to 1200°F.—1 hr. each. A. C.

| Mark	Melt Treatment	Yield Point lb. per sq. in.	Tensile Strength, lb. per sq. in.	Elong., Per cent	Reduction of Area, Per cent
P 11.11	Properly Oxidized	63,500	93,000	27.0	54.0
T-10305	0.06 per cent Al. Poured immediately	70,500	97,000	25.0	45.4
T-10367	0.06 per cent Al. Poured immediately	67,700	93,000	26.0	45.7
T-10314	0.06 per cent Al. Held 2 min.	67,700	93,200	21.0	36.0
T-10304	0.06 per cent Al. Held 4 to 5 min.	70,700	98,700	21.5	33.1
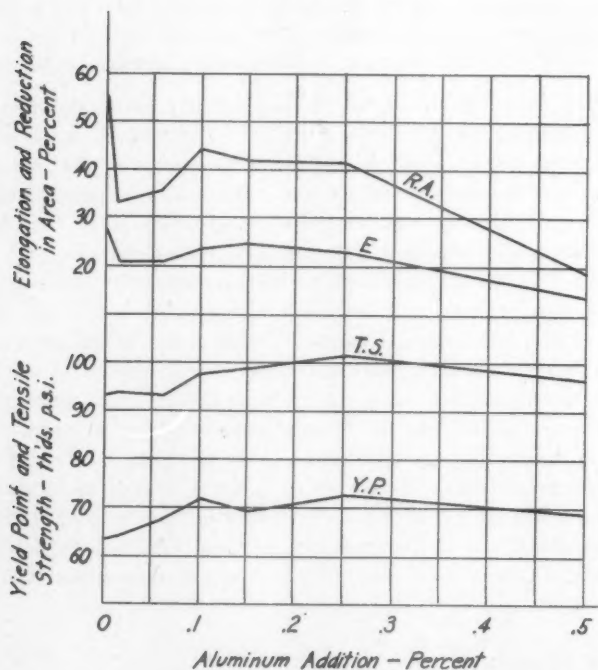
FIG. 9—EFFECT OF ALUMINUM ON THE MECHANICAL PROPERTIES OF NICKEL-MANGANESE CAST STEEL—NORMALIZED AND DRAWN—C., 0.30 PER CENT; NI., 1.5 PER CENT; MN., 1.5 PER CENT, SI., 0.40 PER CENT.

tility can be restored by heavy oxidation as indicated in Table 4. Here, again, the situation is precarious as success depends on the proper amount of iron oxide with no assurance of sound, ductile material.

Hence, these strong deoxidizers offer doubtful aid to the foundryman since their use requires perhaps even better iron oxide control than is required for regular deoxidation.

The Development of an Improved Method of Deoxidation

Calcium, Zirconium and Titanium in Silico-Manganese Deoxidized Steels

19. Intergranular sulphides apparently developed because manganese became impotent as a desulphurizer in the absence of sufficient iron oxide. Hence, the simplest method of overcoming this situation seemed to be in searching for some element with greater affinity for sulphur which would maintain random distribution regardless of the oxide content. Elements reported to have strong affinity for sulphur are titanium, zirconium, calcium, lithium, magnesium and beryllium; all were used except magnesium and lithium which were highly explosive even in a dilute form.

Table 4

Ductility of High Silicon Steels With and Without Oxidation Normalized and Drawn

	Composition				Yield Point, lb. per	Tensile Strength, lb. per	Elong., Per	Reduction of Area, Per
Mark	C. %	Ni. %	Mn. %	Si. %	Melt Treatment	sq. in.	sq. in.	Cent	Cent
T-9072	0.20	1.34	0.98	1.10	None; regular melt	65,400	87,500	21.0	24.8
T-11192	0.20	1.5	0.86	1.14	Heavily oxidized	54,500	85,500	32.5	58.9

Table 5

Effect of Titanium, Zirconium and Calcium on the Ductility of Cast Steel

Composition: C., 0.30 per cent; Ni., 1.50 per cent; Mn., 1.50 per cent— Normalized and Drawn.

Mark	Melt Treatment	Yield Point, lb. per sq. in.	Tensile Strength, lb. per sq. in.	Elong., Per Cent	Reduction of Area, Per Cent
T-10292	0.1% Ti (as FeTi [40%])	67,500	98,000	20.5	41.3
T-10291	0.1% Zr (as SiZr [40%])	64,400	97,500	22.5	49.2
T-10557	0.1% Ca (as SiCa [25%])	57,350	86,800	29.2	59.1
T-10307	0.1% Ca (as SiCa [25%])	67,000	101,000	23.5	48.9

Because these alleged sulphide forming elements also have a strong affinity for oxygen, it was necessary to know whether this affinity predominated over that for sulphur. Consequently, 0.1 per cent of calcium, zirconium and titanium was added to separate, well oxidized nickel-manganese melts and held a few minutes before pouring. The ductility of these steels, Table 5, indicate that calcium and zirconium maintained random sulphide dispersion while lowering the oxygen content; titanium, on the other hand, apparently had a preferred affinity for oxygen thereby promoting intergranular sulphides. Microscopic examination checked these observations.

Calcium-Aluminum Treated Steels

20. Calcium and aluminun were added to steels to determine whether calcium could maintain random sulphide distribution in spite of the drastic reduction in oxide content promoted by aluminum addition. A series of high frequency induction furnace melts, made in the regular manner with silico-manganese and properly oxidized, were finished with simultaneous additions of 0.06 per cent aluminum and various amounts of calcium. The melts were held several minutes after the calcium-aluminum additions to allow the sulphur time to dispose itself as it would. The mechanical properties in Fig. 10 show that the melts with more than 0.05 per cent calcium are excellent, having good ductility, high yield ratios, exceptional toughness, all in combination with very low oxygen contents.

21. Apparently, if calcium is added to well oxidized melts, the free oxide content can then be lowered without disturbing the sulphide distribution. However, if it were added to melts low in oxygen which would have poorly distributed sulphides without further treatment, could it retrieve and disperse this type of sulphide? To assist in answering this question three melts were made which achieved low oxide contents in three ways, (1) by adding aluminum first with a several minute interval before the calcium addition, (2) by melting under a lime-fluorspar slag excluding as much air as possible, (3) by adding the deoxidants as soon as the charge was molten to take advantage of lower oxygen solubility. The low ductility of all these melts, Table 6, testifies to the presence of intergranular sulphides and indicates that calcium is unable to disperse sulphides in a melt which already has predisposition to the intergranular condition.

22. It is strongly emphasized that calcium must be added at

Table 6

EFFECT OF CALCIUM ON DUCTILITY
WHEN ADDED TO LOW OXIDE CONTENT MELTS

Composition: C., 0.30 per cent; Ni., 1.50 per cent; Mn., 1.50 per cent; Si., 0.35 per cent.

Mark	Melt Treatment	Yield Point, lb. per sq. in.	Tensile Strength, lb. per sq. in.	Elong., Per Cent	Reduction of Area, Per Cent
T-10559	Melted under a lime-flourspar slag, the 0.1 per cent Ca., 0.06 per cent Al. added together 62,000		91,700	22.0	35.4
T-10401	C., Mn., Si., 0.06 per cent Al., 0.1 per cent Ca., added at a low temperature, poured at the usual temperature 71,500		98,800	20.0	35.1
T-10405	0.06 per cent Al. added, held 3 min. before the 0.1 per cent Ca. addition 62,000		88,000	24.0	41.0
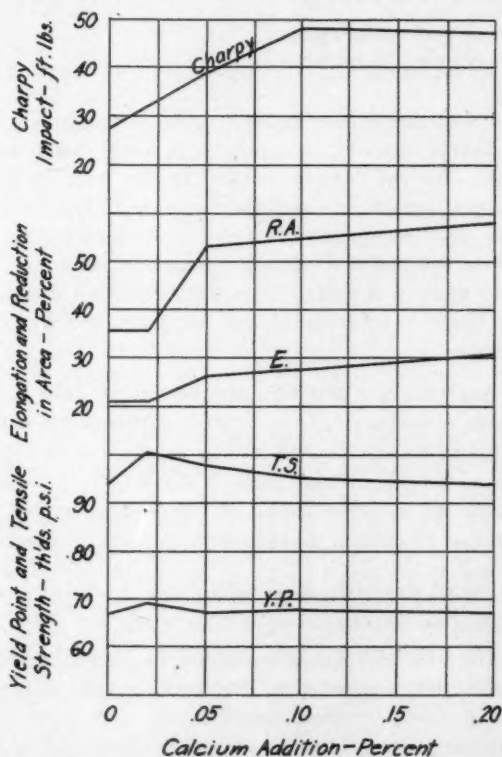
FIG. 10—EFFECT OF CALCIUM ON THE MECHANICAL PROPERTIES OF ALUMINUM FINISHED CAST STEEL — NORMALIZED AND DRAWN — C., 0.30 PER CENT; NI., 1.5 PER CENT; SI., 0.40 PER CENT; MN., 1.5 PER CENT; AL., 0.06 PER CENT.

a time when the sulphides are in an oxidized state predisposed for
random distribution; the aluminum and calcium may be added
simultaneously or the calcium may be added before the aluminum.
Another precaution is that calcium must be forced into the melt;
if added to the surface, it simply burns away with little entering
the melt, even though added as an alloy of silicon as it was in all

Table 7
EFFECT OF OXIDATION AND DEOXIDATION ON ANALYSIS

Mark	Melt Treatment	C. %	Mn. %	Si. %	S. %
T-10794	Regular Si + Mn., ½ of melt poured as such	0.31	1.45	0.35	0.039
T-10794	Heavily oxidized with 0.57 per cent Fe₂O₃	0.27	1.32	0.30	0.038
T-10563	Si.-Mn. with 0.1 per cent Ca				0.037
T-10564	Si.-Mn. with 0.06 per cent Al. 0.1 per cent Ca.				0.036

these tests. Calcium-silicon has been added readily to acid electric
and open-hearth melts by wrapping it in heavy paper bags which
were tossed into the tapped stream during pouring and imme-
diately carried beneath the surface of the melt. For induction fur-
nace heats, small screen containers were attached to iron rods and
plunged into the melts. One more precaution in connection with
this type of melting is that at high temperatures (above 3000°F.),
which are likely to be achieved, the calcium-silicon has tendencies
to be explosive.

23. Analyses of a few significant heats in Table 7 show that
the oxidation procedure and deoxidation with calcium has no effect
on the sulphur content; the treatment achieves its benefits by sul-
phide dispersion. Moreover, melt T-10794, which was oxidized
about double the amount necessary for good properties, shows a
very small loss in carbon, manganese and silicon contents.

Zirconium, Titanium, Beryllium, Molybdenum and Tantalum in Low Oxygen Content Melts

24. The ability of these elements to maintain random sul-
phide distribution was tested by adding them just before the alu-
minum, to properly oxidized heats, which were then held a few
minutes before pouring. Table 8 shows that beryllium is very good,
0.02 per cent being equivalent to about 0.1 per cent calcium. Zir-
conium is also useful but not quite as effective as calcium in the

Table 8

EFFECT OF BERYLLIUM, ZIRCONIUM, TITANIUM, TANTALUM AND
MOLYBDENUM WITH 0.06 PERCENT AL. ON THE DUCTILITY
OF CAST STEEL

Composition: C., 0.30 per cent; Mn., 1.50 per cent; Ni., 1.50 per cent;
Si., 0.40 per cent—Normalized and Drawn.

Mark	Melt Treatment	Yield Point, lb. per sq. in.	Tensile Strength, lb. per sq. in.	Elong., Per Cent	Reduction of Area, Per Cent
T-10319	0.02 per cent Be. 0.06 per cent Al.	67,000	98,300	26.7	56.3
T-10404	0.10 per cent Zr. 0.06 per cent Al.	65,000	91,300	27.0	51.6
T-10556	Heavily Oxidized 0.10 per cent Zr. 0.06 per cent Al.	64,000	90,500	28.0	52.0
T-10403	0.10 per cent Ti. 0.06 per cent Al.	67,300	96,600	20.7	35.7
T-10555	Heavily Oxidized 0.10 per cent Ti. 0.06 per cent Al.	74,100	98,400	20.0	28.2
T-10341	0.20 per cent Ta 0.06 per cent Al.	73,000	100,700	21.0	36.0
T-10320	0.10 per cent Mo. 0.06 per cent Al.	68,600	101,500	18.5	32.2

Table 9

COMPARISON OF VARIOUS COMPOSITIONS FINISHED REGULARLY
AND WITH CALCIUM-ALUMINUM NORMALIZED AND DRAWN

Type	Nominal Composition % C.	% Mn.	% Si.	% Ni.	% Mo	Method of Finishing	Yield Point, lb. per sq. in.	Tensile Strength lb. per sq. in.	Yield Ratio, Per cent	Elong. Per cent	Reduction of Area, Per cent	Impact†
2 per ct. Ni.	0.19	0.65	0.35	2.0	.....	Ca.-Al.*	57,000	76,000	75.1	34.0	60.4	50C
						Regular	45,000 to 55,000	75,000 to 85,000	60.0 to 64.6	32.0 to 25.0	65.0 to 45.0	80-40Z
Ni.-Mo.	0.30	0.75	0.35	1.7	0.25	Ca.-Al.	69,300	88,900	78.2	26.0	52.5	41C
						Regular**	60,500	91,000	66.7	24.3	54.5	
Mn.	0.30	1.5	0.35	.....	.....	Ca.-Al.	59,300	84,900	70.0	30.0	57.6	48C
						Regular	55,000	86,000	64.0	30.0	60.0	30C
Medium C.	0.30	0.75	0.35	.....	.....	Ca.-Al.	47,600	70,700	67.2	32.5	46.0	29C
						Regular	43,000	74,000	58.1	33.0	50.0	24.5C

* 0.1 per cent Ca., 0.06 per cent Al.
** Average of eight heats.
† C = Charpy Test; Z = Izod Test.

same amount. The other elements seemed to have no merit at all, since the ductility achieved was comparable to that expected with aluminum alone.

25. Hence, three elements have been found useful in fixing sulphides—beryllium, calcium and zirconium. An examination of the inclusions of aluminum-finished steels containing these elements showed a very few, small silicates. Examples of the sulphides are given in Figs. 11 to 13. The sulphides of the aluminum-beryllium and aluminum-calcium are randomly distributed, and tend toward angularity, suggesting that precipitation occurred early with an attempt on the part of the inclusions to assume crystallographic shapes. The sulphides of the aluminum-zirconium steel are not as angular as those promoted by calcium and beryllium; moreover they vaguely outline primary grain boundaries, a location which is substantiated by the slightly lowered ductility.

Comparison of Aluminum-Calcium, Calcium, and Regular Cast Steel

26. One natural question in regard to the calcium-aluminum method deoxidation might be, why not use calcium alone since it is in itself a strong deoxidizer? The mechanical properties of three nickel-manganese cast steels in Fig. 14, finished regularly, with calcium alone, and with aluminum-calcium, normalized at 1500, 1600 and 1700°F., then drawn at 1200°F., assist in answering this question. The aluminum-containing steel has nearly double the impact resistance of the other two and a better yield ratio because of its fine grain size; furthermore, aluminum prevents grain growth at the higher temperatures, thus maintaining the high toughness, while the other two steels sustain heavy losses in impact resistance because of a coarsened grain. This virtual insensitivity to normalizing temperature may be of commercial advantage.

Calcium-Aluminum Deoxidation in Various Compositions

27. Most of the deoxidation study was confined to nickel-manganese cast steel simply because this was a familiar composition on which there existed a good background of properties. The aluminum-calcium method of deoxidation was tried on various classes of cast steel to show that it was readily applicable to other types of steel. The properties of induction furnace melts made with this method are compared, in Table 9, with generally accepted properties of similar analyses made regularly with only silico-manganese deoxidation.

28.   Aluminum-calcium deoxidation has improved the yield ratios of these steels, and the toughness, where it is directly comparable, without changing the strength or ductility. The most important difference is that the aluminum-calcium steels have very low susceptibility to porosity while the regular silico-manganese killed steels have rather high susceptibility.
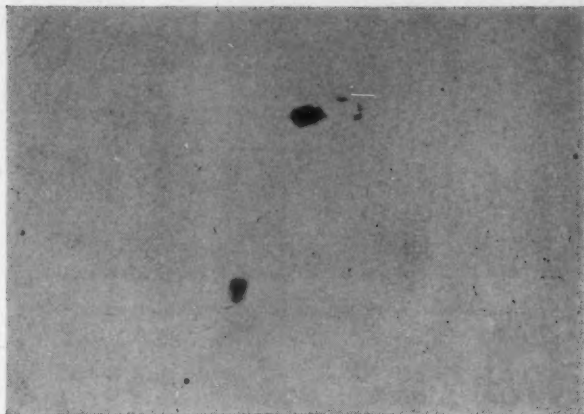
FIG. 11—(LEFT) SULPHIDES IN ALUMINUM-BERYLLIUM STEEL—500x.
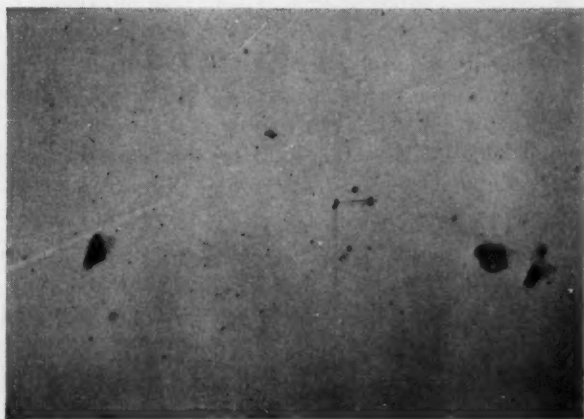
FIG. 12—(RIGHT) SULPHIDES IN ALUMINUM-CALCIUM STEEL—500x.
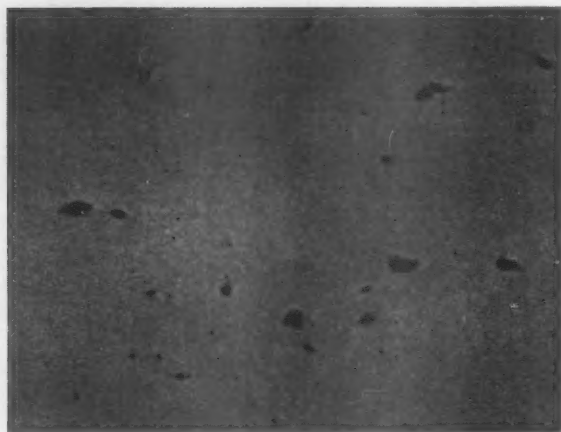
FIG. 18—SULPHIDES IN ALUMINUM-ZIRCONIUM STEEL—500X.

29.   The proposed method of deoxidation allows better control of the variables of cast steel production. With ordinary deoxidation procedures, a rather critical amount of iron oxide is necessary because an excess results in porosity and a deficiency causes low ductility. With the proposed method, steel is made in the regular manner, oxidized, and then finished with calcium and aluminum. The oxidation step is not critical as the oxide content can vary between wide limits above the minimum requirement since the relatively large amount of deoxidizers precludes the possibility of porosity. The introduction of a strong sulphide forming element, such as calcium, (beryllium, or zirconium), in the presence of sufficient oxide, concludes the influence of oxygen on the sulphide form and distribution. The oxide content can then be drastically lowered by a strong deoxidizer such as aluminum and random sulphide dispersion will be maintained. Consequently, it should be readily possible to consistently produce steels of assured soundness and maximum ductility.
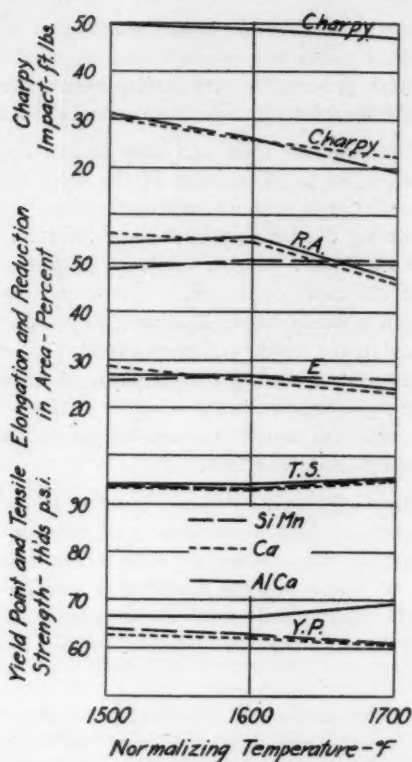
FIG. 14—EFFECT OF NORMALIZING TEMPERATURE ON THE MECHANICAL PROPERTIES OF STEELS FINISHED WITH SILICON-MANGANESE, ALUMINUM-CALCIUM, AND CALCIUM ALONE—DRAWN 1 HR. AT 1200°F.—C., 0.30 PER CENT; NI., 1.5 PER CENT; MN., 1.5 PER CENT; SI., 0.40 PER CENT.

SUMMARY

30. The data given in this paper may be summarized as follows:

(1) Wide variability in the ductility of cast steels, similar in composition and heat treatment, occurs because of differences in sulphide inclusion distribution. The sulphides vary between two extremes, a thin, elongated intergranular type encircling primary grains causing low ductility, and a globular, randomly-distributed type allowing maximum ductility. Between these two extremes were partially agglomer-

ated sulphides outlining the grain boundaries and resulting in intermediate ductility. Silicate inclusions, the other predominant type present in cast steels, were always globular and randomly distributed with no apparent effect on ductility.

(2) The sulphide form and distribution apparently depends on the iron oxide content of the melt at the time of pouring. Low oxygen contents, achieved either by melting conditions or strong deoxidizers, such as aluminum, promote intergranular sulphides. Higher contents permit some agglomeration and sufficient oxide content results in globular sulphides randomly distributed. Manganese appears to be a poor desulphurizer in the absence of proper oxide content, sulphides are precipitated late during solidification of the steel, do not agglomerate rapidly and are thus caught in the primary grain boundaries. Silicates simply became larger and more profuse with increasing oxide contents.

(3) Since the oxide content is affected by the charge, slag, temperature, and deoxidants and because there is no rapid method of determining its concentration, the consistent production of sound, ductile material is problematical. Cast steel is quite likely to be deficient in oxygen and thus have poor ductility, or to contain excessive oxygen and so develop porosity.

(4) A new method of deoxidation is proposed, combining very low oxide contents with good ductility. It consists of producing steel in the regular manner, oxidizing it, then finishing with calcium (0.10-0.20 per cent) and aluminum (0.05-0.10 per cent). The addition of a strong sulphide forming element such as calcium, in the presence of a sufficient oxide content, maintains random sulphide dispersion in spite of the drastic reduction in oxide content promoted by a strong deoxidizer such as aluminum. Two other strong sulphide forming elements were found, beryllium which is as good as calcium but rather expensive, and zirconium which is not quite as effective as calcium.

(5) Beside insuring soundness with maximum ductility, the method has several other marked advantages. It produces fine grained steels which have high ductility, toughness, and yield ratios and allows wide latitude in normalizing temperature without grain corsening.

ACKNOWLEDGEMENTS

31. The author is indebted to Prativu Guha-Thakurta for much of the experimental work, to Norman B. Pilling for encouragement and permission to publish the results, and particularly to Dr. John T. Eash for helpful suggestions and criticisms, and cooperation during the entire investigation.

BIBLIOGRAPHY

1—Sims, C. E., and Lillieqvist, G. A., *"Inclusions—Their Effect, Solubility and Control in Cast Steel,"* A.I.M.E., 1932, vol. 100, p. 154.

2—McCrae, J. V., Dowdell, R. L., Jordan, Louis, *"A Study of the So-Called Overreduced Condition in Cast Steels,"* BUREAU OF STANDARDS, 1930, vol. 5, p. 1123.

3—Comstock, G. F., *"Aluminum and Titanium as Deoxidizers,"* IRON AGE, Dec. 4, 1924, p. 1477.

4—Campbell and Comstock (Modified by Wohrman), *"Method for Identification of Inclusions in Iron and Steel,"* NATIONAL METALS HANDBOOK, 1936, p. 561.

5—Melmoth, F. A., *"Variables in Steel Foundry Practice,"* TRANS. A.F.A. (1928), vol. 36, p. 323.

DISCUSSION

C. E. SIMS[1] *(written discussion)*: Mr. Gagnebin obviously started the investigation on which he has reported with the same general objective in mind as the authors of the contemporary paper on the "Effect of Aluminum on the Properties of Medium Carbon Cast Steel," namely, of obtaining the advantages of strong deoxidizers like aluminum without suffering concurrent disadvantages of low ductility. Many of his data parallel and corroborate those of the paper just mentioned. The principal points of difference seem to be in the interpretation of the results, and that always allows room for argument.

Most of the work of the present paper was done on medium manganese-nickel steels which are sufficiently different from plain carbon steels that the results should not be too freely extrapolated to apply to the latter. For instance the melting and deoxidation practice described in paragraph 3 should always give good ductility when used with medium carbon cast steel.

I would appreciate having the author explain Fig. 1 more fully. It is clear that the three fracture tests are from steels shown in Table 1 but it is not definitely indicated whether the photomicrographs above the fracture tests represent the same steels. It would be unexpected to find sulphides like those in the top row of pictures in Fig. 1, in steels like those shown in Table 1, and it would be most unusual to find sulphides and silicates like those of the first vertical row occurring in the same steel.

[1] Supervising Metallurgist, Battelle Memorial Institute, Columbus, O.

In paragraph 10 the author concludes that manganese is a poor desulphurizer in the absence of iron oxide. There is reliable evidence to show that manganese is always a strong sulphide former but that MnS varies in solubility in molten steel with variation in FeO content. In steels devoid of FeO, MnS has high solubility and precipitates as a eutectic.

In speaking of porosity (paragraph 13) it is said that extreme care is necessary to prevent porosity after a steel has been highly oxidized. This would presume the evolution of CO, and I believe it can be shown that any silicon content over 0.25 per cent is ample assurance against CO evolution in any steel, however highly oxidized. A high degree of oxidation, on the other hand, if accompanied by a vigorous boil will tend to remove dissolved hydrogen and nitrogen and thus help to prevent subsequent porosity.

The data of Fig. 9 corroborate the data in Fig. 3 of the paper by Sims and Dahle. In the latter paper the aluminum additions were not carried above 0.25 per cent.

In Table 3 the aluminum addition of 0.06 per cent is just about the minimum quantity necessary to give recovery of ductility and is apt to give erratic results. Thus in Table 3 when the steel was poured immediately the steel had a fair ductility but when it was held long enough for some of the residual aluminum to be oxidized, as in heats T-10314 and T-10304, the ductility dropped. Oxidation occurs very rapidly in a small induction furnace.

In discussing the effects of calcium, zirconium and titanium (paragraph 19) the author starts with an incorrect premise in assuming that intergranular or eutectic sulphides form because manganese is impotent as a desulphurizer in the absence of sufficient iron oxide. It is apparently true, however, that the solubility of MnS can be lowered, not only by FeO, but also by the presence of small quantities of other sulphides such as Al_2S_3. Naturally it is only that portion of a metal that remains unoxidized after the equilibrium with oxygen is established that can form sulphides.

In Table 8 it is shown that the combination Be-Al and Zr-Al gave good ductility whereas combinations of Al with Ti, Ta and Mo gave much lower ductility. It was also shown in Fig. 9 that 0.10 per cent Al gave better ductility than 0.06 per cent Al. Both Be and Zr are strong deoxidizers and both 0.02 per cent Be plus 0.06 per cent Al and 0.10 per cent Zr plus 0.06 per cent Al are equivalent in combining power to 0.10 per cent Al. Is it not possible, therefore, that these combinations acted just as would larger additions of aluminum?

The Ti is a weaker deoxidizer and Ta and Mo are not deoxidizers at all. These elements form very stable carbides, however, which might explain their different effect.

The foregoing explanation is supported by the photomicrographs in Fig. 11 of an Al-Be steel, Fig. 12 of an Al-Ca steel and Fig. 13 of an Al-Zr steel. The sulphides in these steels are very typically those

obtained with recovered ductility when an excess of Al is used to leave some residual Al.

The action of the combination Ca-Al might be explained in a similar manner, except that it seems to have an effect beyond that of the equivalent amount of Al used alone. Calcium is known to form a very stable sulphide which is, moreover, quite insoluble in steel. Calcium alone does not seem to be a strong deoxidizer but in combination with aluminum it exhibits some interesting and useful properties. These data constitute a decidedly worth while contribution to our knowledge of cast steel.

C. H. JUNGE[2] AND W. K. BOCK[2] *(written discussion):* This interesting paper evidently offers another method of fixing the sulfides of an aluminum treated steel so that they will be distributed in a harmless form. In Sims' and Dahle's paper presented previously at this meeting, it was shown that a large excess of aluminum could also be used to fix the sulfides in order to accomplish a similar result.

The data for this paper were obtained on small induction-melted heats. The commercial operator is obviously greatly interested to know if the process described has been tried on commercial open hearth and electric furnace heats, and what results were obtained.

The author said that poor ductility was due to the shape and the distribution of the sulfides. Is it impossible or improbable to have poor ductibility in a heat having good sulfides?

The statement that desirable sulfides come out of solution at high temperature leads to a question bearing on the mechanism of the process. Does the calcium react with the materials which form the sulfide to produce a high melting complex or one which is immiscible in the liquid steel?

It was shown that the detrimental effect of aluminum increased with time. Are there any data to show that holding a heat for a time after the calcium addition does not tend to produce bad sulfides also?

From the discussion of the paper by Sims and Dahle it appeared that it was the experience of many that steels containing less than 0.03 per cent sulphur were not particularly susceptible to impairment by the use of aluminum, yet in the presentation of this paper we have just seen examples on the screen of several steels of this sulphur content made by different processes which show very bad intergranular sulfides, and the accompanying poor ductility values. Sims' results were for Grade B steel, with normal manganese. Could the higher manganese of this intermediate-manganese type of steel give greater trouble with intergranular sulfides than the Grade B steel, the sulphur being held constant?

It seems curious that calcium, a powerful deoxidizer, is able alone to accomplish the fixing of the sulfides, while if added after the

aluminum, also a powerful deoxidizer, it is unable to fix the sulfides. Does that indicate that if a portion of the calcium is added, then the remainder later, that the sulfides will not be fixed?

Another item of interest is what happens in the author's process when a large excess of aluminum is added after the calcium addition. Since the sulfides obtained in the operation of the author's process are angular, while those obtained by Sims' are larger and of duplex nature, it should be rather easy to determine which type of sulfide fixation occurs in the presence of an excess of aluminum.

In order to use the author's process successfully, it appears to be necessary to oxidize the heat to a point where globular sulfides would be obtained if the heat were then cast, and follow this with the calcium-aluminum addition. In the method suggested by Sims, such preliminary conditioning of the bath is unnecessary, and therefore this process appears to be more easily controllable in commercial practice. An expression from the author as to the relative merits of the two processes would be very helpful to the commercial operator.

H. D. PHILLIPS[3] *(written discussion)*: It is a privilege to be able to discuss this paper, containing as it does, much worth while information. Its study is well repaid. It is personally gratifying to note that the author found even in the induction furnace process, a real necessity for heavy oxidation during the making of his heats. We have, in our effort to secure uniformly ductile and sound steel castings by the acid electric process, been criticized for over-oxidation of our melts. The results secured, and those of the author, however, prove the worth of this practice.

The term "over-reduced" as applied to steels was coined by Batty, I believe. In his studies of these steels he observed that they all showed poor fluid "life" but not necessarily poor ductility. As against this, he found that an addition of 0.10 per cent Al to a normal steel did not cause poor fluid "life," but did give poor ductility in some acid electric steels.

Figs. 3 and 4 are excellent and show very nicely the cause of failure in bars having a net work of sulfides around the primary grains. They also show why certain heats of steel have a greater tendency to hot-tearing in the mold, as a result of hindered contraction stresses, than do others, for while steel may not go through a mushy stage during solidification, this eutectic at the grain boundaries is quite likely to be still liquid at 2,350°F., at which approximate temperature Briggs and Gezelius have found that most hot-tearing occurs in steel castings.

I would differ with the author's statement that the acid electric process is probably more oxidizing than the basic open-hearth. The acid electric practice quite commonly followed several years ago, and still being followed in some foundries today, could hardly be called strongly oxidizing. In fact, melts were made repeatedly in which the residual silicon content did not go below 0.15 per cent. The FeO content of the

metal in equilibrium with this silicon is not much over 0.03 per cent at 2900°F. The fallacy of this practice has been well demonstrated, however.

It is interesting to note the reference to one foundry being unable to meet high ductility specifications with basic electric steel. It is my experience that properly made basic electric steels, with aluminum contents up to 0.15 per cent, and more, show excellent physical properties. In fact, in impact strengths they surpass properly made acid electric steels, unless carefully selected low sulphur scrap is used in the acid furnace and controlled Al additions are made to the melts.

In connection with the author's statement that for assured soundness in cast steel, it is necessary to use a strong deoxidizer such as Al, especially in converter and acid electric steels, I would say that Al is added to practically all steels made by any process when they are to be cast in green sand molds at a high temperature and the resultant castings have to pass a rigid surface examination. The amount used and its effect may vary, but it is an element common to all melting and pouring floors where small and medium size steel castings are made in green sand.

While the author has very definitely demonstrated his ability to get good results by the advocated calcium-aluminum treatment, we are securing results quite as good by an addition of 0.10 per cent to 0.15 per cent Al alone. It is not my intention, however, to discredit the use of Ca-Si in conjunction with high Al additions, for I know of several commercial investigations showing that it has some merit, but at the same time, the physical properties and inclusion types found in the Ca-Al treated steels can be duplicated in the straight Al treated steels.

The results shown in Fig. 9 do not agree with those we have secured by varying Al additions. With no Al, the results are quite good, while with approximately 0.05 per cent Al they are very poor, but as the Al is increased to 0.10 per cent and even up to 0.20 per cent, the maximum amount tried, we secure from 90 to 100 per cent of our normal properties, except in the case of impact strengths and yield ratios, which are greatly improved.

Table 3, showing the effect of time interval between Al addition and pouring on the ductility of cast steel, can be questioned. The results shown by the author differ with those reported verbally by John Howe Hall, who found that as the time interval increased, the physical properties improved. They also differ from our carefully observed results on steels treated with 0.125 per cent Al. We have found absolutely no difference in bars poured early in the casting of an 8,000 lb. heat and those cast late, the elapsed time interval sometimes being 50 minutes. We have also found in the treating of small quantities of metal in hand ladles with Al, a wide difference in results in bars poured early, but bars poured after an elapsed time of several minutes have been consistently quite good. The explanation for this lies in the amount of Al added, time, mixing and exposure to oxidizing conditions.

Al is a low melting point element, quite soluble in molten steel.

When it is added to a ladleful of metal, it goes into solution quickly, and unless good mixing is assured, certain parts of the metal may contain an excess of the element, be well deoxidized and ductile, because some Al sulfide is formed during solidification, while other parts, robbed of their full amount, may be deoxidized thoroughly and evidence a network sulfide structure and poor ductility. This has been demonstrated in commercial practice. It is not my thought that this happened to the author. However, the 0.06 per cent added to his heats was possibly not sufficient to ensure the required excess after exposure to oxidizing conditions for 4 or 5 minutes. Sims and Dahle, in their experiments with induction furnace melts, secured good ductility with additions of 0.05 to 0.075 per cent Al, but in commercial practice, with its less efficient manner of introduction and increased liability to oxidation, an addition of 0.10 to 0.15 per cent is necessary to duplicate their results.

As a result of this research by the author, and the work done by Sims and Dahle, it now is shown that Al, long considered a necessary evil by steel foundrymen, to be used only in preventing porosity in green sand castings, is now capable of conferring the desirable properties of improved grain growth inhibition, increased impact strengths and higher yield ratios, when used in controlled amounts.

Fig. 14 is a very interesting demonstration of the beneficial effects of Al.

In conclusion, I wish to congratulate and thank the author for this excellent paper.

WERNER FINSTER[4] *(Written discussion):* Based on Messrs. Sims' and Dahle's investigation on the effect of high aluminum additions we made fifteen heats to which from 2.5 to 3.5 lb. aluminum per ton were added to the ladle. On three of these heats one pound calcium silicide per ton was added to the ladle prior to the addition of three pounds aluminum per ton.

After normalizing, test bars from the heats treated with three pounds Al per ton had 80 per cent of the ductility normally obtained on similar steel treated with six to eight ounces Al per ton, whereas the bars from the heats treated with one pound calcium silicide per ton plus three pounds Al per ton had 88 per cent of the ductility normally obtained.

After a normalize and draw treatment the ductility of the heats treated with three pounds Al per ton was 90 per cent of normal ductility against 100 per cent for the heats treated with one pound calcium silicide per ton plus three pounds aluminum per ton.

A. B. GAGNEBIN *(Written closure):* In reply to Mr. Sim's question concerning the relation of the photomicrographs to the fractures in Fig. 1, the author wishes to state that the micrographs represent typical inclusions of the tensile specimens whose fractures are shown. The three steels are from a group of thirty heats and were examined

together endeavoring to show the few small silicates and intergranular sulfides that occur with low oxide contents and also how increasing oxidation increases the size and number of silicates and agglomerates, the sulfides.

The evidence of this investigation is not in agreement with Mr. Sim's conception of the mechanism of sulfide dispersion promoted by Zr and Be in conjunction with aluminum. He suggests that Zr and Be, by greatly reducing the iron oxide content, permit the existence of some unoxidized aluminum which therefore forms sulfides. If this were true the sulfides formed should resemble those of a steel finished with 0.1 per cent Al, such as shown by Sims and Dahle which were of a duplex nature containing a black constituent. The sulfides of Al-Zr, and Al-Be steels were light gray in color, containing a single constituent just as sulfides in Al-Ca steels. Moreover, such a theory presumes a strong affinity of aluminum for sulfur and neglects any affinity which Zr or Be have for sulfur.

Ca, Zr and Be probably form stable sulfides which are able to maintain a random distribution in spite of the drastic iron oxide reduction caused by aluminum. Nickel-manganese steels finished with one of these elements and aluminum have higher ductility than can be achieved with aluminum alone.

The term "desulfurizer" in the sense used in the paper means an element which renders the sulfides least harmful, viz., globular and randomly distributed. While manganese is of course a strong sulfide former it does not always remove the sulfur from a harmful location. With low oxide contents resulting in manganese sulfides strung out around the grain boundaries, a location which would be assumed by iron sulfide if no manganese were present, it cannot be said that manganese is effective as a desulfurizer.

Messrs. Junge and Bock bring up an interesting question when speaking of the change in the mechanism of sulfide rejection resulting from a calcium addition. The fact that the sulfides are angular suggests considerable mobility during formation which may result because of a high temperature of precipitation.

The question relating to the ineffectiveness of calcium if added after the aluminum cannot be answered at present. The calcium can be added in several portions but must precede the aluminum addition in order to be effective.

In regard to the causes of low ductility in cast steel, the author hastens to say that there are several. Sulfide distribution is a very important factor influencing ductility and in this study an effort was made to reduce all other factors to a minimum.

Poor distribution of sulfides in intermediate manganese steel containing 0.03 per cent sulfur causes greater loss in ductility than would occur in Grade B steel probably because the former is a harder material and not because it contains more manganese. With harder

material and greater forces required for plastic deformation, less discontinuity can be tolerated at the grain boundaries.

The author agrees heartily with Mr. Phillips that heavy oxidation is an essential part of the operation of producing good quality steel. It is a point which has received far too little attention.

In connection with the oxidation characteristics of the various steel melting processes, much depends, of course, on the particulars of operation. It is true that many acid electric furnace heats are not as well oxidized as some open hearth heats, but in general, we think, aluminum is more often added to acid electric heats than to basic open hearth heats which might be taken as an indication of greater susceptibility to porosity in the acid electric heats.

Mr. Finster's results are quite interesting although by increasing the amount of calcium to 0.07-0.1 per cent it should be possible to obtain 100 per cent of the normal ductility in both the normalized and normalized and drawn conditions.

Refractories for Foundry Ladles

BY C. E. BALES* AND A. R. BLACKBURN**, IRONTON, O.

Abstract

*This paper describes the use of special fire brick shapes
for pouring ladle service. Pouring ladle brick prevent
slag particles from getting into the molds; and the use
of ladle bottom tile overcomes "boiling," gives a jointless
bottom, and a more economical means of installation than
fire brick bottoms. Plastic ramming mixes also are
described for use in desulphurizing ladles.*

1. Progressive foundrymen are interested in refractories, for
they realize that these heat resistant materials represent an im-
portant part of their production costs. If better refractories could
be obtained or if the existing refractories could be used to better
advantage, the life of the linings would be greater, the equipment
could be used longer, maintenance costs would be reduced, and the
profits of the foundrymen would be increased.

2. During the past several years, discussions on refractories
and refractory service conditions in foundry furnaces have been
given an important place on the programs of this association [1, 2, 3].
More emphasis has been placed on cupola refractories because nor-
mally about 4 per cent of the total quantity of fire clay brick pro-
duced in the United States is used in cupolas[4]. Bung brick and side
wall brick for malleable iron air furnaces[5, 6] also have been dis-
cussed, and some attention has been directed to electric furnace
refractories[7].

4. The purpose of this paper is to discuss some of the newer
developments in refractories for ladle use as these refractories are
just as important in the efficient operation of a foundry as cupola
blocks and air furnace brick.

* Vice President, The Ironton Fire Brick Co., Ironton, Ohio.
** Ceramic Engineer, The Ironton Fire Brick Co., Ironton, Ohio.
[1]Schaller, G. S., *"A Study of Cupola Lining Refractories,"* TRANS. A.F.A., vol. 35, pp. 266-274 (1927).
[2]Bales, C. E., *"Refractories for the Cupola,"* TRANS. A.F.A., vol. 36, pp. 683-696 (1928).
[3]Marbaker, E. E., *"Cupola Refractories,"* TRANS. A.F.A., vol. 41, pp. 491-494 (1933).
[4]Sullivan, J. D., *"Notes on Some Foundry Refractories,"* TRANS. A.F.A., vol. 44, pp. 254-280 (1936).
[5]Bales, C. E., *"Manufacture and Properties of Refractories for Air Furnaces,"* TRANS. A.F.A., vol. 30, pp. 360-380 (1922).
[6]Hewitt, L. C., *"Malleable Furnace Refractories,"* TRANS. A.F.A., vol. 33, pp. 51-54 (1935).
[7]Hewitt, L. C., *"Notes on Refractories for Electric Furnaces Producing Special Irons,"* TRANS. A.F.A., vol. 42, pp. 577-583 (1934).
NOTE: This paper was presented at a session on Refractories held at the 1938 Convention of A.F.A., Cleveland, Ohio, May 16, 1938.

LADLE POURING BRICK

5. When molten metal is tapped from the cupola or air furnace, it is practically impossible to make a perfect separation of the slag and metal. Very frequently, small bits of slag and other foreign materials will float on the iron and unless extreme care is used, these particles will be poured into the molds, causing the loss of many castings. This is the cause of many of the so-called inclusions in castings, which, when found after a job has been half finished in a machine shop, represents a considerable financial loss.

6. Ladle pouring brick, shown in Fig. 1, are fire clay shapes in the form of half cylinders of varying diameters and lengths, made for different sized ladles. The use of these brick enables the foundryman to secure the benefits of bottom pouring from either hand, bull, or crane ladles. They are very simple to use. When the mud lining has been placed in the ladle, the brick is
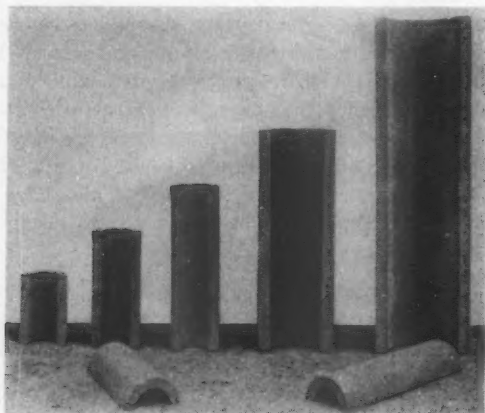
FIG. 1—LADLE POURING BRICK.

pressed into the mud to a depth of about one quarter inch, and about 2-in. above the bottom, as shown in Fig. 2. The lining is then dried thoroughly in the usual manner, and the ladle is ready for use. In pouring, the clean iron passes up through the brick, and the slag remains in the ladle.

7. These brick are most successfully used in hand and bull ladles, but some fire brick manufacturers make them for ladles as large as those up to 2-ton capacities. The writers, however, feel

that a brick for a 3000 lb. ladle is as large as can be used satisfactorily. They are not suitable for steel foundry ladles as the metal nearly always will freeze in the spout. Nor will they resist the high temperatures and iron oxide attack encountered in steel practice.

LADLE BOTTOM TILE

8. Ladle bottom tile, shown in Fig. 3, are circular disks of various diameters made to fit all sized ladles, from small hand to 2-ton ladles. They usually are made from highest grade fire clays to which is added a large percentage of calcined flint clay, to make them resistant to temperature shock without cracking.

9. When ordinary ladle mud is used to line a bottom, it is practically impossible to dry it thoroughly, because the fire clay used in the mud contains about 10 per cent chemically combined water, as well as certain gas forming minerals which are not driven out until heated to approximately 1800°F. These conditions account
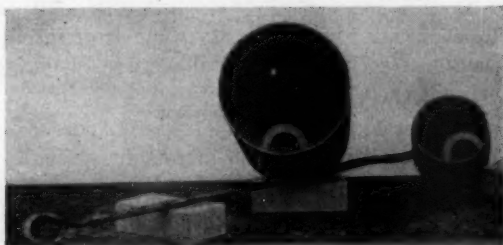
FIG. 2—POURING BRICK INSTALLED IN HAND AND BULL LADLES.

FIG. 3—LADLE BOTTOM TILE.

for the disagreeable "boiling" seen in many shops when hot iron is poured into a green ladle. On the subject of "boiling," it was recently brought out in a discussion at the Michigan State College Regional Foundry Conference of the A. F. A. that this "boiling" from wet bottoms is one of the principal causes of inverse chills in castings. Furthermore, it is well known that a dry bottom gives cleaner castings.

10. In lining ladle bottoms with 9-in. fire brick or splits, considerable chipping and fitting is necessary, and even then, the bottom is full of joints. If one has had any experience with the molten metal working through these joints and melting a hole in the ladle bottom, he will agree that while it is spectacular, it is also very expensive and disagreeable in the foundry.

11. Ladle bottom tile are easy to install and economical to use. A layer of mud is spread over the bottom to act as a cushion, and the proper size bottom tile placed directly on top of this mud. The sides are lined in the usual manner, dried thoroughly, and the ladle is ready to use. The bottom will be dry, so there will be no boiling. The tile makes a solid bottom free of joints, and there is very little chance of a bottom spill. The cost of installing these ladle bottom tile is very low compared to the cost of fitting 9-in. straights or splits.

RAMMED LININGS FOR MIXING LADLES

12. The use of the forehearth in foundry practice has been greatly extended during the past few years, especially in those foundries making specification castings where desulphurizing is accomplished with fused soda ash. Different types of forehearths for mixing and refining have been described by Evans[8].

13. Fused soda ash and soda bearing slags are very corrosive to most refractory materials, and if desulphurizing were to be advanced to its present high efficiency, some type of refractory had to be developed that would withstand this action to a commercially satisfactory degree. In working on this problem, the writers have had the complete cooperation of the fused soda ash producers, the ladle manufacturers, and a number of foundrymen making different types of castings.

14. It was found that ordinary fire brick, even though highly refractory, gave very poor service in these ladles. The rapid failure was due to shrinkage and vitrification of the exposed face, and corrosion between the joints by the soda slags. These eroded

[8]Evans, Geo. S., "The Forehearth as Used in Iron Foundry Practice," TRANS. A.F.A., vol. 32, pp. 1-20 (1934).

joints soon allowed the lining to fall into the ladle. Fig. 4 shows an example of this type of corrosion.

15.   Steel plant ladle brick were used in some foundries with satisfactory results, because of their high density and their tendency to expand and fuse at the joints, at the temperature of molten iron.   Other foundries, however, found that these brick were not the answer to their problem because of their low refractoriness, poor resistance to soda slags, and the high cost of maintenance.

16.   Ordinary ladle mud, made from mixtures of silica sand, molding sand, and fire clay could not be used at all because of excessive corrosion by the soda slags.   At the temperature of
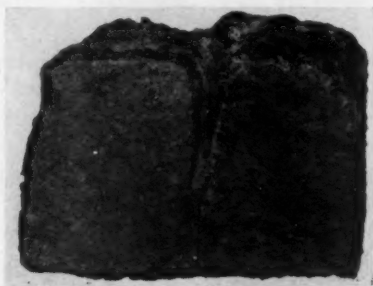
FIG. 4—CORRODED BRICK FROM LADLE LININGS.

cupola metal, these slags severely attack free silica and fire clay, form new silicates, and dilute the refining action of the soda ash.

17.   Various types of plastic-refractory ramming materials were then tried, and during the past three years, they seem to have given better results than other materials available.   When made from highly calcined, highly refractory materials, with crushed, splintery grains, properly sized and classified, then blended with a minimum amount of refractory bonding material, these plastics give a hard, dense, refractory lining—one that is free of joints and uniformly resistant to the cutting action of soda ash and soda slags.

18.   Not all plastic refractories will work satisfactorily in ladle linings.   Laboratory tests indicate that they vary greatly in P.C.E. value (fusion point), in the degree of shrinkage causing the formation of joints and cracks, in their density, and in their resistance to soda slags.   Plastics made from reclaimed brick bats were found to be unsatisfactory for ladle use.   They made such

a porous lining, that large amounts of corrosive slag were soaked up by the lining, causing premature failure. Furthermore, it was difficult to ram such plastics, and shrinkage and cracked linings resulted.

19. Rammed refractory linings are being used successfully in many types of forehearths and ladles; for example, the box, drum, and tub-type forehearths, teapot ladles, standard mixing ladles, elongated teapot ladles, "U"-shaped teapot ladles, covered and insulated ladles, and standard crane ladles.

20. These ramming refractories are used in automotive foundries, pipe foundries, radiator shops, foundries making high grade pressure castings, hardware foundries, agricultural equipment foundries, sanitary ware shops, malleable foundries, and in non-ferrous foundries.

21. Plastic refractories for foundry use usually are packed in 500 lb. metal drums to preserve the moisture content. If the material should happen to dry out, easy workability can be restored by moistening with water. If it should be too moist for proper ramming, it can be dried slightly by opening the drum, cutting the material into chunks, and allowing it to air dry for a short time.

22. In lining ladles with a rammed refractory, a layer of fire brick is placed against the shell or against the insulation. Then a greased, collapsible, wooden form is placed inside of this backup lining. The refractory is fed in around this form in small chunks and rammed solidly with an air hammer. The more it is rammed, the better service it will give. In large ladles, the rammed lining should be at least 5-in. thick. This type of refractory may also be used in ramming the bottom, but as mentioned previously, ladle bottom tile eliminate boiling, and other disagreeable phenomena. After the ramming has been completed, the ladle should be dried thoroughly in a drying oven, or with a torch. Teapot spouts and covers of insulated ladles also are lined with this type of refractory material.

23. A pipe foundry using a ten ton, "U"-shaped, elongated, covered, teapot ladle lined the cover with 1400 lb. of a plastic ramming refractory, and it gave three months service with fused soda ash. A foundry making Diesel engine castings found that they secured cleaner iron from rammed ladle linings.

24. In using a 5-in. rammed lining in a 9000 lb. receiving ladle, one foundry secured forty, 50-ton heats from the lining.

However, some patching was necessary during the last ten heats.

25. Several foundries have reported that the labor cost of installing a rammed lining is less than half the cost of placing

Fig. 5—Teapot Ladle Lined with Rammed Refractory.

Fig. 6—Teapot Ladle with Rammed Lining in a Pipe Foundry.

a brick lining, and the rammed lining lasts twice as long as fire brick.

DISCUSSION

Presiding: ELMER J. CARMODY, Ada, Mich.

THOS. R. WALKER, JR.[1]: I do not quite understand what the material was that was used in these linings. Was there some silica pebble mixed in with that lining? If not, has Mr. Bales had any experience with the mixture of silica pebble and refractory bond or clay applicable especially to ladle lining?

MR. BALES: I have not had any experience with the use of silica pebbles but sometime ago, I was told there were some foundries using that material. How it has worked out, I do not know, but I believe that if you get too much silica gravel in there against these soda slags, the life would be quite short. Most of these ramming materials are made of calcined flint clay. Clay usually has a higher fusion point than silica pebbles.

JOHN R. CLAUSEN[2]: I feel that the speaker has a very good point in his brick for hand ladles. We have used a very deep clay skimmer on hand ladles. In pouring stacked clutch castings, we have had to resort to a pouring basin to stop the slag. The slag will sneak in along the sides of the clay skimmer. This brick should overcome that trouble.

A. W. GREGG[3]: Mr. Bales, what size ladles would you recommend for use with the bottom slabs? Another question is on these desulphurizing ladles. In the pictures that I have seen and the installations I have heard about, it is not the brick, it is the joint material that fails. Is not possible to find some material that will stand up better between the brick than the material we have now? I have heard good reports on brick linings and at the same time the brick fails at the joint.

MR. BALES: On the size of the ladle bottom tile, they have been made in various diameters. They have been made as large as 30-in. in diameter. If you get them any larger than that, they are rather difficult to make without developing cracks in them. Incidentally, that 30-in. brick is 4-in. thick instead of 2-in.

Now on desulphurizing, the question of density that Mr. Donoho brought up in a previous paper is important there. I have never seen a ladle lining, or any other brick work, that would not have some kind of joint in it. Now these refractory cements, and there are a lot of

them on the market, unfortunately are very severely attacked by fused
soda ash. Sometime ago, I believe it was at the Rolla, Mo. conference
of the St. Louis Chapter, one speaker gave a paper on extremely hard
burned brick and his experience seemed to indicate that they gave
better results than regular fire brick. But he had the problem, that
he could not get anything that would make a satisfactory joint. I do
not know of anything you can use that will produce a joint that will be
equally resistant to that slag action as the brick itself.

MR. WALKER: It might be interesting to this assembly to know
there has been a mixture of refractory fire clay and silica pebbles, in
use in linings for a transfer ladle which is used on a crane for handling
gray iron, where we get an average of about five weeks' life out of a
transfer ladle with this type of lining. It is a very easy job and a
rather quick affair.

CHAIRMAN CARMODY: In that case, what is your binder, a clay
binder?

MR. WALKER: We use a refractory clay binder. It is a New Jersey
clay. It is mixed with five parts of silica pebbles, which probably run
between 90 to 95 per cent silica, and about three parts of a refractory
bond clay and about one part of a subangular type of sand, which runs
around 85 to 90 per cent silica.

MEMBER: I would like to ask has anyone tried a basic lining in the
forehearth, where the soda ash is used? Does anyone know of any
reason why it would not work?

E. M. SARAFF[4]: Basic ladle linings are standard practice for
handling certain types of steel, especially the manganese steels. I do
not know of the use of basic linings to withstand the action of sodium
carbonate, as, for example, in a desulphurizing ladle. For the latter
practice, my suggestion would be to use an extra hard burned refractory
of the super duty clay brick type, using minimum joints of finely ground
mixture of clays corresponding to those in the brick themselves.

JOHN LOWE[5]: These ladle bottom tile have cut our lining costs
approximately 50 per cent at our No. 3 plant in Muskegon.

MEMBER: Mr. Bales stated he obtained forty 50-ton heats out of a
rammed ladle. I wonder if that has been done with intermittent
patching?

MR. BALES: The last ten heats were patched. Patch work is
adaptable to that ramming method.

CHAIRMAN CARMODY: Mr. Lowe brought up the cost question, and
Mr. Bales mentioned the price. Would Mr. Bales mention something
about the comparison of the price of his solid tile as compared to his
brick? Have you any actual figures for that?

MR. BALES: No, I do not. I was hoping some of the foundrymen
here would give us some such figures. The reason I made the statement
was because when you are in a foundry and see the brick masons or
ladle men doing a lot of trimming with splits and 9-in. brick, and then
see them simply drop the ladle bottom tile in place, I know it costs a
whole lot more from the labor point of view to use splits and 9-in.

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[4] Harbison-Walker Refractories Co., Cleveland, O.
[5] Melting Supt., Centrifugal Fusing Co., Lansing, Mich.

brick than it does to use the bottom tile. But, so far as actual cost figures are concerned, I do not have them.

A. S. NICHOLS[6]: I wish to ask Mr. Walker a question about the ladle lining being made with silica pebbles. If you are using silica pebbles with a New Jersey clay high in silica content, isn't that practice contrary to what Mr. Bales was talking about in soda ash?

MR. WALKER: I am not prepared to answer that particular question. I am more or less confining myself to facts in connection with our practice. We are using a silica pebble lining. It is a mixture of this New Jersey clay, as I said, with silica pebble, as a lining for a transfer ladle. It is standing up quite well. We are getting some 25 to 30, (on an average) 8-hour days out of a transfer ladle which is handling some 80 to 100 tons of iron a day. And we are using, you might be interested to know, soda ash in the ladle ahead of this ladle. That is, we are bringing our iron and slag out of the cupola together into a forehearth in which we are using soda ash at the rate of one to one and one-half pounds per ton. The soda ash is, of course, a sodium carbonate flux, and we are getting what we consider to be economic results from our mixture in our rammed linings.

MR. BALES: I think one reason Mr. Walker can use a mixture of that type and some of the automotive shops cannot, is because of the lower temperature of pouring. He is making pipe which is poured at much lower temperatures than cylinder blocks and other automotive castings. Furthermore, he is not using soda ash in that ladle.

MR. WALKER: We are bringing our iron out of the cupola at some 2700° to 2750°F. and it is going into this ladle at some 2650° to 2675°F. on an average.

MR. BALES: I think that is the answer. The people who make automotive castings and thin section castings have to pour at a much higher temperature.

MEMBER: Mr. Walker, what size are these silica pebbles?

MR. WALKER: I do not know what screen they might pass. My offhand judgment is they are an average of a quarter of an inch across, the maximum size, a quarter to five-sixteenths, something like that. They run rather uniform.

A question has just been asked me about handling the iron in the well zone of a front slagging type of cupola. For those having this particular problem with that phase of cupola operation, I would be very glad to recite our experience. We have been using a special block in the spout of the cupola. We have three cupolas with an average trough length of some 12 ft. We are using a block for the trough which we have made up specially for us. A trough is built into the block. The blocks have an over-all dimension of 6-in. high by 9-in. wide by 12-in. long. A trough is cut in the blocks which is the shape that we desire in our cupola spout. We lay these in and cement them together with ordinary refractory clay, and get very satisfactory life from these blocks. They are not particularly expensive. We replace them at the rate of about less than one per shift of 8 hours. It works quite satisfactorily.

⁶ Illinois Clay Products Co., Chicago, Ill.

Cupola Operation

BY DONALD J. REESE,[1] BAYONNE, N. J.

Abstract

The author first gives his broad definition of the complete melting process, then discusses the cupola as a long time investment. One of the most important factors listed is that of the amount of coke on the bed charge. The use of the smallest size of cupola consistant with production demands is stressed. The author then goes into a detailed discussion of fuel burning rates, melting rates, normal operating conditions, wind pressure requirements, coke charges and metal charges giving formulas and examples. Then he discusses such factors as getting hot iron, use of steel and limestone. The following section of the paper is devoted to factors in determining the requirements of the measurements of a cupola and parts. Tuyere performance, blower types, gauges and metering devices are then discussed. The author concludes his paper with sections on control of moisture in the blast, preheated air and factors related to temperature of iron at spout and at pouring.

1. To the writer, the term "cupola" means the complete melting process from the purchase of raw materials, such as pig iron, scrap iron, steel, coke and limestone, to the pouring of molten metal into molds. Even so, it is not always feasible to drop responsibilities for the quality of the metal when it is ready for the mold, for the size and location of gates, the size and location of risers and other factors pertaining to the mold may have a definite bearing on the finished product.

MELTING AND MOLDING RESPONSIBILITIES OVERLAP

2. The responsibilities of melting and molding departments overlap and unless there is mutual appreciation of this fact, it may be the prime cause of failure to produce the best quality most

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economically. For instance, some years ago, a foundry was having a run of hard luck in making one type of casting which weighed from one to ten tons each. The melting department claimed the cause of rejection was some mold condition, while the molding department was sure the metal was at fault. While each department maintained the position that the other was at fault, the loss continued. The trouble was finally traced to a rusty core wire used to assemble two halves of a very large core. A large tonnage of rejected castings testifies to the unnecessary expense incurred when these two departments failed to collaborate fully.

3. Unless those responsible for the quality of the product and the economy of the melting process have a thorough knowledge of their jobs and complete supervisory authority over the melting process, either the quality of the product will fail to meet specifications or the costs of production will be more than the traffic will bear. As the casting of metals progresses from an "art" to a "science," what is discussed here will become more and more important if we, as individuals in the foundry industry, are to survive and if our particular foundry is to remain a part of the industry.

SCOPE OF CUPOLA OPERATION

4. To clarify what is meant here by "cupola," to repeat, it means all the raw materials used, the method of charging them into the cupola, the cupola with all its accessory equipment such as the blower, blower motor, pressure gauges, volume gauges and air metering devices, slagging and tapping arrangements, preheated air equipment, humidity control equipment, etc.

5. The usual product of a cupola is gray iron but it is also used to produce white iron, malleable iron, bronze, copper and also other metals when used in conjunction with another furnace as in a duplex process. In a general way, this discussion will be applicable to any of these metals but, unless otherwise stated, it is to be assumed that gray iron production is meant and more specifically a high quality grade of gray iron.

THE CUPOLA — A LONG TIME INVESTMENT

6. In approaching the selection of equipment, how many foundrymen realize that they are making a long time investment? The normal life of a cupola, when reasonably maintained, is between 20 and 30 years, seldom less than 10 years and occasionally more than 50 years. Undoubtedly, there are many cupolas in

England or on the continent that have seen many years of service beyond the 50 year mark.

7. The total cost of a cupola erected and lined, ready for operation, and including accessory equipment ranges from $1,000, for a very small cupola, to about $10,000, for a very large one. These figures, of course, are averages for it is possible to have a much larger investment. In a large foundry which only has one cupola, the investment might be as much as $30,000 or more. The important point, however, is that this investment, when pro rated over 20 to 30 years production and with the large tonnages of metal produced, is, in reality, a very small item.

8. To illustrate the point, it is possible to melt 5,000 tons of iron per year in a cupola lined to 27 in. which in a period of 10 or 20 years should amount to 50,000 or 100,000 tons of iron. The original cost of equipment when pro rated over this tonnage would be in the neighborhood of five cents per ton. The bearing on production costs of the original investment is probably of less importance than any other single item included in production costs.

Importance of Amount of Coke on Bed

9. During the life of the cupola, it is probable that it will be operated some 5,000 to 10,000 times or more, and it is in this daily repetition of performance that costs should be studied. For example, if a cupola is lined to 60 in. when one lined to 54 in. would serve equally well, then about 500 lb. more coke is used in the coke bed each day than is necessary, for the bed coke of a 60 in. cupola is about 2,800 lb. and that for a 54 in. cupola is 2,300 lb. In the life span of the cupola, this item alone might easily be from five to ten times as much as the original investment.

Importance of the Smaller Cupola

10. When selecting melting equipment, it is desirable to select the smallest size that will render satisfactory performance when all factors have been duly considered. If the problem is to melt 20 tons of iron per hour for eight hours per day, then there is little choice other than to select a cupola lined to 72 in., except that certain operating conditions might make it feasible for the job to be accomplished in a 66 in. or that it might be necessary to select one lined to 78 in.

11. Though the hourly tonnage of metal which the foundry can handle is the most important guide in the selection of the proper

size cupola, it is also important that the type of castings to be made
and the size of scrap materials available for the cupola charge
be considered.

12.   For example, if the largest casting to be made requires
ten tons to pour, then it may be necessary to have a cupola large
enough to melt ten tons of iron in less than one hour, even though
the average requirement is only six or seven tons per hour. It
might also be possible that the foundry is making high strength
gray iron, in which case, the risers might be of considerable size,
making it necessary to select a cupola in which this returned
material can be charged without danger of "bridging" the cupola.

INSIDE DIAMETER AND MELTING RATE

13.   The cupola furnace is a fuel burning unit just as is the
furnace in your home or the furnace in a steam generating plant,
but instead of producing hot air or steam, it produces hot metal.
When the inside diameter of the cupola furnace is fixed, the grate
area and to a large extent the fuel burning rate is fixed.

14.   The next step is to select a proper coke charge. In the
selection of the coke charge, the operator has the privilege of
deciding to a large extent whether the cupola will have a high
or a low degree of thermal efficiency. By thermal efficiency is
meant the percentage of heat units contained in the molten iron
when the fuel is burned in perfect combustion. The average thermal
efficiency of the cupola furnace in the industry is about 25 per cent.
A very high degree of thermal efficiency is as much as 65 per cent.
It is well to point out, however, that even though the thermal
efficiency of the average cupola furnace is only 25 per cent, this
figure is about twice as good as other types of furnaces used for
melting iron. It is also known that only a small part of successful
cupola operation is attributed to good thermal efficiency, and unless
other important factors are considered, nothing can be accomplished
by limiting our thoughts solely to thermal efficiency.

Table 1

CUPOLA FUEL BURNING RATES

Cupola Diameter....	24	30	36	42	48	54	60	66	72
Fuel Per Hour (coke)	450	700	1000	1400	1800	2300	2800	3400	4000

15.   Table 1 shows the "Fuel Burning Rates" of nine sizes
of cupolas. The "Fuel Burning Rate" of the cupola, however, is

dependent on the amount of air that is delivered to the fuel consuming zone of the cupola (between the tuyeres and the melting zone). The actual amount of air reaching this zone varies from 50 per cent of the normal air to 150 per cent for any one size cupola in the practices that have been established in different foundries. This does not differ from the furnace in your home or the furnace in a steam generating plant where the furnace load may be reduced to some low point when the demand for heat or steam is low or increased when the demand is high. It is also true that any fixed combustion system has a definite range in which it can work satisfactorily with a minimum and a maximum.

16. In approaching an understanding of the melting rate of a cupola furnace, it is possible to get a much clearer conception of what may be expected from a given size cupola operated under definite conditions by considering the "Fuel Burning Rate" and a very simple formula can be devised to check performance. This is shown in Formula 1, where P equals wind box pressure.

FORMULA 1:

$$\frac{\text{Fuel Burning Rate (in lb.)} \times \text{Fuel Ratio} \times \frac{1}{4} \sqrt{P}}{2000} = \text{Melting Rate (tons per hr.)}$$

MELTING RATE

17. In Table 1, the normal fuel burning rate of a 48-in. inside diameter cupola is 1800 lb. per hr. Thus, if enough air is supplied to cause a wind box pressure of 16 oz. and the fuel ratio is 10 to 1 (10 iron and 1 coke), the melting rate would be 9 tons per hr. as found by substituting these values in Formula 1 as:

$$\text{Melting Rate} = \frac{1800 \times 10 \times \frac{1}{4} \sqrt{16}}{2000} = 9 \text{ tons per hr.}$$

18. If the fuel ratio was 8 to 1 instead of 10 to 1, the melting rate would be 7.2 tons per hr., or if the amount of air was only enough to give us a 9-oz. wind box pressure, even though our fuel ratio was 10 to 1, the melting rate would be about 6.8 tons per hr. If this cupola happened to be in use in the automotive industry, where fuel ratios are 9 to 1 and wind box pressures of 25 oz. are not uncommon,·the melting rate would be 10.1 tons per hr.

19. It is to be noted that if it is desirable to slow down the melting rate of a cupola, it can be done by using more coke or less air, and if speeding-up the melting rate is required, use more air

or less coke. The melting range of any cupola may be anywhere between one-half normal to one and one-half normal or from 4.5 to 13.5 tons per hr. in a cupola lined to 48-in.

NORMAL OPERATING CONDITIONS

20. Table 2 shows the normal operating conditions for nine sizes of cupolas using a 16 oz. wind box pressure as an indication of normal operation with a 10 to 1 fuel ratio.

Table 2
NORMAL CUPOLA OPERATING CONDITIONS

Cupola Diameter..	24	30	36	42	48	54	60	66	72
Tons per hr.......	2¼	3½	5	7	9	11½	14	17	20
Air—cu. ft. (Std.) .	1000	1570	2270	3040	4025	5100	6300	7610	8930
Air in lb.........	78	123	177	237	314	398	492	593	696

PRESSURE GAUGE INDICATIONS

21. Most foundrymen are not clear as to the significance of what the cupola pressure gauge indicates. The cupola pressure gauge indicates the resistance the cupola offers to the flow of air through it and, contrary to general belief, the area of greatest restriction is not at the tuyeres but inside the cupola in the melting zone area.

Tuyere Ratios and Areas

22. Most foundrymen talk in terms of ratios between the area of the cupola and the total area of the tuyeres. A common ratio is 4 to 1 which means that the total area of the tuyeres, when the area is measured at the cupola shell is ¼ the area of the cupola. On large cupolas, the tuyeres area is more likely to be 1/6 of the cupola area, while those foundrymen who believe in large tuyere areas might have them enlarged enough so that this area is equal to ½ the cupola area.

23. When measured in percentage, the tuyere area is normally 25 per cent of the cupola area, is as low as 16 per cent in large cupolas and sometimes as large as 50 per cent of the cupola area. However, that area which is inherent within the cupola, and which determines what the cupola pressure is to be, is only about 1½ per cent of the cupola area. In other words, in a cupola having a tuyere area of 1/6 of the cupola area, the tuyere area is still about 10 times as much as the bottle neck within the cupola while with a tuyere area of ½, the tuyere area is some 35 times as much as the limiting area.

24. The area of a cupola lined to 48 in. is 1810 sq. in. The area of the tuyeres would be approximately 450 sq. in., whereas the effective area of the bottle neck of the cupola is only 27 sq. in. In effect, this is equivalent to suspending a 6-in. inside diameter pipe in the center of the cupola at a height of the melting zone and then sealing off the rest of the cupola in this area from air passage. In reality, this limiting area is probably some six times larger than the 27 sq. in. which was calculated, for it must accommodate gases which are close to 3000°F. in temperature, but when measured in terms of cold air, the effective area is only equivalent to 27 sq. in.

25. There are two ways in which foundrymen may direct their thoughts toward the bottle neck of a cupola inherent in the melting zone and away from the tuyeres. One way is to have an air tight system between the blower and the melting zone of the cupola with an accurate means of determining the air volume and wind box pressure. If this equipment were available, the air volume could be lowered and the resultant wind box pressure read, then increase the air volume and check the resultant wind box pressure. If this was done on a 48-in. inside diameter cupola, the result would be that with an air volume of 3960 cu. ft. the pressure would be 16 oz., with 1980 cu. ft. the pressure would be 4 oz. and with 5940 cu. ft. the pressure would be 38 oz.

26. One of the standard formulae used in determining the pressure of liquids in pipes is $V = 4005 \sqrt{P}$, where V is velocity and P or pressure is measured in inches of water. Formula 2 is this same formula in which P is changed from inches of water to ounces.

FORMULA 2: $V = 5280 \sqrt{P}$

27. Although V measures velocity, it is easily converted to volume by dividing by the area of the cupola bottle neck or an area equivalent to 1.5 per cent of the cupola area. Thus, in a 48-in. cupola, the area of the bottle neck is 27 sq. in. If the volume of air being used is to be determined with a 16 oz. wind box pressure, substitute in the formula $V = 5280 \sqrt{P}$ and the velocity is 21,120 ft. per min. If the velocity through 27 sq. in. is 21,120 ft. per min., then the number of cubic feet is 27/144 of 21,120 or 3960 cu. ft. Table 3 shows the relationship between wind box pressure and normal air delivery reaching the cupola melting zone.

Table 3

NORMAL AIR DELIVERY TO CUPOLA MELTING ZONE

Wind Box Press. (oz.)	4	8	12	16	20	24	28	32	36
Per Cent of Normal Air	50	70.6	86.5	100	111.8	122.5	132	141.5	150

PRESSURE GAUGES

28. Attention is to be directed toward such a simple device as a cupola pressure gauge, which most foundrymen either ignore or do not even consider important enough to install on their cupola, or, if installed, fail to repair it in case it is broken. When a person is ill and requires the services of a doctor, the first thing the doctor does is to get out his stethoscope and check the patient's heart condition. When a cupola is ill, the first thing a cupola doctor would like to check is the pressure gauge. Unfortunately, more than 75 per cent of the cupolas are not equipped with this instrument which costs less than $10.00.

29. When using the formula $V = 5280 \sqrt{P}$, a figure of 21,120 ft. per min. for the velocity of air up through this cupola bottle neck was desired. When the air is cut in half from 3960 cu. ft. per min. to 1980 cu. ft., the velocity is reduced from 21,120 ft. per min. to 10,560 ft. per min. and when the air is increased from 3960 cu. ft. to 5940 cu. ft. the velocity was increased from 21,120 to 31,680 ft. per min.

30. Although the reaction between incandescent coke and pre-heated air is almost instantaneous, there is some small increment of time involved and assuming that a 48 in. bed is good for a cupola operating at 16 oz. pressure, it is possible to calculate the time element, knowing the velocity of air travel (21,120 ft. per min.) and the height of the bed, which is 48 in. or 4 ft. This time factor is about 1/88 of a second. When the volume of air is reduced, the velocity of air travel in the bed zone of a cupola is reduced and therefore the height of the coke bed necessary to consume all the free oxygen in the air before it reaches the melting zone is less. Table 4 shows bed heights for various cupola pressures.

Table 4

CUPOLA BED HEIGHTS

Pressure (oz.)	4	8	12	16	20	24	28	32	36
Minimum Bed—(In. above tuyeres)	21	30	36	42	47	52	56	60	63
Recommended Bed (In. above tuyeres)	27	36	42	48	53	58	62	66	69

COKE BED HEIGHT

31. In foundry discussions on cupola melting practice, the question on "what is the proper coke bed height for a cupola?" is always raised. In Table 4, for a cupola operating at 4 oz., a bed of 27 in. is sufficient; for a cupola operating at 16 oz., the bed should be 48 in., and if the cupola is operating at 36 oz., the bed should be 69 in. Thus, the answer to a question on the proper coke bed height depends on the type of practice that has been established.

32. In the discussion so far, points have been covered that may be confusing to foundrymen, inasmuch as few of the points covered have ever been called to attention. Now some phases of equipment and operation will be covered.

SIZE OF MATERIALS

33. Assume a cupola that would be lined to 48 in. It would be desirable to use a 4-in. coke and limit the size of scrap and pig to 14 in. Table 5 lists the sizes of materials for nine sizes of cupolas. Attention is directed to the fact that most foundrymen use larger materials than are desirable for the best type of cupola operation.

Table 5

CUPOLA MATERIALS SIZES

Cupola Diameter	24	30	36	42	48	54	60	66	72
Size of Coke	2	2½	3	3½	4	4½	5	5½	6
Size of Scrap	7	9	11	13	14	16	18	20	22

SIZES OF CHARGES

34. After considering the size of materials in the cupola charge, it is necessary to consider the quantity. Suggested coke and iron charges are shown in Table 6.

Table 6

SUGGESTED COKE AND IRON CHARGES FOR CUPOLAS

Cupola Diameter (in.)	24	30	36	42	48	54	60	66	72
Coke Charges (lb.)	40	60	90	120	160	200	250	300	360
Metal Charges (lb.)									
With No Steel	400	600	900	1200	1600	2000	2500	3000	3600
With 20% Steel	360	540	810	1090	1440	1800	2250	2700	3240
With 40% Steel	340	510	765	1020	1360	1700	2125	2550	3060
With 60% Steel	320	480	720	960	1260	1600	2100	2400	2880

35. It will be noted that the figures in Table 6 are based on using fixed coke charges for any given size cupola and any adjustment is to be made in the iron charge. Normally, common practice is to decide the weight unit of the cupola iron charge and then decide what the fuel ratio will be from which the coke charge is derived.

36. For example, if a 2000 lb. iron charge is used and an 8 to 1 fuel ratio, the coke charge is 250 lb. Then, if the iron is not sufficiently hot, foundrymen use more coke, winding up with 300 lb. or more of coke. Assume a 48 in. cupola. That would mean that a 300 lb. coke charge would be about 11 to $11\frac{1}{2}$ in. deep. It would result in poor combustion conditions when this coke charge finally reached the cupola melting zone tending to lower the melting rate of the cupola and offering little possibility of producing high melting temperatures. Some coke charges are large enough to mean a depth of 15 in. or more.

GETTING HOT IRON

37. When the weights of the coke and iron charges are large and cupola trouble develops, which generally means difficulty in getting the iron as not as desired, the first thing usually done, is to use more coke. In most cases, this may do more harm than good for the coke charges are already too much for the cupola. In any event, the foundryman will do a lot of sweating before he secures the favorable change desired which most likely will be when he reduces the amount of coke below the original amount.

USE OF STEEL

38. In Table 6, it will be noted that the cupola metal charges decrease in weight as the amount of steel in the cupola charge increases. Many think that a cupola operated with steel in the charge requires more fuel to melt the steel and also that the cupola melting rate is lowered. When recommending less weight in a cupola charge containing steel than in one where no steel is used, the prime reason is to still have sufficient coke left to replenish the coke bed after the steel containing charge has dissolved carbon from the fuel. A charge containing 60 per cent steel may dissolve as much as 1.75 per cent of its carbon from the fuel, which means about 35 lb. of carbon in each ton of iron or about 40 lb. of coke.

39. As a matter of fact, when steel is used in the cupola charge, the amount of fuel that the charge will dissolve is seldom compensated fully and, therefore, the over-all practice is better

than on those charges not containing steel. Inasmuch as steel-containing charges result in higher metal temperatures at the cupola spout, a double effect is secured, one of producing an iron containing more heat units and the other of gaining the added temperature with less fuel. The melting speed of a cupola improves with the fuel ratio and when melting steel, the melting rate is generally higher than when melting no steel.

LIMESTONE

40. The amount of limestone that should be used in a cupola charge depends on the type of materials used in the charges. The writer uses an amount equal to 30 to 40 per cent of the weight of the coke charge and prefers to use on top of the coke bed an amount equal to three or four of the following limestone charges. It is desirable to have a volume of slag equivalent to 60 to 90 lb. per ton of iron. A cupola slag serves to protect the molten iron in the cupola well from the cooling air currents of the entering cold air. The slag also acts as a filter to separate the non-metallic impurities attached to the drops of iron as they move from the melting zone into the cupola well.

APPLYING THEORIES — EQUIPPING FOR MELTING

41. Up to this point, some of the theories of operation have been covered. It might be wise now to see how to apply these theories in starting from scratch to equip a foundry for melting iron.

Days Melting Demand

42. Before interviewing the representatives of the manufacturers of equipment, the demand of the melting equipment each day and preferably each hour of the day must be determined. Further, just what type or types of castings the foundry will make must be decided.

43. In a general way, these considerations will determine whether the cupola and its accessory equipment shall be small, medium or large. If it is a small cupola, the possible selection is narrowed down to one size of several possibilities of say 24, 30 or 36 in. inside diameter, or to a medium size which may be 42, 48 or 54 in. inside diameter or if it's a large cupola, selecting from a choice of 60, 66 or 72 in.

Vertical Dimensions

44.   After selecting the proper size cupola, vertical dimensions will have to be established.

45.   *From Base Plate to Charging Door:* What shall the height be from the base plate of the cupola to the lower sill of the charging door? A satisfactory dimension is 18 to 22 ft. If the height is less than 18 ft., full advantage cannot be taken of the hot gases passing up through the cupola to preheat the materials going down. This is a point of fuel economy that may save several thousand dollars in lower fuel costs each year.

46.   *Height of Bottom Doors above Foundry Floor:* Next consideration is the height of the bottom doors, in raised position, above the cupola floor line. This height will be about 7 ft. 6 in. for an average size cupola, if we wish to use a reservoir type of ladle. However, this dimension will vary somewhat with the size of the cupola, being lower for small cupolas and higher for large cupolas.

47.   *Over-all Height of Cupola:* Next to be established is the over-all height of the cupola. This is sometimes regulated by city ordinances which may specify that the height shall not be less than 10 ft. above the peak of the foundry roof. If there are other buildings nearby, especially office buildings, it will be desirable to have the cupola stack extend above these buildings so that the cupola gases will not blow into them.

Apron for Stack Opening

48.   An apron must be placed around the cupola at the point where it passes through the roof to keep the elements (snow, rain, wind) out of the foundry and it will be necessary to determine the slope of the roof in order to build the apron or roof hood, as it is generally called.

Blast Pipe Connection

49.   Next a decision must be reached on just where the blast pipe from the cupola will be attached to the wind box and whether the connection will be made at one or two points. In general, one point should be selected as a double wind box entrance involves some fancy blast pipe which eventually will wear out and need replacement. Being considerably more expensive than a single blast entrance, the replacement will be delayed as much as possible.

50.   In fact, the foundry may try to get along with a very poor

blast line for years. This point may not seem very important but the air loss between the blower and the melting zone of the cupola in most foundries is seldom less than 20 per cent and frequently is as much as 50 per cent of the total air delivered by the blower. Should an air metering or air weighing device be installed in a leaky air circuit, the instrument will only tell how much air is passing the metering point and will only be relative instead of absolute as compared to the quantity delivered to the melting zone.

51. Up to this point, our most important selection has been the determination of the inside diameter of the cupola and next in importance has been the determination of the cupola working height between the base plate and the lower sill of the charging door.

Spout, Tuyeres, Segments, etc.

52. If the person constructing the cupola is well versed in cupola operation, he will pay considerable attention to the location of the tap spout, the slag spout, the tuyeres and type of tuyeres, the type of angle supports to hold the lining and where they are located, the construction of the cupola around the charging door and of the re-enforcement of the cupola in this zone. If he decides to use mechanical charging, that means a large door. The selection of a suitable spark arrestor will be dependent on the locality in which the cupola is to be installed.

53. *Cast Iron Lining Segments:* These are better than angle iron because they will last longer in the presence of heat.

54. *Breast Opening:* The breast opening for the tap spout should allow sufficient height for the sand bottom in the cupola and provide for the tap-out block.

55. *Slag Spout:* The slag spout should be 5 or 6 in. below the lower tuyere plate.

56. *Elevation of Tuyeres:* The elevation of the tuyeres will be largely dependent on the type of operation used. Normally the bottom plate of the lower tuyeres is 20 to 24 in. above the base plate of the cupola.

SLAGGING SPOUT

57. A modern trend is to use a front slagging spout, shown in Fig. 1. When using this type of spout, rear slagging is eliminated. The construction of the spout is dependent on the wind box pressure, and the elevation of the tuyeres in the cupola must be

high enough so that when the blast is shut off, any metal which backs up into the cupola will not raise the slag line inside the cupola to such an extent that slag will flow into the tuyeres and seal them off from the wind box.

58. The end of a front slagging spout away from the cupola is at a higher elevation than the top edge of the tap hole. For example, the elevation of the spout on a cupola operating at 16 oz. blast pressure would be 4½ in. (divide the wind box pressure by 4 and add ½ in.; i. e. 16 oz. pressure divided by a 4 + ½ in. = 4½ in.). This head of metal in the tap spout is just sufficient to

X Inches	P Ounces
1.5	4
2.5	8
3.5	12
4.5	16
5.5	20
6.5	24
7.5	28
8.5	32
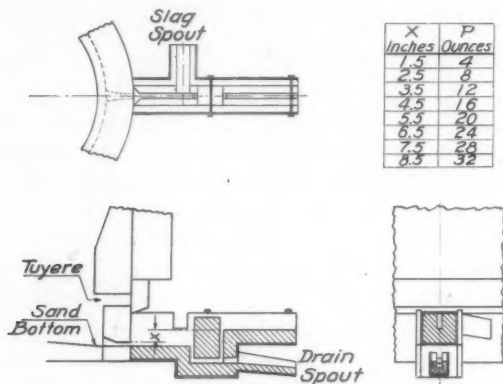
FIG. 1—DIAGRAM OF CONSTRUCTION OF A FRONT SLAGGING SPOUT.

overcome the pressure inside the cupola and keeps the tap hole sealed so that air will not blow through. If the safety head of metal in the spout is made too great, the slag will not be able to flow through the tap hole.

59. A spout of this type should be fabricated of ½-in. plate with tie bars across the dam and at the far end of the spout to keep it from spreading. A dam is inserted in the spout to separate slag from metal and a slag spout is attached at some point between the cupola and the dam. The depth of metal at the dam should not be less than 10 in. to get a positive separation of slag and metal.

60. When the spout is put into service, the brick should be preheated to as near a white heat as possible. An auxiliary spout will allow the first metal melted to run on through the spout as with an ordinary spout. Then after a small quantity of metal has been

melted, the auxiliary spout may be botted up just as a cupola is botted-up.

61.   The width of the spout should be narrow, from 1½ to 2 in., so as to hold down the volume of metal held in the spout, but the spout may be flared at the cupola shell enough to permit the cupola man to work on the cupola breast.

62.   Front slagging spouts are most desirable for continuous tapping, being excellently adapted to continuous melting on cupolas melting less than 5 tons of iron per hour. However, front slagging spouts may be used on cupolas in which several different types of iron are being melted provided a spreader coke charge is used between these irons. It is very desirable to use where low total carbon-high strength irons are desired.

TUYERES

63.   Most foundrymen do considerable thinking about the design of their tuyeres and few have refrained from making a tuyere pattern of their own and installing the tuyere castings made from their pattern in their cupolas. Hundreds of dollars have been spent changing tuyeres and those who have been in the foundry for 15 or 20 years have probably designed and installed several different types of tuyeres.

64.   To the writer, a good tuyere will prevent, as much as possible, slag, slag coated coke and patching from accumulating at the tuyere and a poor tuyere will not. We believe this is accomplished largely by having an overhanging and replaceable cover plate over the top of the tuyere.

65.   To the writer, there are many details of cupola construction and operation that are considerably more important than the tuyeres and he would prefer to leave tuyere changes alone, until other features have been corrected. The actual number of rows of tuyeres that are used will be discussed further on in this paper.

SELECTION OF BLOWERS

66.   The selection of a suitable blower is probably more important than the selection of the proper size cupola, and because most foundrymen find the subject of air requirements rather difficult to understand, they often select a blower that is ill adapted to the job to be accomplished.

67.   Blowers are of two general types, positive displacement and centrifugal.

Positive Displacement Type

68. The positive displacement blower is commonly called the positive pressure blower. This type blower has been very successful on cupolas as it serves both as a source of air and as a meter for the air, delivering a volume of air just under the displacement per revolution multiplied by the speed and against any wind box pressure, provided the motor is large enough to carry the load. An ideal installation for this blower would include a means of variable speed control with V-belt drive. Blowers of this type, when reasonably maintained, render many years of satisfactory service.

Centrifugal or Fan Type

69. The other general type of blower is the centrifugal or fan type blower. This is an excellent type of blower when properly selected, but quite frequently a poor selection is made. Blowers of this type are rated as being able to deliver a definite volume of air at a given temperature and pressure at the discharge point of the blower. For example, a blower of this type might be rated at a capacity of 4400 cu. ft. of air, when the air is 60°F. and the resistance at the discharge point of the blower is 20 oz.

70. There may be several ounces of pressure loss in the blast pipe circuit between the blower and the cupola. If the blower intake is piped to the outside of the building and if a screen or louvre arrangement is placed over the inlet of this pipe more resistance is added in the pipe circuit with the result that a 20 oz. centrifugal blower may only be able to show a 16 or 17 oz. pressure on the wind box of the cupola.

WHAT PRESSURE GAUGE TELLS

71. Earlier in this paper, what the pressure gauge on a cupola attempts to tell us was discussed. Table 7 analyzes the conditions a blower must satisfactorily overcome if it is to be entirely satisfactory.

72. Suppose the normal air required for the cupola is 1000 cu. ft., then the wind box pressure will be about 16 oz. and the line loss will be about 3 oz. making the pressure at the discharge point of the blower 19 oz. If a blower was purchased having a capacity of 1000 cu. ft. against 20 oz. of resistance it would be entirely satisfactory.

73. If an air metering or weighing device was installed in

Table 7
BLOWER DATA

Vol. of Air, Cu. Ft.		Wind Box Pressure oz.	Line Loss oz.	Pressure-Blower Discharge oz.
500	(50%)	4	3	7
750	(75%)	9	3	12
1000	(100%)	16	3	19
1250	(125%)	25	3	28
1500	(150%)	36	3	39

conjunction with the blower, a larger blower might be purchased, supposedly for the purpose of having capacity available should it ever be needed. It is here that a mistake is made, for the blower would usually be purchased with a capacity of 1500 cu. ft. against 20 oz.

74. Provision has been made for an increase in volume, but none for a very sizeable increase in pressure. A glance at Table 7 shows that if cupola operation was modified and 25 per cent more air was desired, the pressure at the blower would increase from 19 to 28 oz. If a 50 per cent increase in air was desired, pressure at the blower discharge would be 39 oz. A centrifugal blower designed to deliver air against 20 oz. of resistance will not be able to deliver air against 28 or 39 oz. In fact, this type of blower will not deliver any air at these pressures if it was originally designed for 20 oz.

75. Therefore, instead of selecting a blower that will deliver 1000 cu. ft. against a resistance of 20 oz., generally it would be better to provide for some future change of practice and select one that will deliver 1250 cu. ft. against a resistance of 29 oz., or, to go still further, one that will deliver 1500 cu. ft. against a resistance of 39 oz. The latest blower installations of this type in the automotive industry are capable of operating up to 48 oz.

76. Table 7 may be corrected for volumes above 1000 c. f. m. without changing the pressure conditions. For example, if the cupola had a normal air requirement of 5000 cu. ft., the wind box pressure would still be 16 oz. and the pressure at the blower discharge 19 oz.

AIR METERING OR WEIGHING DEVICES

77. The writer likes to compare air metering or weighing devices with the speedometer on his automobile. The speedometer

tells him that he is going 60 miles per hour, but his personal
knowledge tells him whether his tires, or the conditions of the high-
way will stand 60 miles an hour. Air metering or weighing devices
will tell how much air is passing the metering or weighing point,
but they will not tell how much air is lost beyond the measuring
point or if the actual amount reaching the cupola bed zone is cor-
rect for the best conditions of operation. However, after satis-
factory operation has been established then the measuring device
will enable one to repeat this performance day in and day out.

SOURCES OF AIR LOSS NOT INDICATED ON METER

78.   There are two sources of serious air loss for which an
air measuring device will not compensate, one is the loss through
joints, blast gates, man hole covers on the wind box, etc., in the
air circuit and the other is the air loss through the slag hole of
the cupola.

79.   If the slag hole is small one day and large the next, or
if the slag hole is burning out quite a bit during a day's run, then
the air which is lost through the slag hole may cause quite a varia-
tion in cupola practice regardless of the expensive instruments
purchased to tell something about the air supply. Slag holes should
be just as small as they can be made, and should be maintained at
this size during a heat.

80.   It is the practice in some foundries to leave the tap hole
open after the wind is put on for some 7 to 10 min. in order to
preheat the refractory. If the tap hole is large enough, enough air
may be lost here during the first few minutes of the run to adversely
affect the possibilities of attaining good melting conditions. During
this time, the coke bed height is adjusting itself and after the tap
hole is botted up, more air is pushed up through the bed zone so
that a second adjustment is necessary. This error is much larger
on small cupolas than on large cupolas.

81.   A cupola blast gate should be air tight. It is better to do
without a blast gate than to install a cheap affair that will allow
a considerable loss of air from the line to the cupola room.

CONTROL OF MOISTURE IN CUPOLA BLAST

82.   The foundry industry at the present time is giving
thought to conditioning the cupola air supply by preheating and
by controlling the water content of the air.

83. At least two foundries have been controlling the water content of the air for several years. Both foundries produce piston rings and report very satisfactory results with this equipment. It is generally supposed that water in the cupola blast tends to increase the stability of iron carbide so that when small castings are made, such as individually cast piston rings, a high percentage of castings are rejected in the machine shop due to unmachinable hard spots. Some of the automotive foundries make an effort to compensate for various amounts of water in the cupola blast by using additional coke. One such foundry uses a 4000 lb. iron charge varying the coke charge between 450 and 500 lb. depending on the water content of the air.

84. A cupola melting 200 tons of iron per day would use about 5,000,000 cu. ft. of air. The water content of air will vary between 2 and 12 grains per cu. ft. from a very dry air to a very wet air. There are 7000 grains to the pound so that on a 200 ton melt, the water blown into the cupola may be as low as 1500 lb. or as much as 9000 lb. The writer feels that any piece of equipment that will eliminate a variable from cupola practice stands an excellent chance to make good, although he is not prepared to offer further data at this time.

PREHEATED AIR

85. Undoubtedly, the principal of supplying preheated air to the cupola as against cold air is desirable. The tuyere zone of a cupola is really a cooling zone and the substitution of hot air for cold air lessens the cooling effect on both metal and slag so that it is possible to tap hotter metal and run for longer hours when using preheated air. Metallurgical melting losses of silicon and manganese are cut in half when using preheated air. Preheated air systems cost from $10,000 to $30,000 per cupola, installed, varying with the type of preheater and being more on the larger cupolas.

TEMPERATURE OF IRON

86. When iron is tapped from the cupola, the temperature may be in the range of 2700 to 2850°F. but when poured into the mold, the temperature may be in the range of 2350 to 2650°F. There is a large loss of metal temperature from tapping to pouring into the mold, seldom less than 250°F. and sometimes as much as 400°F. This sacrifice of metal temperature is due largely to the size, types and number of ladles used in the metal handling circuit. In that the quality of the product is in many cases dependent on

the proper pouring temperature, the metallurgist often misses an opportunity to make a casting to very high specifications by pouring too cold.

87. The desirable pouring temperature for castings for high physical strengths of, say 60,000 lb. per sq. in. tensile strength, or more, is above 2650°F. Lower pouring temperatures will result in lower physical properties. It is possible to get strengths as high as 75,000 lb. per sq. in. and as low as 50,000 lb. per sq. in. from the same metal poured from the same ladle but at different temperatures. The standard objection to hot pouring from the foundry is that "the molding sand will not stand that temperature" while the standard answer of the metallurgical department is "get a sand that will."

88. Until recently, the gray iron foundryman limited his metallurgical to thinking largely in terms of silicon but today they do their heavy thinking in terms of carbon control. The more they learn how to control carbon, the wider the range in which silicon may be used.

89. At the present time, cupola carbon control is thought of as something between 2.50 and 3.50 per cent carbon, experiencing difficulty in getting carbons higher than 3.50 per cent or lower than 2.50 per cent.

90. With a 60 per cent steel mix in the cupola charge, it should be possible to get carbons down under 2.50 per cent, yet with this amount of steel, the range may vary from 2.5 to 3.5 per cent. The most important item in carbon control is the type of coke. A coke readily soluble in molten iron will give a product from 0.50 to 0.75 per cent higher in carbon than when identical practice is used but with a coke less readily soluble in iron. Hard, dense, high ash cokes are less soluble than soft, open, low ash cokes even though both types may work very well as far as producing satisfactory melting rates and metal temperatures are concerned. If metal is not allowed to accumulate in the well, where it is in contact with coke, it probably will be 15 to 20 points lower in carbon than when the metal is allowed to accumulate in the well. When the melting is fast, the carbon will be lower than with slow melting by some 15 or 20 points.

91. In order to control cupola carbons, it is necessary, first, to have a suitable charge, next, to have a suitable coke, then to have a suitable melting technique, and finally to have a suitable tapping technique.

DISCUSSION

Presiding: HORACE DEANE, Deere and Co., Moline, Ill.

NOTE: As this paper was presented before a shop operation course session, no record was kept of the oral discussion.

E. F. GREEN[1] *(Written discussion)*: Mr. Reese is to be congratulated upon the lucid presentation of his ideas upon cupola operation. His remarks concerning coke ratios and control of air volume are particularly pertinent. I would like to advance the opinion that the reason a moisture laden air blast increases carbide stability is because of absorption of hydrogen by the molten iron.

It is realized that time and space considerations probably prevented Mr. Reese from enlarging upon certain details of cupola operation such as slag control. In recent years the steel industry has made tremendous progress in improving the quality and uniformity of steel by means of slag control. I believe the gray iron foundrymen would do well to use the steel industry's knowledge of slag control. We would like to see Mr. Reese compare the use of limestone, and dolomite as cupola slag making agents? In making high strength iron, is it advisable not to remove burnt on sand from return scrap? What is the most advantageous slag makeup to use in melting high-strength iron?

DONALD J. REESE: As Mr. Green implies, I wrote the paper to fit a technical session of 45 minutes duration and, therefore, could only discuss the high spots of cupola operation.

The water content of air varies normally between 2 and 12 grains per cubic foot of air, or with a cupola melting 200 tons of iron per day the daily amount of water blown into the cupola may be as low as 1500 pounds or as much as 9,000 pounds. I feel that the variation in the amount of hydrogen present from the dissociation of water acts to stabilize cementite in varying degree. The result is that light section castings, when made during a humid season are likely to have hard spots and that very heavy sections are likely to have minute pin holes, due to minute quantities of hydrogen released after the metal has become too viscous to permit the hydrogen to get out of the iron.

In one large foundry using a 4,000 lb. cupola charge, the coke charge varies between 450 and 500 lb., varying with the amount of water in the air. This method of correction probably acts toward higher carbon irons causing the iron to trend from hypo-eutectic to the eutectic composition, which in effect means that the carbide becomes less stable.

There are several foundries using air conditioning equipment to control the water content of air between 3 and 4 grains of water per cubic foot.

Actually we know very little about the value of cupola slags. I like to operate cupolas with a volume of slag equivalent to 90 lb. of slag per ton of iron. Because the tuyere zone of a cupola is very definitely a cooling zone, the slag layer covering metal in the cupola reservoir serves as a heat insulator. I also like to use this slag layer as a filter to clean globules of iron dropping down of non-metallic impurities.

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[1] Axelson Mfg. Co., Los Angeles, Calif.

In practice I use a flux on the coke bed equivalent to the amount used on 3 or 4 intermediate coke charges and on every coke charge an amount equal to 30 to 40 per cent of the weight of the coke charge.

I prefer not to remove sand from sprues, gates, risers and scrap castings. Both iron oxide and manganese oxide act as strong fluxes in the cupola and, unless there is enough silica material present in the cupola charge, the cupola refractory lining has to furnish it. I have seen rotary furnace linings go out in 8 to 10 heats due to manganese oxide and last 200 heats when the manganese oxide was neutralized with silica sand.

Apparently the alumina content of cokes and slags is important in cupola operations. Investigations so far seem to indicate that low total carbon irons are obtained when the alumina content is high and vice versa. This seems to be true of coke and indicates that slags containing appreciable quantities of alumina will help produce low total carbon irons, that is, down to 2.50 per cent. In the future I believe that we will be able to go as low as 2.20 per cent total carbon in cupola melted iron.

The writer likes the dolomitic limestones because they are more fluid than the straight calcium carbonates at the normal cupola temperatures.

I think of high strength iron as a material above 50,000 lb. per sq. in. and preferably above 60,000 lb. per sq. in. To me, the essence of quality is the control of carbon and to do this properly the major factor is the quality, size and amount of coke used on the bed and in each intermediate coke charge. The cokes I have best luck with contain 11 to 12 per cent ash and 32 to 35 per cent alumina in the coke ash. Table 6, gives coke and iron charges for various size cupola and, though these figures are normally about half of what the average foundrymen uses, I consider them maximum for high quality iron production and, whenever possible, I use coke and iron charges of only ⅔ the weights listed in this table.

Foundry and Laboratory Characteristics of Cupola Coke

By Hugh O'Neill[1] and J. G. Pearce[2], England

Summarized for the Transactions of the
American Foundrymen's Association by
J. A. Bowers[3], Birmingham, Ala.

Preface

The original of this paper was presented by the authors, Messrs. O'Neill and Pearce, to the Thirty-Fourth Conference of the Institute of British Foundrymen, held in Derby from June 8 to 11, 1937. Because of the interesting and valuable data presented in this paper, it was recommended that a summary be prepared for presentation to the members of the A.F.A. This task of summarizing was undertaken by Mr. Bowers. The paper, as originally presented, appeared in the Foundry Trade Journal, vol. 57, nos. 1091, 1092 and 1093, July 15, 22 and 29, 1937, pp. 46-50, 66-72 and 87-90 respectively and will appear in the Proceedings of the Institute of British Foundrymen, vol. 30, 1936-37, complete with discussion.

1. This paper summarizes the results of tests on 14 different English foundry cokes and is divided into three sections. The foundry trials were carried out at the locomotive works of the L.M.S. Railway Co. The experimental cupola tests and the laboratory determinations were made under the auspices of the British Cast Iron Research Association. The main object was to endeavor to correlate laboratory tests of cupola cokes with their behaviour in the foundry.

Section I: Foundry Trials

2. These tests were carried out during a normal day's operation in a cupola melting from 8 to 9 tons per hr. for the manufac-

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ture of rail chairs. Conditions as to cupola diameter, iron-to-coke ratio, flux, metal charge, and the air supply were standardized for all tests. The weight of patching material necessary after each heat was debited against each coke. (A cross sectional view of the cupola is shown in Fig. 1).

3. The procedure followed was the same in all tests with reference to the preparation of the bed and was according to that followed in regular production practice. Pressure and volume readings were made at frequent intervals during each heat. Atmospheric conditions were also noted, and metal temperatures were determined by means of an optical pyrometer, care being taken to avoid changing atmospheric conditions in the foundry. The amount of slag was carefully measured for each heat. Test castings for strength determinations were cast at frequent intervals and the per cent of scrap chairs was recorded with a notation that the loss could, of course, have been caused from something other than the iron.

4. Chemical analysis of both metal and slag were determined. Complete slag analyses are shown in Table 8, and the average analysis of the metal for each heat is given at the top of Table 3.
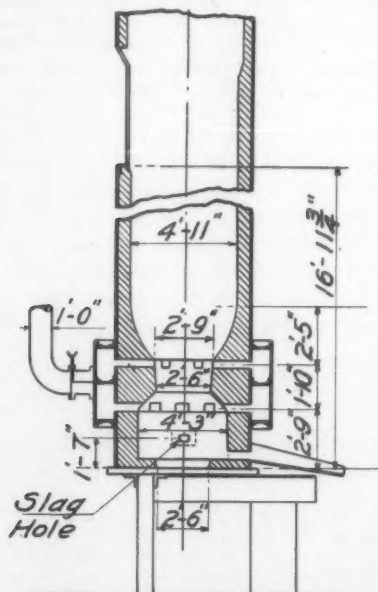
FIG. 1—DIAGRAM OF CUPOLA USED IN LARGE SCALE TESTS

Table 1

CUPOLA COMBUSTION DATA

Coke Reference	Mean Metal Temp., °C.	Melting Rate, tons per hr.	Mean Vol. of Air, cu. ft. per min.	Mean Pressure of Air, ins. water	Blast Temp., °C.	Atmospheric Conditions		
						Humidity, per cent	Barometric Pressure, ins.	Direction of Wind
L1	1,340	8.3	3,675	16.0	14	43	29.70	S.W.
L2	1,320	8.9	3,725	15.5	24	:	30.00	S.
L3	1,320	9.0	....	...	:	:	....	N.E.
G1	1,320	7.4	3,525	16.5	11	56	28.90	S.
G2[a]	1,310	9.3	3,900	15.0	21	:	30.10	S.W.
P	1,315	8.9	3,750	15.5	11	47	30.30	N.W.
A	1,310	8.7	3,750	15.5	14	41	30.30	N.W.
F	1,310	8.0	3,700	16.0	9	55	29.95	N.W.
C	1,305	8.4	3,800	15.0	13	75	30.40	N.W.
I	1,305	8.1	3,675	17.0	9	62	29.70	N.N.E.[b]
H[b]	1,300	8.4	3,800	16.5	10	77	30.50	N.W.
B	1,300	8.9	3,750	16.5	10	41	30.50	N.
E	1,295	8.2	3,675	17.0	11	47	29.45	W.S.W.
M[c]	1,295	7.7	3,600	16.5	8	66	28.60	N.
K	1,290	7.9	3,575	17.9	9	66	29.20	N.W.[b]
D	1,290	7.9	3,650	16.5	10	67	30.05	S.[b]
J1	1,270	8.5	3,750	16.0	8	75	30.20	S.S.E.
J2	1,295	9.0	3,875	15.5	22	:	30.00	W.
316[d]	1,295	8.1	3,800	15.0	10	47	29.90	N.
317[d]	1,290	7.7	3,700	17.0	11	60	29.70	S.W. to W.

[a] In this case the cupola throat was patched to a larger size and so the repeat test is not quite comparable.
[b] Strong wind.
[c] Coke of rather small size.
[d] A bed of coke A was used in these tests.

Table 2

SLAG FORMATION AND ASH CONTENT

Coke Reference	Mean Ash Content, per cent	Patching Material Used per Ton Metal Melted, lb.	Total Slag Produced Cwts.	Slag in Lb. per Ton Metal Melted
L1	6.45	17.2	29.25	64.8
L2	...	13.2	32.75	64.0
G1	8.7	18.8	27.0	63.1
G2	...	...	30.75	60.5
P	8.95	10.6	31.0	64.6
A	6.8	17.7	32.75	68.0
F	4.7	24.6	30.0	65.5
C	8.65	19.1	32.5	65.3
I	5.7	23.9	30.5	67.7
H	3.65	19.6	29.75	62.9
B	8.75	13.4	32.25	66.5
E	7.65	20.1	30.75	67.0
M	8.8	18.0	23.75	56.9
K	9.4	17.0	31.0	70.2
D	5.0	24.0	24.5	54.6
J1	6.45	19.3	38.75	81.4
J2	...	...	33.25	66.3
316	7.15	14.2	31.5	66.5
317	7.0	19.0	33.75	71.4

Discussion of Results

5. These cokes were rated in an order of merit based on spout temperatures as shown in Table 1, which contains other information obtained during these tests. A study of Table 2, which shows the slag formation and patching material required for repairing the cupola, shows that hotter melting cokes do not necessarily require more patching material, and, therefore, the order of merit of these cokes does not have to be changed because of the refractory requirements.

6. Mechanical tests on the test castings show no evidence of inferior metal from any particular coke, as shown on Table 3. This table also shows the per cent of scrap castings with each coke, but, as stated before, the authors call attention to the fact that this scrap figure is not a reliable criterion because of varied conditions in molding and cleaning the rail chairs. The authors state that as the total carbon in the melts rose, the sulphur content fell and vice-versa. Although no figures are given to prove this statement, they also say that the curves for temperature and sulphur content show a marked similarity.

Table 3

QUALITY OF METAL

Av. Comp., per cent at 2 p.m.; G.C., 2.26; CC., 0.76; Si., 1.62; Mn., 0.33; S., 0.12; P., 1.38.

Coke Reference	Transverse test (1.2-in. bars as cast)—Rupture stress on 18-in. centers —tons per sq. in.				Tests on chairs (transverse breaking strength, tons per sq. in.)				Scrap Castings, per cent produced during trial
	First metal	11 a.m.	2 p.m.	4 p.m.	First metal	11 a.m.	2 p.m.	4 p.m.	
L1	19.3	23.9	20.5	20.0	18.0	26.0	24.5	24.5	6.2
L2	20.3	23.4	16.9	18.8	15.5	22.5	21.5	17.0	6.5
G1	23.0*	21.3	23.0	22.5	21.0	19.0	26.0	24.0	4.1
G2	14.4*	25.2	24.3	21.6	15.0	27.0	17.0	25.5	7.5
P	18.1	19.4	22.2	22.6	21.0	22.0	27.0	24.5	3.9
A	21.1	20.7	21.6	22.2	16.0	18.5	25.0	25.5	5.0
F	17.1	21.1	26.1	20.4	21.5	25.0	24.0	26.0	3.8
C	20.5*	24.3	23.1	18.3*	20.0	23.5	23.0	22.75	5.4
I	18.4	21.9	21.4	24.5	15.75	28.5	27.5	23.5*	3.5
H	19.0	22.0	25.4	22.3	15.5	27.5	23.0	23.0	4.0
B	19.2	17.9*	21.0	24.2	22.5	26.0	23.0	24.0	3.9
E	20.2	16.5	20.6	18.6*	22.0	22.0	25.5	22.5	4.2
M	16.4*	22.7	22.6	19.2	16.0	17.0	24.5	22.0*	3.3
K	17.9	21.6	22.9	24.7	15.0	25.0	23.0	19.0	4.2
D	17.9	20.1	24.6	24.6	15.0	22.0	26.0	17.0*	5.6
J1	17.1	20.4	21.8	26.8**	15.0	20.0	14.0	24.0	6.3
J2	17.0*	25.2	21.1	24.4	17.0	21.5	24.0	22.0	5.0
316	19.0*	22.7	21.7	...	15.0	22.0	26.0	18.0*	4.3
317	18.3	23.4	21.2	20.9	19.0	29.0	25.0	24.5	3.3

* Flaw present.
** Broken on 12 in. centers.

SECTION II: EXPERIMENTAL CUPOLA TRIALS

7. In these experiments, a balanced-blast cupola with an internal diameter of 24-in. at the main tuyeres was used. The materials and methods again were standardized although the iron-to-coke ratio was changed from 16 to 1 to 10 to 1. The metal charges were made of 50 per cent steel and 50 per cent pig iron, since the main object of these tests was to study the carbon and sulphur pick-up. The air supply was the same in all tests and any variation in the recorded pressure and volume would be due to the inherent characteristics of each coke. The results of pressure and volume readings were given in Table 4.

8. Although the investigators failed to take metal temperature readings, they state that metal temperatures were higher than those in the first series of tests. No strength test castings were made, but test blocks for chemical analysis were cast from each tap.

Discussion of Results

9. These tests show that almost without exception the sulphur content of the iron is highest on the first ladle and slowly diminishes as the heat progresses. This is shown in Table 5 with the total carbon of the iron.

10. An effort was made to correlate the sulphur pick-up with the combustible sulphur in the coke rather than the total sulphur. Although there would seem to be a closer relationship between these two, in at least one instance this contention did not hold true as shown in Table 6. Speaking broadly, however, the higher the total sulphur and combustible sulphur in a cupola coke, the higher the sulphur pick-up. Fig. 2 shows graphically the sulphur pick-up

Table 4

BLAST PRESSURES AND VOLUMES

Coke Reference	Mean Blast Pressure in Wind Belt, ins. W. G.	Mean Volume of Air, cu. ft. per min.
B	9.65	1,130
D	9.15	1,120
F	9.70	1,135
J	8.55	1,140
L	10.20	1,285
I	9.60	1,245
H	10.40	1,085
316	6.30	1,775
317	9.80	1,180
M	9.30	1,235

Table 5

Carbon and Sulphur Contents of Metal, Per Cent

Coke Reference	Tap 1		Tap 2		Tap 3		Tap 4		Tap 5		Tap 6	
	T.C.	S.	T.C.	S.	T.C.	S.	T.C.	S.	T.C.	S.	T.C.	S.
B	3.03	0.103	3.11	0.085	3.12	0.084	2.93	0.073	2.81	0.070	3.03	0.072
D	3.24	0.112	3.15	0.105	3.06	0.095	3.12	0.091	3.13	0.081	3.09	0.084
F	3.24	0.104	3.29	0.082	3.17	0.075	3.15	0.076	3.19	0.072	3.20	0.067
J	3.31	0.125	3.37	0.114	3.02	0.097	3.05	0.079	3.00	0.079	2.78	0.084
L	3.07	0.083	2.87	0.117	3.14	0.090	3.05	0.076	3.00	0.079	3.07	0.079
I	3.00	0.123	3.11	0.094	3.05	0.082	2.95	0.082	2.97	0.078	2.96	0.075
H	3.21	0.086	3.06	0.082	3.12	0.076	3.04	0.015	3.12	0.067	3.09	0.068
316	3.32	0.122	3.26	0.103	3.06	0.086	3.07	0.083	2.99	0.082	2.94	0.082
317	3.02	0.112	3.13	0.096	2.84	0.080	2.91	0.072	2.97	0.067	2.99	0.069
M	2.90	0.138	3.02	0.115	3.10	0.108	3.08	0.105	2.86	0.099	2.89	0.093
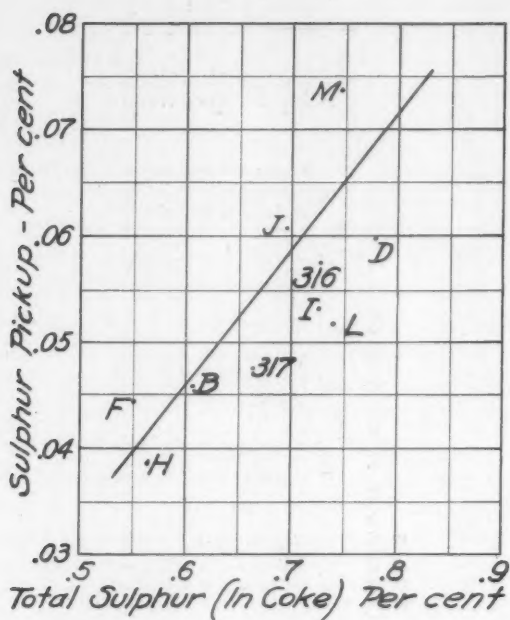
FIG. 2—SULPHUR PICK-UP AND TOTAL SULPHUR

Table 6

CARBON AND SULPHUR PICK-UPS

Coke Ref.	Mean Carbon Pick-up, per cent	Mean Sulphur Pick-up, per cent	Sulphur Per Cent in Coke		
			Total	Combustible	Fixed
B	0.94	0.046	0.59	0.25	0.34
D	1.07	0.060	0.77	0.66	0.11
F	1.15	0.044	0.55	0.51	0.04
J1	1.03	0.061	0.69	0.67	0.02
L1	0.97	0.052	0.73	0.68	0.05
I	0.93	0.053	0.72	0.68	0.04
H	1.03	0.039	0.56	0.55	0.01
316	1.07	0.058	0.72	0.63	0.09
317	0.92	0.048	0.66	0.55	0.11
M	0.87	0.074	0.75	0.65	0.08

of these irons plotted against the total sulphur. Further study of
Table 5 indicates that in most instances the carbon pick-up is higher
on the second tap.

11.   The authors' explanation of this is that the metal did not
reach its maximum temperature until this time. Most of the cokes
tested show a more or less rapid fall in carbon as the heat pro-
gressed. Since none of the cokes gave inferior metallurgical proper-
ties, the authors consider their order of merit based on metal
temperature correct. It is interesting to note that some of the cokes
standing highest in this order of merit are from 30 to 40 per cent
cheaper than some of the more popular brands included in these
tests.

SECTION III: LABORATORY TESTS

12.   The following tests were carried out in the laboratory:

Chemical:	Analysis
	Wet oxidation test
	Combustibility test
	Reactivity test
Physical:	Shatter
	Calorific Value
	Porosity
	Compression Strength
	Abradability
Structural:	Macrostructure (Rose's Method)
	Microstructure

Chemical Tests

13.   *Analysis:* The results of the chemical analysis are given
in Table 7. The most important result from chemical analysis of
coke would probably be the carbon content, but this is of no par-
ticular help in these tests.

14.   *Wet Oxidation:* It has been stated that high graphite is
desirable in a foundry coke and, since graphite oxidizes more
rapidly in this test than amorphous carbon, this test was chosen
as one of the indicators of the quality of a foundry coke.

15.   *Combustibility:* Combustibility might be a slight indicator
since the authors conclude that a decrease in a cupola temperature
corresponds to a decrease in the $CO:CO_2$ ratio.

16.   *Reactivity:* The reactivity test, which is a measure of the
reaction of the coke with carbon dioxide, would seem to be at least

Table 7

CHEMICAL ANALYSES AND CALORIFIC VALUE OF COKES

Coke Reference	Moisture per cent (105° C.)	Sulphur, per cent			Proximate Analysis Calculated to Dry Coke			Gross calorific value B. T. U. per lb.	
		Combustible	Fixed	Total	Mean ash, per cent	Fixed carbon per cent	Vol. Matter, per cent	By Calculation	Determined
L1	0.5	0.68	0.05	0.73	6.4	92.7	0.9	13,350	13,220
L2	0.4	...	...	0.78	6.9	92.5	0.6	13,300	.....
G1	0.9	0.69	0.04	0.73	8.7	90.5	0.8	13,000	12,930
G2	0.6	...	...	0.78	7.4	92.0	0.6	13,200	.....
P	0.6	0.31	0.29	0.60	8.9	89.4	1.7	13,000	12,990
A8	0.7	0.65	0.05	0.70	6.8	92.0	1.2	13,300	13,230
F	0.4	0.51	0.04	0.55	4.7	94.5	0.8	13,600	13,540
C	1.7	0.48	0.24	0.72	8.6	90.4	1.0	13,000	12,790
I	2.6	0.68	0.04	0.72	5.7	93.4	0.9	13,450	13,465
H	1.3	0.55	0.10	0.56	3.6	95.6	0.8	13,750	13,675
B	0.9	0.25	0.34	0.59	8.7	89.5	1.8	13,000	13,020
E	0.5	0.62	0.05	0.67	7.6	91.5	0.9	13,150	13,160
M	0.9	0.65	0.08	0.73	8.8	90.1	1.1	13,000	12,800
K	5.0	0.53	0.05	0.58	9.4	90.1	0.5	12,900	12,820
D	2.1	0.66	0.11	0.77	5.0	94.1	0.9	13,550	13,440
J1	3.6	0.67	0.02	0.69	6.4	92.7	0.9	13,350	13,260
J2	0.3	...	...	0.74	8.9	90.7	0.4	13,000	.....
316	0.7	0.63	0.09	0.72	7.9	90.8	1.3	13,100	13,260
317	1.1	0.55	0.11	0.66	7.0	91.3	1.7	13,250	13,100

an indication of quality. The *"RI value"* is the primary reactivity value of a coke with the first 100 millileters of CO_2 and which has a theoretical maximum value of 200. The *"RIII value"* is obtained by continuing the experiment until the volume of gas obtained by successive determinations is practically constant. The correct method of making use of this test, according to the authors, is to subtract the RIII value from the RI value, and when the remainder is high, it indicates that the cupola value of the coke is high.

Physical Tests

17. *Shatter:* The shatter test seems to be a fairly good indicator as the shatter index tends to decrease as the cupola merit of the coke decreases.

18. *Calorific Value:* The calorific value does not seem to offer much help as an indicator of the quality of a coke.

19. *Abradability:* Although the abradability test is not always reliable, it is an indicator to some degree.

Structural

20. *Macrostructure:* This test is handicapped by the difficulty of securing a representative sample. It does, however, show a tendency of the size and shape of the fissures and according to the merit rating in this paper these should be large and, if possible, rectangular in their formation. A reproduction of the cokes, "L", "H", and "D" is shown in Fig. 3.

21. *Microstructure:* The authors attempted to determine the cell size distribution and the cell space ratio from photomicrographs by actually measuring the intercepts of the black cell walls and the white cell space. They use this measurement as one of the indicators of a good cupola coke. Table 9 gives the results of five laboratory tests separated into three classes; high, medium, and low. It is a significant fact that the cokes of high merit tend to rate high by these five tests.

CONCLUSION

22. The correlation between these cupola and laboratory tests, unfortunately, proves to be rather unsatisfactory. Under the circumstances, none of the laboratory tests can yet be relied upon to predict the practical cupola properties of a coke of unknown origin. The most helpful of all seems to be the shatter test.

23. The following generalizations are hinted at from the
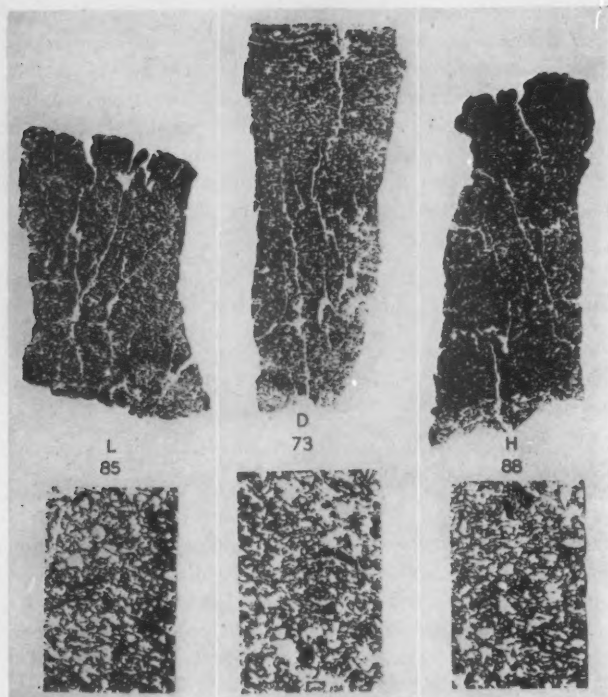
FIG. 3—MACROSTRUCTURES AND MICROSTRUCTURE OF COKES *L*, *D* AND *H*

Table 8

ANALYSIS OF SLAG

Coke Reference	SiO$_2$ %	Al$_2$O$_3$ %	FeO %	Fe %	TiO$_2$ %	MnO %	CaO %	MgO %	P$_2$O$_5$ %	CaS %	Alk. and Undet. %
L1	52.3	9.4	10.9	1.3	2.4	4.7	18.1	0.6	0.2	Tr	0.1
G1	49.5	10.9	12.5	1.2	2.4	4.6	17.3	0.9	0.1	0.2	0.4
P	51.9	7.8	11.8	0.8	2.3	4.5	10.3	5.4	0.7	Tr	4.5
A	50.6	8.1	14.8	1.5	2.6	4.3	17.2	0.7	0.1	0.1	----
F	51.7	0.8	19.9	0.5	2.7	4.7	17.8	0.6	0.7	Tr	0.6
C	48.5	10.1	13.5*	----	1.4	4.5	19.0	1.1	0.4	0.2	1.3
I	50.0	8.2	14.9	1.2	1.7	4.9	18.0	0.7	0.3	0.1	----
H	49.6	7.7	15.0	1.3	2.3	5.1	17.6	0.7	0.3	0.1	0.3
B	48.3	7.6	13.6	1.1	2.8	3.0	19.0	1.3	1.8	0.3	1.2
E	47.9	8.8	14.4	1.5	2.0	4.4	18.0	1.2	0.1	0.2	1.5
M	48.2	11.0	11.5	1.4	2.6	4.3	16.7	0.5	0.6	0.2	3.0
K	47.6	9.5	15.5	1.0	2.9	4.5	16.9	1.3	0.5	0.2	0.1
D	50.0	5.1	18.5	----	2.0	3.6	17.9	0.3	1.7	0.1	0.8
J1	44.7	10.8	25.2	----	2.0	1.3	14.2	0.6	0.8	0.2	0.2
J2	----	----	12.6	----	----	5.0	----	----	----	----	----
316	50.3	8.7	14.8	0.7	1.5	4.4	17.6	0.3	0.4	0.1	0.7
317	49.7	11.7	14.9	1.1	2.5	1.9	16.8	0.3	0.4	0.1	0.1

* Individual FeO contents of slag: (1) 10:45 a. m., 16.4: (2) 11:50 a. m., 14.7;
(3) 2:25 p. m., 6.9: (4) 3:45 p. m., 12.0; and (5) 4:45 p. m., 12.5 percent.

Table 9
H = high
L = low

Coke Reference	Cupola Temp. H < 1,305 deg. C	Shatter Index (2-in.) H < 80	Wet Oxidation Test, H < 187	Openness of Microstructure (per cent black cell wall), H > 60	N.C.R.C. Resistance to Abrasion, H < 60	Reactivity Test, (RI-RIII) H < 30
L	H	H	H	H	H	H
F	H	H	H	H	H	H
A	H	H	H	H	H	H
G	H	H	H	H	H	H
I	M	H	H	H	H	H
H	L	H	H	H	H	L
E	H	H	H	H	H	H
P	H	H	L	H	L	H
C	H	H	L	H	H	L
M	L	L	L	L	L	H
D*	L	L	L	H	H	L
B	L	L	L	L	L	L
K	L	L	L	L	L	L
I	L	L	L	L	L	L

* Beehive coke

results: Cokes giving low metal temperature are low in (1) shatter test, (2) rate of wet oxidation, (3) (RI — RIII) reactivity value, (4) resistance to abrasion, (5) $CO:CO_2$ ratio in the combustibility test; and they are high in density, cell structure, and resistance to static compression.

DISCUSSION

Presiding: FRED J. WALLS, International Nickel Co., Detroit. Mich.

L. L. ANTHES[1]: In this abstract it was said no record was kept of the temperature. But on this experimental work it says the authors had considerably higher than ordinary cupola practice temperatures. The word "considerably" has a large range and I wonder if there is any record at all of what the difference was, because it is in the practical end that we are really interested.

MR. BOWERS: The authors made no temperature readings whatsoever in the experimental cupola tests. They stated the atmosphere around the cupola in which they made these tests was such that pyrometer readings would not have been reliable. So it was just a matter of opinion among those conducting the tests that the temperature was considerably higher, and that is just the way they stated it.

CHAIRMAN WALLS: There is one interesting thing to me and that is the cupola illustrated. It has a 13-in. bosh for a small cupola. That is a little larger than most of the boshes used in this country.

Another significant point was the fact that the price of coke did not measure its quality.

MEMBER: If I remember correctly, there was a difference in carbon picked up in both types of cupolas, experimental and practical cupolas. Would that account for the two different types? The question is, did the balanced blast or single row tuyeres have any difference on the carbon pick-up?

MR. BOWERS: Frankly, I do not recall the figures of the analysis of the foundry trials, but the reason for the difference in carbon pick-up that was mentioned here, was explained by the authors as due to the change in temperature. The metal reached its maximum temperature on the second tap and that was the tap that gave the highest carbons. The experimental cupola heats were run solely to study the carbon pick-up on the same cokes that had been used in the foundry cupola but the character of the metal charges were entirely different and I doubt whether this would be comparable. Furthermore, the iron to coke ratio was different.

J. G. PEARCE[2] *(Written Comment):* Mr. Bowers has stated the position correctly in his reply to Major Anthes, but the actual temperatures obtained from the small cupola, not recorded in the paper

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[1] Anthes Foundry, Ltd., Toronto, Canada.
[2] Director, British Cast Iron Research Association. Birmingham, England.

for the reasons given, were approximately 2750° F. to 2800° F., corrected. Marked discrepancies have been noted between cupola temperatures recorded in Great Britain and in the United States, the latter being invariably higher. Experienced observers, however, believe that the actual temperatures obtained in the two countries are similar for similar types of melt. The large cupola used was working on high phosphoric iron and on account of both this and of the product being made, a very high temperature was not required. The discrepancy noted is possibly due to the fact that optical pyrometer temperatures in England are regarded as comparative only and for the one plant only. They are rarely corrected, and they are not considered as actual nor as a basis for comparison with other plants or furnaces. Furthermore, the higher the temperature, the greater the correction and the greater the margin for possible error.

With regard to carbon pick-up, the large cupola mixture was relatively high in carbon and in the small cupola designedly low in carbon to show a relatively high pick-up. General experience shows that change in carbon is less dependent on type of cupola than on fuel, mixture used, etc., but other things being equal, the higher the temperature, the greater will be the tendency to pick up carbon. The authors are greatly indebted to Mr. Bowers for his able abstract and hope that the paper will increase interest in the important part played by coke in cupola melting.

Importance of Maintenance to Foundry Operations

By W. C. Bliss*, St. Louis, Mo.

1. Foundry methods, patterns, sands and molds have changed very little since the art of molding began but the methods of making these molds, treating the sand before and after the metal is cast and the handling of the castings themselves has developed rapidly with the mechanical age, until the maintenance of the various mechanical devices has become a problem upon which the production and costs depend.

2. It was not many years ago that foundry methods were almost entirely a hand proposition. If a molder broke his shovel, he was given a new one at the store room but the present demand for high speed production has built up problems far more complicated than that.

3. Foundry machinery in our plant, like in many others, must operate under very trying conditions 24 hr. per day, making lubrication a problem. The ever present sand, if permitted to mix with grease or oil, makes an excellent grinding compound and everyone knows the ultimate result. Hence, moving parts must be lubricated and, at the same time, protected to prevent excessive wear and the resulting high cost of maintenance.

Breakdowns Expensive

4. The mechanical handling of sands, molds and castings has created machinery which is complicated, expensive and costly to maintain and the operating heads are being faced constantly with the danger of a breakdown which may tie up the shop and severely cripple production. Insurance companies cover plants for periods when production is stopped by fire, storms or other causes. Breakdowns must be regarded in the same light and money spent

on spare parts, extra belts and alternate machines as insurance against costly delays.

5. In our effort to produce castings quickly and at a *low* cost, perhaps we have gone too "mechanical"—perhaps we have placed too many of our eggs in one basket. There is undoubtedly a happy medium beyond which it is certainly uneconomical to go. It is folly to install machinery to cut costs and then to spend money saved—on maintenance.

6. Fortunately for the foundry, iron and steel are magnetic and the electric magnet plays a prominent part in production. The charge is loaded quickly and cheaply with a magnet which is adequately protected and economical to maintain. Magnetic pulleys play a prominent part too and save many dollars worth of chills, jaggers, fins and spills by removing them from the sand.

7. For those who have installed molding machines and sand systems, what is said will be an old story; for those contemplating or installing a system, perhaps one or two suggestions may be helpful. If so, then the author will feel amply repaid.

REPLACE SPARE PARTS

8. Your jolter, sand slinger or other high speed unit is the heart of the foundry and it is there that the highest efficiency must be maintained. Spare parts, extra fittings, valves, bearings and even a spare unit must be on hand in case of a breakdown. Every minute the unit is down means a loss in the shop. Spare parts, which are used, must be replaced immediately so that they will be on hand. Parts which wear out quickly should be studied and, if possible, replaced with a more permanent piece. The unit must function perfectly—for an inferior mold is worse than no mold at all. Experienced trouble shooters should be on hand at all times and make frequent inspection to detect any wear which later may give trouble. Catch a breakdown before it comes and you will be money ahead.

9. High speed molding machines must be served to make them pay for their installation. Do not install sand slingers, jolters or other high speed units without making provisions to supply the required amount of sand and also a shakeout to take care of the flasks and return them in time to the molding machines.

10. The question arises "Shall we install one sand system for all the units, one for each unit or one for each pair of units?"

Both have their advantages and disadvantages. One system means a major breakdown stops production while with several units, production may continue at a reduced rate, until the necessary repairs are made. During slack times, one of the multiple units may be closed down with the resulting saving in power, wear and tear on the unit. One system must be run whether one molding machine is operating or all are operating. The layout of the shop and the class of work handled also will determine the number of sand systems and units to use.

BELTS

11. Now as to a few details of difficulties we have had with a sand system and what we did to overcome them. The sand must be elevated to supply a molding machine. There are two methods in general use, (1) an elevator or (2) an inclined belt. We have both. For an elevator, get the best quality belt obtainable. It will pay in the long run. The belt has a heavy load to carry and may be subjected to several strains. A broken belt means a costly delay; hours of hard work digging out the sand and restoring the belt to working order. Do not try to economize on belts. For an inclined belt, side aprons were supplied us. Troughing rolls are more desirable. The side aprons must rub the belt to prevent the loss of sand and the point of contact wears the belt. Troughing rolls were installed and give the same results without wearing the belt.

SAND MILLS

12. Sand mills are a very necessary part of the sand system and present an acute problem in maintenance. It is there that the moving parts are constantly in contact with the sand and again proper lubrication is a problem. For a stationary pan and revolving mullers, a rugged and dependable center bearing is desirable. Our mills were supplied with bronze center bearings which lasted 3 weeks. They were replaced by roller bearings and ball bearing thrust bearings which have been changed once in 6 years. The initial cost was higher but think of the repairs saved.

SHAKEOUTS

13. Flask shakeouts are necessary for high speed units. The flasks must be shaken out after the castings have been removed to keep the molding machine gang going. Air or mechanical shakeouts may be used. We feel a mechanical shakeout is more efficient and easier to maintain.

14. The sand from the shakeout is still quite warm and cannot be handled by a rubber belt. This means a slow moving apron conveyor consisting of pans connected by links and a sprocket drive. Shear pins must be provided to prevent excessive strains on the conveyor. The conveyor itself is difficult to lubricate properly and case hardened links have been tried to reduce the cost of maintenance.

15. For a casting shakeout, an air driven hammer seems to be desirable. Castings are suspended in front of the hammer and the castings are picked up and laid down by some form of movable hoist. The sand knocked out of the casting is removed by a belt which passes over a magnetic pulley. The sand may be reconditioned with the help of a vibrating screen and used over again. The hammer and hoist are not costly to maintain and, if proper care is taken, the belt and magnetic pulley will give many months of service.

Core Room

16. The core room may be as mechanical as you care to make it. A sand mill is necessary with perhaps a skip hoist and a car for distributing the sand to the bins. Core racks and core trucks or a crane to handle racks to and from the ovens.

17. The cleaning room consists chiefly of annealing, sandblasting and chipping. An efficient tool room that keeps the chipping hammers in first-class operating order is indispensable. The men will do better work, cut more pieces and be better satisfied with good hammers.

Furnace Insulation

18. For the annealing furnace, insulation will show a material saving. I heard the dean of open hearth men say, "The time is coming when we must insulate to stay in business." His remark applies to annealing furnaces, core ovens and dry sand ovens as well. We have not taken the matter of insulation seriously enough. Insulation will reduce the amount of fuel used and may be used over and over again in case of furnace or oven repair. An insulated furnace will spall less and show a lower repair cost than one that is not insulated. It is well worth the money invested.

BLAST CLEANING

19.   About one year ago, we installed a sand-blast which we believe is new in our district. The blast chamber is a long steel room, open at both ends. The castings to be cleaned are suspended from a continuous chain and move past the operators, who are on the outside of the room. There are four stations, two on each side, which are staggered and the operator looks through the window. The nozzle is at the end of a pipe about 4 ft. long which the operator controls with a hand wheel. Dust collectors are provided for visibility only, the operator working without respirator, goggles or any protective device. We have blown side frames, bolsters and yolks, and with the proper rigging, miscellaneous castings may be blown. We feel this set-up is a decided step toward better working conditions for the sand-blast operator.

20.   The proper illumination of a sand-blast chamber is important. Lights must be placed to the best advantage and still resist the abrasive action of the shot. The protective globes supplied us lasted about one week. Our final solution was a steel box with a heavy glass inserted in the wall of the chamber to the right and left of each operator. Light is supplied by a globe and reflector in the box which can be changed in a very few minutes. We feel this arrangement is the best suited for a sand-blast installation.

21.   The inside of the blast, which is subjected to blasting, will wear out and its protection is a problem which has not been solved satisfactorily. Old pieces of belting, loose chain and even steel plates which can be replaced will cut down the repair bill on the blast.

22.   On a new installation of any kind, do not forget to make provisions for the repair or replacement which is bound to come. Have an I-beam over a motor to be used when the armature is changed and do not put in a concrete wall which has to be chipped out before a part can be removed. A little thought beforehand will save hours of costly delay and keep production up to standard.

23.   When making a layout, keep in mind the safety of your men. Make provisions for the trouble shooter to work and work safely. A costly accident has to be paid for out of profits and means many hours of suffering that might have been avoided.

CONCLUSION

24. In conclusion, there is one more essential we must not fail to mention, and that is an efficient master mechanic. He must be a man who never gets excited, knows what to do when a breakdown occurs, always has a new part ready to replace the broken one and seems to have the faculty of knowing where to look for trouble. The master mechanic's job in a foundry is no bed of roses and, after all is said and done, he plays an important part in the monthly costs of foundry operations.

DISCUSSION

Presiding: E. W. BEECH, Campbell Wyant and Cannon Foundry Company, Muskegon, Mich.

L. E. EVERETT[1], *(Written discussion):* Mr. Bliss has given a very excellent picture of the extent of the mechanization trend in the modern foundry. In some plants maintenance becomes a major cost item, and the importance of proper organization for the work can not be over emphasized. In too many foundry organizations, the maintenance department, particularly in these depression times, becomes a repair department and I want to reiterate and emphasize what Mr. Bliss has already emphasized, "Experienced trouble shooters should be on hand at all times and make frequent inspections to detect wear that may result in trouble later on." "Catch a breakdown before it comes, and you will be money ahead."

Irrespective of the size of the plant there are five main divisions of a foundry maintenance department that must function if the plant continues to operate. These are engineering, operations, inspection, service, and record-keeping, and they are all required in some degree at least in every maintenance department.

Under engineering on all the matters of design, redesign, location, arrangement, installation, and in many cases, specification of equipment, etc.

Under operations, I class the actual labor involved in the construction, destruction, repair and maintenance of buildings and equipment.

Inspection takes care of routine and emergency inspection of equipment, and buildings to see that the operating division functions properly.

The service division takes care of the sanitary, and comfort facilities, yard maintenance, and the services such as electricity, gas, air, water, heating, etc. In many plants they also have charge of operating service departments such as overhead cranes, yard locomotives, etc.

[1] Foundry Superintendent, Key Co., East St. Louis, Ill.

The record branch includes the provision for maintenance reports, space parts files, equipment, orders, etc., and is frequently the most neglected phase of maintenance work. Whether the maintenance department is a one man or one hundred man job all of these five functions must be diligently handled and if any one of them is slighted, delay and excessive cost will be surely reflected. In addition to the maintenance of the mechanized equipment used for producing the foundry's output of castings, we are today forced to concern ourselves with the maintenance of good working conditions, to an extent not dreamt of even ten years ago.

The epidemic of silicosis disability law suits, resulting in our present occupational disease legislation has made foundry operators extremely health conscious.

Ventilation, sanitation, hygiene and safety are problems of operation today. The maintenance of a plentiful supply of clean, dust-free air throughout the breathing zone in the shop is a major responsibility of the maintenance department. The great strides made in recent years by foundry equipment manufacturers in enclosing and exhausting dust creating operations has made it possible to maintain satisfactory air conditions (as regards health hazards) throughout the shop. The development of the wet type dust collectors permits the effective collection of hot and moist dusts, which were very troublesome to handle in a dry type collector.

General heating, adequate day lighting, proper artificial lighting, dust removal and collection, mechanical ventilation, sanitation and hygiene and properly supervised safety procedures are all being given much more attention today than formerly, and are all effective in the maintenance of proper working conditions necessary for optimum productive effort.

JAS. THOMSON[2]: For a number of years we have discussed in one way or another the problem of maintenance. Years ago, that did not seem to be a very important problem to many people, but you who have to deal with cost sheets every month know very well that you get a statement about your maintenance costs for the month and it always looks too high.

This maintenance problem can be attacked in different ways and it is steadily gaining more attention from the management of the different foundries. Some people have not awakened to the value of the proper consideration of maintenance yet, but some of the foundry companies have, and those that have are quite anxious and willing to assist in passing out all the information they can to help the other fellow lick his maintenance problems. This is especially so in Mr. Bliss' paper.

H. L. McKINNON[3]: There is one thing Mr. Bliss neglected to state, and that is that while he bought the equipment for 90 tons an hour, he used it at the rate of 180 tons an hour. That was the reason

[2] Chief Plant Engineer, Continental Roll & Steel Foundry Co., East Chicago, Ind.
[3] C. O. Bartlett & Snow Co., Cleveland, O.

for the change of the vibrator on his belt. Some of you may perhaps have attempted such changes when you expected to get twice as much out of an equipment than for which it was originally designed.

MR. BLISS: I will agree that we have put a terrific load on this system. We have overloaded it and it has done remarkably well. There is an old saying in the foundry that when you have done everything else, then you have to make it "niggerproof," and you cannot make a thing too strong in a foundry. Many of you may have found that out.

MEMBER: I would like to have Mr. Bliss discuss the proposition of using wheels on an apron conveyor every other pitch instead of every pitch. I presume that he has run across that construction in his experience and I believe it would be a good point to bring out at this time, because there is quite a saving in supplying that type of equipment. You can leave out every other set of wheels and, therefore, save some money in the initial cost.

This proposition is basic and fundamental. You have pitches on apron conveyors. Each pitch has a pan attached to it. Now the pans can be attached to each pitch and still the support of these pans is on an axle with rollers at the end of each axle and spaced at every other pitch.

MR. BLISS: These remarks are correct, and not only does it save the cost on the original installation, but it will save on repairs. I believe all are familiar with the usual type of apron conveyor. The pans have ears on the bottom that carry the shaft and there are wheels at the outer end that ride on a track. These pans are in sections, of course. The wheels ride on a rail on each side. The wheels are quite small; naturally, they make many revolutions in their path as they make a good many trips every day. There is a tendency for the sand to leak out of the joints of the apron conveyor. Unless there is a good tight sand seal at that point, the sand gets into the axle bearing, the wheel begins to wabble and unless you watch it closely, it falls off. It is important to get a good tight joint to keep out the sand.

As to the comment that instead of putting a wheel on each axle, put a wheel on every other axle. There is a chain in our conveyor; the chain connects each pan and is also used as a link in the driving chain. In other words, the sprocket comes up and drives through a roller which is on that shaft. In the later type, the drive chain has been divorced from the shaft so that it is possible to change the shaft and wheels without disconnecting the chain. If we take one of our axles out, we take a link out of the chain. It is desirable to be able to change the shafting and put new shafting in or new wheels without taking the chain apart. The chain is kept tight by a take-up. When the chain comes apart, some method has to be used to keep it together while the change is being made.

We thought of several things. As with any new piece of equipment, we had some breaks in the conveyor, and it would separate and let go, separate four or five feet, and it was quite a job getting it

back together again and getting all the parts together. The later type provides that safety chain unit. But that takes you back to your shear pin on your drive wheel; that is the fuse. I might also state, as far as this shear pin is concerned, there is a shear pin on the conveyor in our sand blast, so that if a casting falls off of that conveyor chain and gets fouled, the shear pin lets go and we can find out where the trouble is.

There was one thing that we added to our sand blast after we got it. We put a control button on it at each operator's station. Any one of the four operators can stop the chain conveyor at any time, but when he does stop the conveyor by pushing a button, he also lights a light immediately over his head. That is an indication to the trouble shooter that the chain has been stopped at that point. He immediately goes to that operator and sees what the trouble is.

We are not using continuous tanks. We installed the old-fashioned drop type of tank. You might ask "Why not a continuous tank?" Well, it is a continuous operation for it is just one casting right after another. Now, the operators are going to have to take a little time off occasionally, on the average of about five minutes an hour. And when one operator's tank is down, all four of them refill with shot. In that way, we cut the refilling time down to a minimum.

Risers or Gates for Some Special Non-Ferrous and Alloyed and High Test Iron Castings

By C. Brisbois* and A. E. Cartwright**, Montreal, Canada

Abstract

The authors of this paper have experimented with pouring castings directly through risers and have found the practice successful on many types of castings made from both non-ferrous metals and high test and alloy cast irons. The investigations carried on by the authors developed from a desire to reduce the sensitivity to variables prevalent in ordinary methods of gating. The experiments were conducted on castings weighing from a few ounces to those weighing 2 tons. After much experimenting, the authors adopted as the basic form of feeding head a straight-sided, round-shouldered, short-necked feeding head, the advantages of which are explained. They also describe the various types of sand used in their shop and in which castings are molded to take advantage of direct riser pouring. The use of strainer cores in the pouring risers also is outlined and the dimensions and materials from which they are made is given. The authors then describe in detail some of the various types of castings on which the direct riser pouring method has been used successfully, such as bushings, 85-5-5-5 brass funnel castings, nickel silver castings, pure nickel pump castings, silicon bronze castings. In relation to the use of this method for non-ferrous castings, the authors describe some of the troubles encountered, the functioning of indirect and direct filled heads, types of castings on which this method of pouring cannot be used, and the degree of dross formation caused by direct riser pouring. Under high strength and alloy cast irons the authors explain the method of pouring described in the manufacture of annealing pots, bushings, machine tables, pump impellers, etc. They also discuss the effect of direct riser pouring on mold and core surfaces and end with

* Foundry Superintendent, The Robert Mitchell Co., Ltd.
** Metallurgist, The Robert Mitchell Co., Ltd.

a discussion of the degree of replacement of ordinary methods of gating by the direct riser pour method in their shop.

INTRODUCTION

1.   Foundrymen who are concerned with the varying problems encountered in practically non-repetition jobbing work, in which nearly every pattern entering the foundry represents a gating study of its own, possibly will sympathize with the authors' desire to evolve as fool-proof a method of gating as possible, one which could be applied to the varying designs encountered, and one which will give maximum certainty of obtaining positive results in terms of sound, dense castings. A number of variables are present when planning the proportions of gates and feeding heads to secure adequate results from an unfamiliar design of casting, regardless of the molding method used. If these proportions are not accurately estimated, a faulty casting is obtained.   •

2.   It was with the idea of reducing the sensitivity to variables prevalent in the ordinary methods of gating that experiments were conducted along some less orthodox paths. These experiments were concerned with castings weighing from a few ounces to two tons and made from non-ferrous alloys, including the usual tin bronzes, phosphor-bronzes, manganese and aluminum bronze, nickel silvers, monel, commercially pure nickel, and silicon bronze, as well as high test and alloyed irons.

3.   It would be difficult at this date to conceive of a method of gating and feeding castings that might be described as original or as having no precedent in the published data on the subject. No such originality is claimed for any of the examples described in this paper. However, some of the practices introduced are thought to be less known generally than is deserved by their inherent merit and proved usefulness in the authors' foundry.

SOME ALLOY IRON CONSIDERATIONS

4.   The technique of producing dense castings of the modern high strength and special purpose alloyed irons is much more exacting than it was in the days before these alloys were introduced. High pouring temperatures make necessary the use of more refractory molding sands. High shrinkage and short freezing ranges of the low-carbon, high-strength gray irons, ni-resist, and white irons of the ni-hard type, demand more generous feeding measures.

In many instances, the feeding requirements approach those used by the steel founder.

5.   The increasing popularity of alloyed irons during the past 10 years has been largely the result of the discernment on the part of pioneering producers who have provided castings of these metals of unquestionable soundness; otherwise, regardless of the improved metallurgical qualities and higher properties obtained in test bars, such marked success in competition with other metals and methods of fabrication, could not have been attained or maintained.

6.   After considerable experience by the authors in production of these alloys, close study of the defects most commonly encountered and service reports received on the behaviour of the castings, it was realized that the degree of uniformity in structure and density was of as great significance as the composition.

General Gating and Feeding Practice Consideration

7.   While innumerable variations of gating and feeding of castings are practiced by different foundries, the almost universal method is to run castings by introducing the molten metal through one or more gates and to take care of shrinkage by use of feeding heads. Adequate feeding is sometimes accomplished by means of the gate itself, but when heavy sections are involved and more than a very moderate amount of feeding is required, separate feeding heads usually are employed. Such heads, more often than not, represent the last portion of the mold to be filled upwards from the casting, and therefore, must be so proportioned as to provide feeding metal of sufficient quantity and fluidity to take care of the shrinkage requirements of the casting sections.

8.   A number of supplementary methods frequently are employed to attain and maintain the required fluidity of the metal in the feeding head, e.g., (a) interrupting pouring through the gate at a point where the mold cavity is filled and transferring pouring directly into the riser, and (b) connecting the down-sprue with the base of the feeding head at a point where the mold cavity is full. Covering the head with powdered charcoal or other compounds furnishing an exothermic reaction, introducing a small amount of phosphor-copper (in the case of some bronzes), the use of feeding rods, etc., are all methods used to increase feeding head efficiency.

9.   The obvious method of dispensing with a separate introductory gate and pouring directly and entirely through the feed-

ing head is seldom seriously considered. To use the last mentioned
method, it is necessary to ensure that the metal entering the mold
is thoroughly free from dross or slag, and also to provide a mold
surface that will resist satisfactorily the impact of falling metal.
This latter condition commonly is considered sufficiently difficult
of fulfillment to discourage the use of that method of introducing
the metal.

10. This paper is concerned, mainly, with methods developed
to satisfactorily meet the requirements of direct pouring of cast-
ings through feeding heads.

EXISTING REFERENCES TO DIRECT POURING METHODS

11. Dwyer[1] illustrates and briefly describes a system of direct
pouring through a strainer-feeding head, attributing the original
idea to an Italian foundry engineer, Brunelli. Ronceray[2] likewise
credits Brunelli with the initial recommendation. However, Ron-
ceray can be particularly identified with the promotion of the idea
through several papers read before European and American
foundrymen. [2,3,4]

12. In a number of illustrations, referring to both bottom
and top pouring, he has shown the possibilities involved in con-
trolling rates of pouring to the point of diminishing the size of,
and even the elimination of feeding heads on sections of significant
size by progressive feeding during pouring. In the authors' opinion,
it is doubtful that either method of pouring could be made feasible
to such a degree, except, possibly, in connection with alloys of very
moderate shrinkage character, used in conjunction with extremely
efficient repitition practice.

13. No data regarding elimination of feeding heads, to our
knowledge, have been published in reference to such alloys as
monel, nickel silver, steels, or the modern high shrinkage alloyed
irons in casting sections of significant proportions. We have not
carried our own studies and practice to this point.

PRELIMINARY STUDIES

14. Initial consideration was given to choice of available
molding sands best suited to withstand expected severe conditions.
Practical trials were made to determine the most efficient basic
form of feeding heads to accentuate the expected advantages of
the method of direct head pouring.

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1 Superior numbers refer to references at the end of this paper.

15. Co-operative efforts of the metallurgist and practical foundryman were, of course, essential at every stage of the experiments. Critical comparative study of fractures and metallurgical structures was a prime means of evolving basic feeding head designs. Laboratory and practical foundry tests of sands and bonding agents were of importance, as also were thermal tests indicating the relation between actual pouring temperatures and the temperature of the metal in the mold at the end of pouring.

16. These data, considered along with freezing ranges of various alloys, indicated the improvements available by lowering the degree of superheating of the molten alloys necessary for optimum results. Intelligent revision of molding technique to suit the conditions, economical planning of existing equipment, and the addition of a minimum of new essentials, perseverance in overcoming various minor obstacles, and, not least, tact in obtaining the co-operation of the conservative molder in efficiently handling the less usual materials and methods, were all matters requiring careful attention.

Advantages Obtainable by Direct Riser Pouring

17. It was increasingly demonstrated that, properly applied, the principle of direct riser pouring was capable of providing the following advantages:

(a) Much enhanced effectiveness of feeding by having the feeding head filled last with hottest metal, and maintenance of feeding head temperature by virtue of the head being covered with a heated strainer core.

(b) A moderate economy in feeding metal and gates, together with a large economy brought about by reduction in the number of defective castings.

(c) Much lessened sensitivity to feeding head proportions for different casting designs and alloys of different shrinkage characteristics; that is, the feeding head is generally of a size to compensate for a range of shrinkage wider than that provided for by the more usual methods.

(d) Generally lower requisite pouring temperatures for all alloys, with corresponding less necessity for excessive superheating of the molten alloy. The last is an especially attractive feature from the metallurgical standpoint where those alloys most susceptible to oxidation or gas absorption beyond their melting point are concerned.

DESIGN OF DIRECT-POUR FEEDING HEAD

18. The basic design of feeding head chosen was that of a straight-sided, round-shouldered, short-necked bottle. The total height and diameter are variable, according to the size of section to be fed and the alloy concerned. Casting designs vary to such an extent, and position and shape of cores are such as to frequently make modifications of this basic design necessary but, as far as feasible, the following principles of proportion are adhered to:

(a) The diameter of the base (neck) of the feeding head adjoining the casting should be not less than one-half the full diameter of the head, and,

FIG. 1—A STORE OF RISER PATTERNS FOR DIRECT POURING METHODS.

(b) The full diameter of the feeding head should be continued downward to a point as near the casting as is consistent with maintaining a sufficient mold strength at the junction.

19. The importance of an efficient feeding head proportion and the general adaptability of a few sizes and shapes were considered sufficient to warrant making and maintaining a store of patterns. The riser pattern decided upon is handed to the molder along with the casting pattern. This eliminates uncertainties due to any misunderstanding on the molder's part as to the required feeding provision. Some of these patterns are made of wood; others are cast hollow in scrap aluminum. Plain cylindrical or square extensions are provided for varying height, and a coreprint is provided for seating the strainer core used in all direct-pour heads. A store of these patterns is illustrated in Fig. 1.

20. Most of our experiments were carried out with dry sand molding, and all the illustrations given are, with one exception, of castings made in dry sand molds. This procedure was followed

because the more expensive rejections concerned this type of work; that is, fairly large castings of relatively heavy sections. It was found desirable to use a synthetic sand for the higher melting point alloys, such as nickel alloys and alloyed irons. The fact that this sand is constantly available, is the main reason that other alloys, perhaps not strictly requiring such a highly refractory sand, happen to have been cast in it. It is quite likely that, for ordinary bronzes, a natural sand of the Albany type would be sufficiently resistant to the pouring conditions. We often use French sand for direct riser pouring. A fairly open sand is desirable (preferably one with a grain fineness number of 80 to 120 and with high permeability, a sand corresponding to A.F.A. coarse No. 3 or fine No. 4). and the mold should be well treated with a graphite wash to obtain good surface quality. In our practice, we lean generally toward sand of high permeability rather than use of a vent wire on a less open sand.

Synthetic Sand

21.   The synthetic sand mixture chosen for direct riser pouring consists of a pure white silica sand bonded with 10 to 12 per cent bentonite. The heap sand gradually has been built up over a period of two years by preparation of a facing sand including about 20 per cent of new silica sand. When no increase in the quantity of heap sand is required, it is usual to prepare a facing sand from 100 per cent heap sand, one per cent bentonite being added.

Cereal Binders

22.   The addition of 2 to 5 per cent cereal binder is made to the facing sand. This combats the tendency towards surface drying and crumbling generally associated with synthetic sands. The cereal binder is very efficient in maintaining plasticity without leaving behind undesirable fine material of bond destroying qualities when it burns out, a difficulty we have sometimes experienced when using, for example, sea coal.

Mechanical Analyses

23.   The sand, as will be seen from the mechanical analyses of Table 1, is very resistant to grain changes and to reduction to fine material. Sand losses are confined to average burning-out of bond and to relatively small amounts of sand adhering to castings going to the cleaning room. The quantity of water used throughout is about 10 per cent. A typical analysis of new silica sand and, for comparison, analyses of facing sand and heap sand at

time of writing are given in Table 1. The analyses of Table 1 are of new silica as received, and of facing and backing sand after bonding material had been washed out.

Table 1

COMPARISON OF MECHANICAL ANALYSES OF NEW SILICA, FACING AND HEAP SANDS

Retained On	New Silica Per Cent	Facing Sand Per Cent	Heap Sand Per Cent
No. 6 mesh	Nil.	Nil.	Nil.
No. 12 mesh	Nil.	Nil.	Nil.
No. 20 mesh	2.7	2.9	2.6
No. 30 mesh	21.5	17.6	14.1
No. 40 mesh	27.6	21.6	19.4
No. 50 mesh	25.6	14.8	16.5
No. 70 mesh	14.1	11.4	13.2
No. 100 mesh	6.1	8.4	10.4
No. 140 mesh	1.4	4.5	3.6
No. 200 mesh	0.3	3.2	3.4
No. 270 mesh	0.2	2.2	2.1
Pan mesh	0.5	3.5	4.3

Working Characteristics

24. The sand mixture is not readily workable for hand ramming of molds. All molds made of this sand are rammed with an air rammer. The sand mixture is rather sticky, and difficulty at first was experienced in withdrawing the pattern, but this was overcome by spraying the pattern with a light fuel oil before the mold was rammed.

25. Notwithstanding the coarse grade of sand used, a very fine surface is obtained on castings when the mold surface is protected by a wash of the following mixture applied by air spray before the molds are dried:

> Plumbago ....15 quarts
> Glutrine .... 1 quart
> Diluted with water to 24° Be'

STRAINER CORES

26. Strainer cores perform the very important function of providing clean metal to the molds and are used in the top of the feeding heads. These cores are made of a strong oil-sand mixture and in standard diameters ranging from 2 to 10 in. They vary in thickness from ½ to 1½ in. as the diameter of the core increases. The number and size of the holes in the strainer cores varies ac-

cording to the job, generally from 4 to 12 holes of ¼ to ½ in. diameter being used. Except when it is required to avoid striking a centrally located core, the gate holes are arranged near the center of the strainer core. Care is exercised in all cases to avoid metal striking directly on the protruding sand section at the riser-casting junction.

27. Comments on some illustrations taken from our routine practice will give a fair idea of the method of operation in our foundry. These will be grouped into sections, first those of our non-ferrous castings and secondly those of high-test and alloyed iron castings.

Section A — Non-Ferrous Casting Practice

28. Fig. 2 shows a variety of monel castings: (a) two bushings in which the feeding head takes the shape of an extension of the full diameter of the bushings; (b) a design that frequently gave shrinkage imperfections when made by ordinary gating methods; (c) two equally satisfactory methods of making a ring.

29. At the left of (c), an internal strainer core is used at the base of the down sprue that runs the metal into the mold via a feeding head. This method is used largely for green sand work requiring significant feeding and retains, to a large degree, the advantages of direct top pouring through the riser.

30. Actual direct riser top pouring in green sand is shown at (d). The shaded square area on the bottom of the right hand casting represents the relative position of an oil sand core placed in the drag to receive the first impact of the falling metal. We have not continued the attempt to apply this method to green sand work.

Monel Bushings

31. Solid and cored monel bushings are shown in Fig. 3. The core box and prints of the cored bushings were not conveniently arranged to allow for an undivided pouring head, as in the case in Fig. 2a. The practicality of providing a sufficiently long core print in the drag is the determining factor in the matter of necessity of dividing the feeding head. When the core cannot be set firmly enough to eliminate the necessity for support in the cope, the divided pouring heads shown are used.

32. It will be noted that there is a rather excessive quantity of metal above the strainer core. This is due partly to the limited
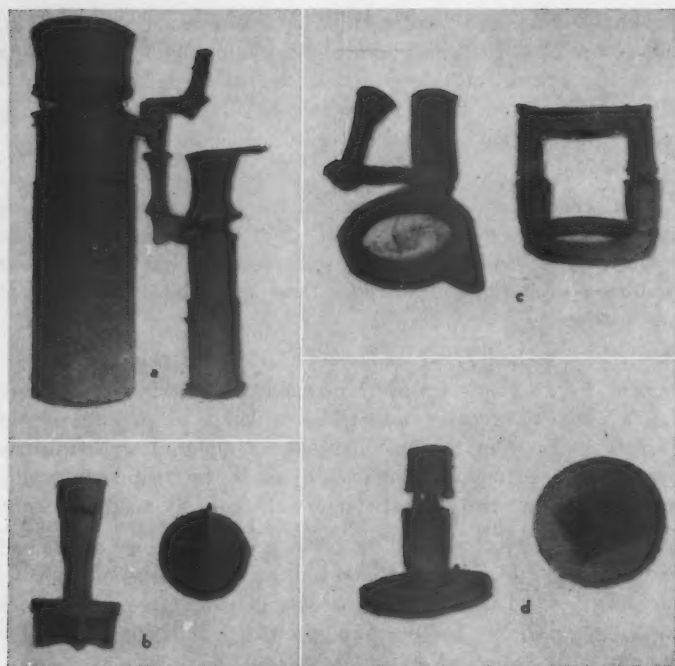
FIG. 2—a—TWO CORED MONEL BUSHINGS IN WHICH CORES ARE HELD VERTICAL BY
EXTENSION IN DRAG. b—A DESIGN WHICH, IN MONEL, FREQUENTLY GIVES SHRINKAGE
TROUBLE WITH ORDINARY GATING METHODS. c—LEFT: MONEL RING, BOTTOM GATED
THROUGH RISER, USING BOTTOM STRAINER CORE. RIGHT: SAME RING BY TOP POURING
RISER METHOD. d—MONEL DISK, 1-IN. THICK, 9-IN. DIAMETER, TOP POURED IN GREEN
SAND. SQUARE SHADED AREA CORRESPONDS WITH POSITION OF AN OIL SAND CORE PLACED
IN DRAG.
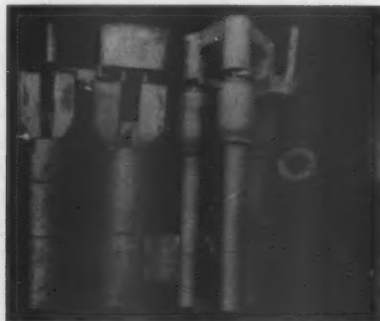
FIG. 3—SOLID AND CORED BUSHINGS IN MONEL SHOWING SINGLE AND DIVIDED FEEDING HEADS.

supply of flask sections of different sizes. Since these photographs
were taken, more attention has been paid to gauging the rising
metal and to stopping of pouring in time to prevent unnecessary
excesses of this kind. It is interesting to note, however, that with
such bushings, the riser below the strainer was fed solid by the
pouring basin excess *through* the $\frac{3}{8}$-in. pop gates used. This indi-
cates the feeding efficiency of the method, despite the interference
of the strainer core.

Large Bushings

33.   Bushings of larger size usually are made with four sep-
arate pouring risers and a circular pouring basin, as exemplified
in Fig. 4, which represents an 80-10-10 bronze casting. For addi-
tional certainty or with alloys of greater shrinkage character, an
extension of the bushing in height, or height and diameter, in
addition to the feeding heads, may be used, as instanced in the
large monel bushings shown in Fig. 5. The extra metal required by
such an extension, plus that in the pouring risers and basin (not
shown), may be regarded by some as extravagant. We do not
consider it such, in view of the certainty of results obtained, and,
on the other hand, the expense of disposing of sizable faulty cast-
ings of such an alloy as monel.

Fig. 4—80-10-10 Bronze, 300 Lb. Bushing with Divided Feeding Heads and Circular
Pouring Basin.
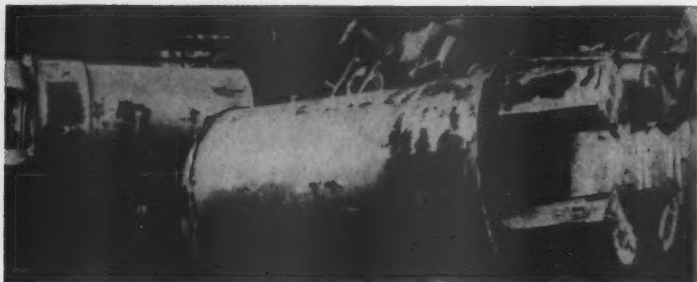
FIG. 5—MONEL BUSHINGS, WEIGHING 700 LB. EACH, SHOWING ADDITIONAL EXTENSION OF
SLIGHTLY LARGER SECTION BETWEEN BUSHING PROPER AND FEEDING HEADS.

85-5-5-5 Brass

34. Very definite reduction in number of rejected castings
was obtained in the case of the funnels of 85-5-5-5 brass shown in
Fig. 6. The funnel section is $\frac{1}{4}$-in. thick up to the junction of the
side tube, which is cast solid.

35. Leakage was responsible for a large number of failures
when more conventional methods were used for gating. Notwith-
standing the many variations of gating tried and the utmost care
in melting, leakage due to ineffective feeding or occluded gases
occurred until direct riser pouring was tried. Since that method
was adopted, rejections have been negligible.

Nickel Silver Casting.

36. The casting shown in Fig. 7 is of federal specification
nickel silver and was not obtained satisfactorily when gated by
other methods involving separate indirect-filled feeding heads. The
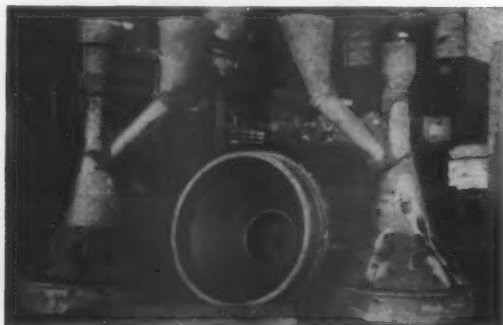
FIG. 6—FUNNEL CASTINGS OF 85-5-5-5 RED BRASS, TOP POURED WITH INDIRECT-FILLED
HEAD ON LARGE SOLID SIDE TUBE.

net weight of the casting was about 260 lb. and the top section containing the cored holes was 3½-in. thick and solid, except for these cores. The pouring-feeding riser was 18-in. high, 8-in. diameter, and the strainer core contained eleven, ⅜-in. holes.

37. *Pouring Temperature:* At the time this casting was made there was some discussion as to the requisite pouring temperature. An authoritative recommendation was 2400° to 2500° F.
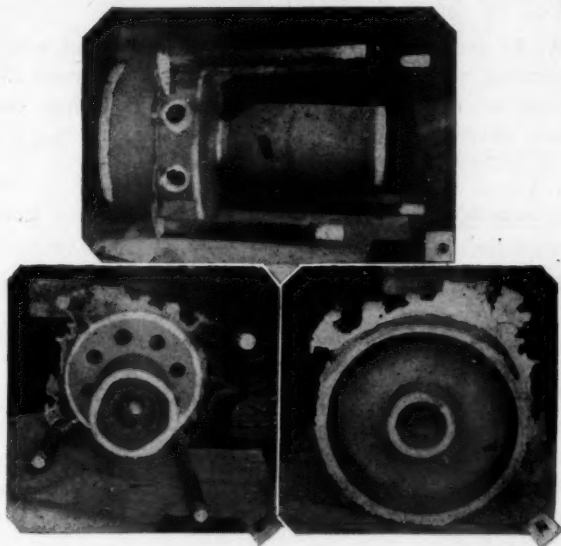
FIG. 7—*a, b, c,*: VARIOUS VIEWS OF NICKEL SILVER CASTING, TOP POURED THROUGH CENTRAL FEEDING HEAD (UNMODIFIED BOTTLE DESIGN OF FEEDING HEAD).

38. Such a pouring range, while quite possibly including the optimum pouring temperature for a gating arrangement of exclusively indirect-filled feeding-heads, was, in our experience, considerably too high for the single direct-poured riser method. With the latter method, the smaller marginal difference between the temperature of the alloy in the ladle and that in the filled mold is a definite factor, and may be one of the major advantages. It also may provide a distinct source of trouble when ignored and a too high pouring temperature used, such trouble generally taking the shape of porosity, large weak crystal zones, segregation, etc., especially in a complex alloy such as the federal specification nickel-silver. This was practically demonstrated by pouring one of

these castings at 2450°F. through the feeding head. This experience resulted in a casting that had serious defects due to porosity and uneven fracture.

39.   The pouring range favored was 2250° to 2300°F. The satisfactory casting shown in Fig. 7 was poured at 2300°F. If we had to repeat this casting, we should be inclined, if anything, towards the lower pouring temperature, but certainly not exceeding 2300°F.

40.   In general, our experience has shown that with direct riser pouring, the required degree of superheat beyond the melting point for most alloys lies between 7 and 12 per cent; for the same alloys with entirely indirectly filled feeding heads, a range of 10 to 20 per cent would be usual.

INDIRECTLY FILLED FEEDING HEADS AND DIRECT POUR FEEDING HEADS

41.   Caution in another respect is necessary when using indirectly-filled feeding heads in conjunction with direct-pour feeding heads. This practice should be avoided, except in certain instances. For example, if the four vent-risers shown in Fig. 7 are replaced by feeding risers of normal size, they would be last, and indirectly, filled. In the present case, the risers, which are of 1-in. diameter, communicate with the casting by a high narrow connection about ¼-in. wide, which connection freezes quickly after it has functioned as a vent.

42.   The size and temperature of the central pouring riser are such that it will maintain fluidity in the casting for a longer period than that of any normal size head which is indirectly filled. The latter will freeze without feeding the casting. As the casting freezes, it will feed the head to a degree depending on its own condition and in its turn will draw on the pouring head almost certainly beyond its available feeding capacity. The result will be unsoundness of very variable location.

43.   It is decidedly preferable to make the pouring head or heads of adequate size to feed the entire casting. When a thin section intervenes and definitely restricts fluid communication between thick sections, all of which may not be separately filled, then indirectly filled risers of suitable size may be used. Even then, chills might find more advantageous use in many instances.

Fig. 8—Pure Nickel Pump Casing Made by Top Pouring Through Feeding Heads

Pure Nickel Pump Casing

44.   Fig. 8 shows an example of direct riser pouring applied to a commercially-pure nickel pump casing required to be leak proof and sufficiently sound of structure to resist strong corrosive liquids. Nickel is one of the more difficult metals to cast satisfactorily because it has a high melting point and very short freezing range. Maximum activity of feeding heads is most important for this metal, and we believe the method shown is most capable of providing it.

Silicon Bronze

45.   Turbine castings of a 4 per cent silicon bronze are shown in Figs. 9 and 10. The casting in Fig. 9 weighs about 300 lb. net, and that in Fig. 10 about 3000 lb. net. The molds for both were made entirely of oil-sand cores of sufficient weakness to collapse as necessary to minimize danger of cracking due to contraction strains.

46.   Both castings were poured directly through strainer cores in the innermost risers. The smaller casting had only vent risers in addition to the pouring risers, but the sections of the larger casting and the distance of the heavy rim from the pouring risers were such that indirectly filled heads on the outer rim were considered desirable.

47.   The pouring heads on the larger casting were of not quite adequate size for complete feeding, leaving some slight shrinkage porosity at the junctures of pouring risers and casting. The indirectly filled heads, in this instance, were not to blame having visibly functioned to effectively feed in their location. The shrinkage

FIG. 9—SILICON BRONZE TURBINE CASTING, 800 LB. NET WEIGHT, FED ENTIRELY BY TWO TOP-POUR FEEDING HEADS.

areas were chipped out and proved to be confined to the small area immediately under the pouring heads. These areas were corrected by burning in, a process quite efficient with the alloy concerned.

48. The surfaces of these two castings were very good and clean, quite free from drossy areas or sand inclusions. One of our early fears had been that this method of pouring would be abnormally severe on the mold surface and, until recently, we have studiously avoided dropping the metal directly onto a core in the mold.

49. The intentional weakness of the cores used for making these silicon bronze castings may be considered as evidence that the method is not excessively severe on mold surfaces. A protective graphite coating was, of course, applied to the core surfaces. We still exercise care, however, to avoid the direct dropping of metal onto thin or delicately shaped mold or core sections.

FIG. 10—SILICON BRONZE TURBINE CASTING, 8000 LB. NET WEIGHT, FED WITH INNER FIVE DIRECT-POUR FEEDING HEADS AND OUTER FIVE INDIRECT-FILLED FEEDING HEADS.

Some Troubles with Single Strainer Feeding Head on a Monel Casting.

50. When marked success is encountered in the use of a particular method, it is not uncommon for the user to attempt to apply that method in a too sweeping fashion. We were not immune to such a mistake, as will be evidenced by the following.

51. An attempt was made to produce the monel casting shown in Figs. 11, 12, 13 and 14 by means of a single strainer feeding head on the hub.

52. The casting weighed about 1500 lb. and was of rather difficult design, as the illustrations indicate. This first effort, the result of which we have no photographic record, resulted in serious cracking of the arms. This was the result of difference in cooling rates.

53. The casting was molded the opposite way up from that shown in Figs. 11, 12, 13 and 14; that is, the hub extension was in the cope and the pouring riser extended from the top of the hub. The mistake was evident on viewing the casting; the central pouring head had been a handicap by maintaining high temperature in the hub, the reverse of actual requirements. Thereby, we were reminded forcibly that the method had its limitations, and was probably the worst possible way of attempting to produce such

FIG. 11—MONEL CENTRIFUGE CASTING SHOWING RESULT OF UNSATISFACTORY GATING METHOD.
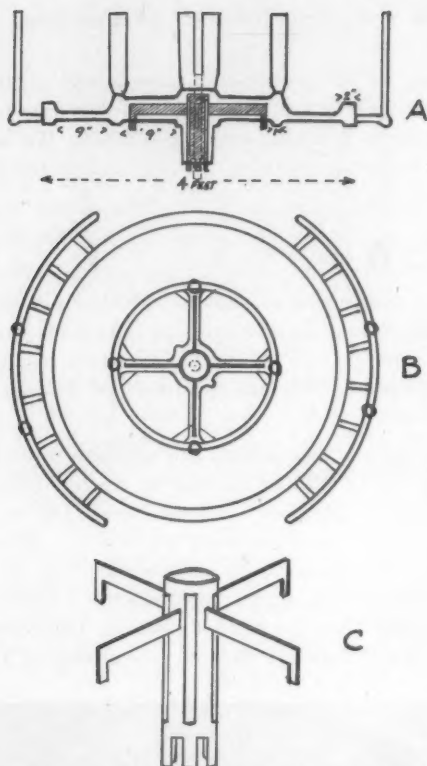
FIG. 12—a AND b—GATING METHOD AND LOCATION OF SKELETON INSERT FOR SATISFACTORY CASTING. c—DESIGN OF INSERT SHOWN IN a AND b.

castings as difficult designs in sprockets, sheaves, etc., whose primary requirement is the equalization of cooling rates and accelerated cooling of heavy hubs.

54. We were recommended to mold the present design as shown in Fig. 11, this time with the hub extension down in the drag and surrounded with chills. Chills also were to be placed under the outside rim. A separate gate entering the hub was to be used, the remains of which are visible in the picture. Between the arms a loose mixture of sawdust, sand, and ashes was rammed and as soon at it was deemed safe after pouring, the cope was lifted and the hub and arms freed. However, the result was very similar to that of the first effort in regard to strain cracks.

FIG. 13—SATISFACTORY CASTING BEFORE CLEANING.

FIG. 14.—REVERSE SIDE OF SATISFACTORY CASTING BEFORE CLEANING. NOTE PROTRUDING
POINTS OF SKELETON INSERT.

55. Following this experience, it was decided to adopt a
rather radical procedure for a third trial. We believe that this
procedure, in view of the result obtained, may be of interest and
assistance to others faced with a similar problem. A skeleton in-
sert of the design shown at *C*, Fig. 12, was cast in monel. The
mold for the main casting was designed with gating· as shown in
A and *B*, Fig. 12. The skeleton, after thorough cleaning by sand
blasting, was placed in the mold as indicated in the diagram, the
extensions on bottom and arms of the insert being firmly bedded
in the main sand mold.

56. The relative section sizes were such as to leave at least
from $3/4$ to 1-in. between the insert and every part of the adjacent
mold wall and core, with the exception of the supporting points.
It was considered that the insert would so function as to acceler-
ate cooling in its locality; that it would either completely melt
into the mass of molten monel or, on the other hand, if it retained
its form,· would act as a reinforcement against possible, though
doubtless lessened, contraction stains in the arms.

57. The casting so produced was apparently perfectly sound
and satisfactory. As will be seen, the only points at which an
examination of the effects of the insert was possible were the sup-
porting prints in the hub and on each arm. Figs. 13 and 14 show
this casting before cleaning, and the protruding ends of the insert
are visible. Each protrusion was examined minutely; chipping,
grinding and filing failed to reveal the outline of the insert below
the surface so treated. Subsequent machining did not uncover
any evidence that complete fusion or welding with the main body
of the metal had not taken place.

DEGREE OF DROSS FORMATION

58. To return to the main theme of pouring castings directly
through a feeding head, the problem of dross formation due to
agitation of the alloy concerned, after it has passed the strainer,
was expected to be a serious handicap with some of the alloys used.
We were agreeably surprised to note how little trouble from this
cause was encountered, particularly with nickel silver containing
up to 10 per cent zinc, and a 4 per cent silicon bronze. The usual
tin bronzes, low in zinc, were free from any drossing trouble.

59. In view of the results obtained with regard to freedom
from drossy surfaces or internally entrapped dross, we determined

to experiment further with more drossy alloys and tried out the method using a manganese bronze containing about 40 per cent zinc and up to 1.5 per cent aluminum.

60. The castings obtained were slightly more drossy on the surface than those poured from the bottom through a reversed horn gate with reservoir (our standard practice), but when the castings were cut up it was seen that they were perfectly sound and free from entrained dross. The experiment with manganese bronze poured by this method was merely an interesting, though logical, sequence. At the present date, we would not make it a standard practice or suggest it as such. We may, however, eventually encounter a situation in which its utilization may be very convenient.

61. The tendency towards dross formation, slight though it is, eliminates the use of a molding technique of this kind on aluminum bronzes. This is most unfortunate in view of the undoubted help it would be toward most efficient feeding of these notoriously high shrinkage and difficult alloys. Whether careful design of pouring basins, coupled with a replacement of the air in the mold by a non-oxidizing atmosphere eventually may permit using this method, can be only a matter for conjecture.

Section B — High Test and Alloyed Iron Practice

62. Figs. 15 to 18 inclusive illustrate the method of production of some forming dies in an alloy iron having a composition of total carbon, 2.90 per cent; silicon, 1.75 per cent; nickel, 2.50 per cent; chromium, 0.50 per cent and showing a tensile strength of 55,000 lb. per sq. in. Fig. 15 shows a closed mold with strainer cores in place; Fig. 16 some drag sections of the same molds; Fig. 17, castings from these molds prior to the removal of the feeding heads; and Fig. 18, a group of these dies ready for shipment. The uniform, close grain obtained, cleanliness over the whole surface, and perfect freedom from external or internal shrinkage, was very notably in advance of similar dies previously made by the use of other gating methods.

Wire-Annealing Pots

63. Fig. 19 shows a wire-annealing pot of a heat resisting iron containing 1.5 per cent nickel and 0.75 per cent chromium. The pot weighs about 700 lb. net, and the design of feeding heads

FIG. 15—MOLDS FOR DIES CLOSED READY FOR POURING, SHOWING STRAINER CORES IN PLACE.

and pouring basin are plainly shown in this illustration. The bottom of the pot is up in the mold, and was shown to be of ideal density and structure when feeding was effected in this manner.

Bushings

64. Incidentally, this type of molding, design of gating, etc. is typical of most large bushings made of alloy irons, but in this
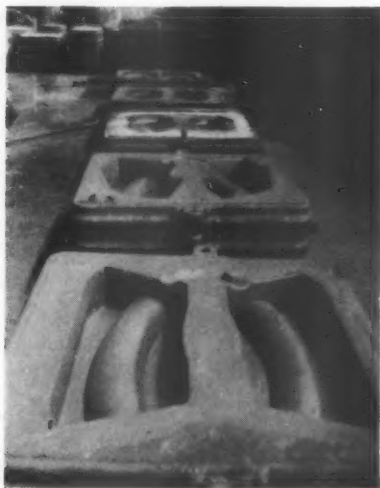
FIG. 16—SOME DRAG SECTIONS OF THE DIE MOLDS.

FIG. 17—DIES BEFORE REMOVAL OF RISERS.

FIG. 18—5 TONS OF DIES READY FOR SHIPMENT.

FIG. 19—COIL WIRE ANNEALING POT SHOWING POURING—FEEDING HEADS AND POURING
BASIN. TYPICAL ALSO OF METHOD FOR LARGE BUSHINGS.

instance the bottom of the pot replaces the open core. Occasionally, the full diameter of a bushing may be continued upwards to act as a feeder but for cast irons, this is limited to those bushings small enough to be readily cut with a power saw. Solid stick bushings are usually extended about 25 per cent of their required net length and top-poured through a strainer core, the feeding excess partly cut off by power saw and then broken.

65.   Some pattern equipment for small bushings is conveniently arranged so that the core can be printed sufficiently, deeply and firmly into the drag, so that the top of the core does not need support. Where this can be done, a solid undivided feeding head is feasible, otherwise, the feeding heads are separated to allow the sand to support the core in the cope.

USE OF INDIRECTLY FILLED RISERS

Machine Tables

66.   A machine table of cast iron is shown in Fig. 20 while Fig. 21 shows in the background the drag of the mold and in the foreground a part of the cope and pouring basin before strainer cores are placed. It was necessary that the flat top of this table should be perfect on machining.

67.   Ordinarily, we should have been inclined to mold this face down in the mold, and the intricate coring shown in the drag would have been a more expensive molding problem. Profiting by our past observation, that this method furnished upmolded surfaces of as perfect cleanliness and density as those molded down in the drag, we used this method very economically from all angles in this case.

68.   This is also one of the very few cases in which ordinary feeding risers were used in addition to the pouring heads. This was due to the rather thin sections intervening between the heavy center and a thick outside section. We were doubtful that the pouring heads would feed the outside walls.

69.   In general, we believe it inadvisable to use indirectly-filled heads in conjunction with the large pouring heads, and prefer to make the pouring heads of adequate size to take care of the entire casting. The reason behind this principle will be appreciated readily when it is considered that the relatively large hot pouring heads are likely to keep the casting fluid longer than an

Fig. 20—Machine Table Made by Direct Pouring Method. Note Indirect-Filled Feeding Heads Used in This Case.
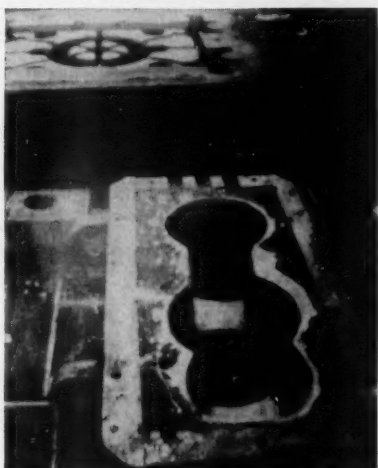
Fig. 21—Mold for Machine Table. Foreground—Cope with Pouring Basin Before Seating Strainer Cores. Background—Drag; Note Heavy Outside Sections Requiring Additional Indirect-Feeding Heads.

Fig. 22—Refrigeration Pump Showing Three Pouring Feeding Heads and Basin. Metal Drops on Intricate Core in This Case.

indirectly-filled head, with the consequence that the latter is fed
solid at the expense of the casting, or feeding metal is drained
from, and beyond the capacity of, the pouring head or heads. This
may result in porosity of very variable location. This was quite
clearly proved in our earlier and less experienced practice when
with less information as to the actual capabilities of the pouring
heads, we retained some of the original indirectly-filled risers pre-
viously found to be reasonably efficient. In the present instance,
where the hot section under the pouring heads is well separated
from the section to be fed by indirect-filled heads, the latter are
permissible and desirable.

EFFECT ON MOLD AND CORE SURFACES

70. Pouring conditions of this method were not so excess-
ively severe on mold surfaces as had been feared. We believe that,
provided the mold surface is such as to resist satisfactorily the
first impact of the falling metal, the ultimate effect on the sand
is no more severe than that due to ordinary methods of pouring.
Care is exercised to avoid dropping the metal on delicate sections
of mold or core and, until recently, we were averse to dropping
the metal directly onto an ordinary oil-sand core.

Refrigerator Pump Castings

71. Misfortune had been our lot in connection with obtain-
ink leak-proof castings of various sizes of a refrigeration pump
when gated and fed by every conceivable ordinary method. With
some trepidation, direct riser pouring was tried, and Fig. 22 shows
one of these pumps so gated. The core used is quite intricate and
good venting is vital. Cores and core wires are required to be
removable from the finished casting. Frequently, pressure tests
would condemn these pumps owing to shrinkage porosity in por-
tions difficulty accessible to feeding. Since conversion to the pour-
ing riser method, we have had no failures due to this or any other
cause at the time of writing.

Sludge Pump Castings

72. Fig. 23 shows a few sludge pump castings made of ni-
hard. For optimum resistance to abrasion, this alloy requires not
only correct metallurgical structure and hardness, but positive
density and clean surfaces free from holes or included sand. In
the course of careful study of used castings of such pumps and
other abrasion resisting castings, we have been impressed by the

Fig. 23—Ni-Hard Pump Parts with Pouring-Feeding Heads.

serious results that may be caused by imperfections in the working surface of such castings. A relatively small sand inclusion often forms the nucleus of a wearing groove that, as it progresses, limits and materially reduces the life of the casting.

73. The same limiting condition frequently is caused by internal shrinkage porosity, to a point at which the hardness and structure of the alloy assume a relatively unimportant place in the value of the casting. The development of the direct pouring riser method has shown splendid results with regard to structural soundness, while the molding and facing materials quoted furnish truly smooth uninterrupted surfaces.

Pump Impellers

74. The ni-resist pump impellers shown in Fig. 24 illustrate two different designs, one suitable for direct riser pouring, the other (the thin one with heavy hub) advisedly not so poured. In the latter instance, which also applies to such castings as difficult designs in sprockets, sheaves, etc., if a direct pouring riser is used on the heavy hub, the maintenance of heat in the riser location will set up strains sufficient to crack the casting.

Fig. 24—Two Pump Impellers in Ni-resist—Left: Bottom Gated on Outside Edge
Right:—Top Poured Through Divided Feeding Heads with Chills Near Base of
Feeding Heads.

FIG. 25—NI-RESIST HEAVY SECTIONS POURED THROUGH FEEDING HEADS FOR MAXIMUM DENSITY.
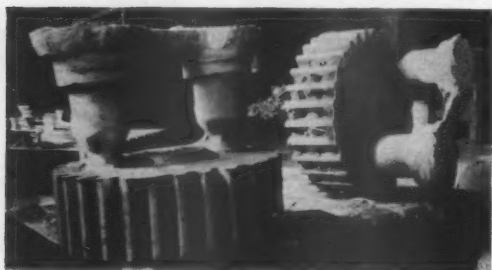
FIG. 26—LARGE NICKEL-CHROMIUM CAST IRON GEARS.
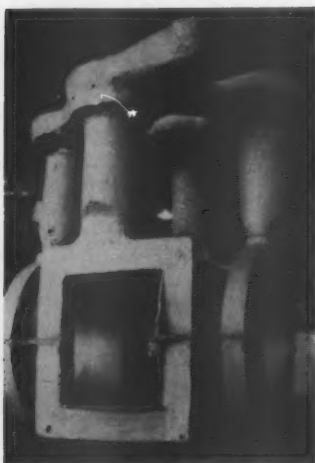
FIG. 27—NICKEL-CHROMIUM CAST IRON VALVE SECTION.

75. The condition here is the reverse of that actually re-
quired for such castings; the cooling of hubs must be accelerated
to equal the cooling of the thin arms or web so far as is possible.
With the small solid impeller, the direct pouring method worked
well, and, in addition, chills were used to hasten solidification.
Fig. 25 shows ni-resist castings, while Figs. 26 and 27 are of a
similar composition as the dies of Fig. 18.

Degree of Replacement of Ordinary Methods

76. The accompanying figures are fairly representative of
the variety of designs to which the direct riser pouring method has
been applied, although they do not convey a quantitative idea of
the degree of conversion to the method.

77. The extent to which direct riser pouring has been intro-
duced into the practice of the authors' foundry by gradual replace-
ment, is large and increasing almost daily. In general, all jobs that
require a feeding riser of significant size are automatically con-
sidered eligible. The exceptions, from the point of view of suitable
alloys, are the yellow brasses, manganese and aluminum bronzes,
and the light aluminum alloys.

78. Probably about 50 per cent of the jobs eligible for this
method of molding are so produced. The remaining 50 per cent
represent those castings which, we are quite satisfied from previous
experience with them, can be soundly produced by the ordinary
methods. It is admitted that the direct pouring method is slightly
more costly and time consuming in molding, so that we do not go
out of our way to make 100 per cent conversion to the direct pour-
ing when no significant benefits are to be expected. It may safely
be stated, however, that the degree of change represented by intro-
duction of this method has been of revolutionary proportion.

79. The previously described practices have been developed
from the point of view and demands of a miscellaneous jobbing
foundry. As applied to the practice of a repetition foundry in which
an exhaustive study of each job generally is possible before produc-
tion is stated, it is certain that improvements could be made in
the matter of precise proportioning of pouring head dimensions,
design of economical pouring basins, and provision of ideal flask
shapes and sizes, to a degree impossible for the jobbing founder.
The latter can afford, however, to be reconciled to a somewhat

lesser degree of economy of this kind in consideration of the very real economic value inherent in a reputation for, and returns from, the production of sound, dense castings.

References

1.   Dwyer, P., *"Gates and Risers for Castings,"* Penton Publishing Co., Cleveland.
2.   Ronceray, E., *"The Making of Castings Without Feeding Heads,"* Proceedings, I.B.F., vol. 15, 1921-22.
3.   Ronceray, E., *"Facts About Sound Castings, Feeding Heads, Runners, and Venting,"* Proceedings, I.B.F., vol. 16, 1922-23.
4.   Ronceray, E., *"Some Considerations in Gating and Pouring Castings,"* Transactions, A.F.A., vol. 38, pp. 41-64, 1930.

DISCUSSION

Presiding: HAROLD J. ROAST, Vice President, Canadian Bronze Co., Ltd., Montreal, P. Q.

A. J. CASSISTA[1] *(Written discussion):* This paper by Messrs. C. Brisbois and A. E. Cartwright on risers and gates will go a long way towards advancing the practice of direct pouring. It has placed the spotlight on its results and advantages. The article gives very good instruction on how to avoid possible trouble that one may encounter while putting this system of gating into practice. Their paper also emphasizes that its application is practical on a wide variety of castings. Their article in my opinion, ranks as one of the most interesting papers I have read on gates and risers.

Previous to reading Mr. Brisbois' and Mr. Cartwright's paper, I have had occasion to make use of direct pouring on a few castings that were not successfully made with ordinary form of gating and feeding. At this time, I should like to have the opportunity of presenting my experience with this method of pouring, not with the idea of shedding new light on direct pouring, but merely as small additional information on what has already been presented.

The outstanding experience was on a small casting, weighing approximately 80 pounds; the design is illustrated in Fig. 28. This casting was a combination of a fan, fly wheel and drive pulley for a small air compressor. On the rim two V-grooves were machined. Porosity in

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[1] Vermont Foundries, Inc., Springfield, Vt.
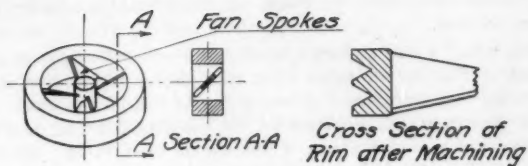
FIG. 28
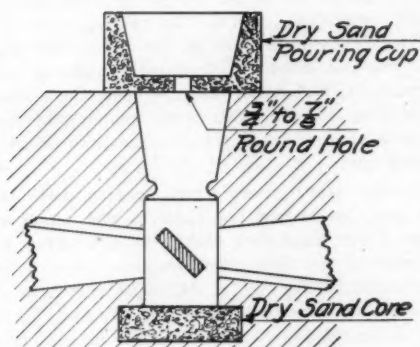
FIG. 29
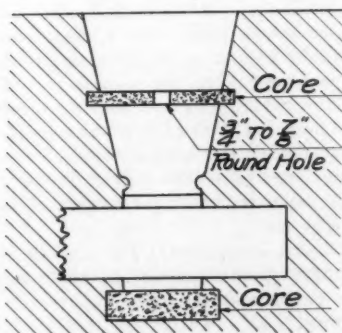
FIG. 30

these grooves was the cause of a high machine shop rejection; also occasional shrink holes were found in the hub. This casting was poured "two-up" with gates and two risers on the rim, also a large riser was placed on the hub.

For a trial, a few castings were made with the gate and riser as illustrated in Fig. 29. A large riser was placed on the hub, and a dry sand pouring cup was placed directly on the top of the riser, after the mold was closed, and the bottom of the pouring cup had a hole; ¾ to ⅞ in. dia. which would direct a stream of iron in the center of the riser. A small dry sand core was rammed on the lower part of the hub, to absorb the drop of the pouring stream. The iron was poured in the cup and after the mold was filled, the pourer would take a small rod and break up the bottom of the cup; by this the effective height of the riser was increased and also the broken part of the cup would float to the top and act as insulation on the top of the riser and would keep it in the molten state somewhat longer. The casting turned out clean and solid. Several hundreds were cast in the same way and the machine shop rejection was cut down to practically nothing and the foundry loss was also materially reduced. In this case, the actual molding cast was reduced, because by using this direct pouring method, two sets of gates and risers were eliminated, which more than offset the cost of the dry sand pouring cup.

In a good many other cases, direct pouring was obtained, without extra molding cost, by having the riser in two parts and in between these two parts a dry sand core slab with a pouring hole which was placed in the center of the riser, was rammed. Construction of this type of gate is illustrated in Fig. 30. A dry sand core was rammed in the mold to come directly beneath the pouring stream. This type of mold was poured by using the top part of the riser as a pouring basin. And here also the core in the riser was broken up after the mold had been filled for the same reasons as mentioned in the previous case. In every instance, excellent results have been obtained with the use of direct pouring.

MR. CARTWRIGHT: Mr. Cassista's suggestions and his experience suggest or confirm what I had in mind, that other people have tried the same idea. And there are so many simple improvements that might be made that we might not happen to think of, and that method of putting the strainer core part way down the riser is one of them. Of course, that can only be done where you are pouring directly into one riser, as far as I can see now, but it has the advantage of lessening the drop of the metal and severity on the sand surface, of course. We would not care to use the one large gate hole there. We would prefer to have a number of smaller holes to lessen the danger of forming a vortex to which the slag might go.

T. J. WOOD[2] (Written discussion): The authors of this praiseworthy contribution to foundry practice are to be commended on the general excellence of their paper and the clear manner in which they have recorded their studies.

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[2] Foundry Engineer, Robins Conveying Belt Co., Passaic, N. J.

The principles of direct-riser pouring, as the authors have shown, entail many advantages difficult to attain through other methods. In our own foundry (in which we have a fairly even division between machine and bench or floor molding) we employ the direct-riser practice on certain classes of work and have obtained excellent results. With respect to repetition work, we have used a slightly modified form of the direct-riser method which may be of general interest. For example, one problem consisted in producing several thousand Ni-Resist bracket castings of the design shown in Fig. 31. Since there were no cores, it will be appreciated that each casting required a substantial reservoir of hot metal for feeding. By pouring through a skim-gate arrangement down through the sprue, the flow of metal was gated directly into the blind risers and from there into the castings. Using a sand of high permeability, it was unnecessary to provide for any additional venting. The slight taper which may be noted on the blind

PLAN VIEW

BLIND
RISER

GATE

SPRUE

GATE

BLIND
RISER

END VIEW OF
HALF SECTION

BLIND
RISER

GATE
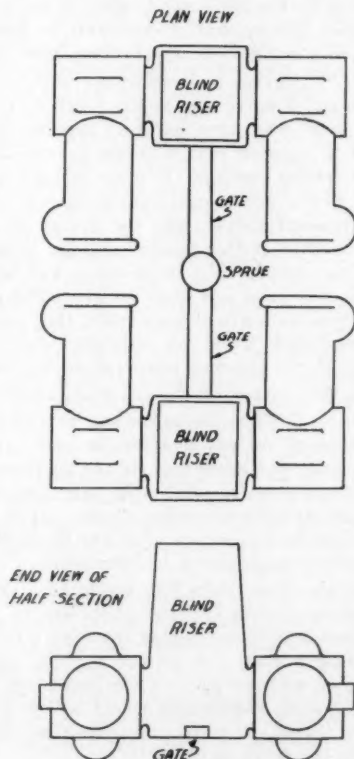
FIG. 31

risers was necessary to ensure a clean draw of the riser patterns, which were, of course, incorporated with the casting patterns on the matchplate. It also may be pointed out that in this type of molding one of the advantages of using blind risers lies in the fact that the thermal action is better than that of the conventional type (of corresponding weight), since advantage can be taken of the insulating effect of the sand. The effectiveness of the method employed is attested by the fact that casting losses, due to all causes, were somewhat less than two per cent.

MR. CARTWRIGHT: That is rather similar to the method we mentioned in the paper whereby we pour a lot of green sand work with an internal strainer core leading through the riser into the casting.

MEMBER: Would Mr. Cartwright give us an idea of how much bentonite was used in the synthetic sand or whether the facing was made with it? Further, was it necessary to dry the mold?

MR. CARTWRIGHT: The entire mold was made with synthetic sand. We prepare a facing of synthetic sand, also; a separate facing, free from lumps, of course. The amount of bentonite we usually use is from 10 per cent to 12 per cent, both in the backing sand and facing sand.

W. M. BALL, JR.[3]: Having had experience with the type of gating mentioned in this paper, I agree that it has much merit. I believe that for the benefit of those who have not had experience with this type of gating it is well to mention that it is not a "cure-all" for all types of bad gating, or defective castings. It must be used with some degree of understanding. There are many non-ferrous alloys that will not lend itself to this type of gating, also the design of the casting will have a bearing on the use of this type of gating. Experience seems to indicate even with the simple alloy of 85-5-5-5, and where the casting has deep well or uneven drag line that we are bothered with foaming and I believe that this gating will accentuate this condition.

MR. CARTWRIGHT: We found no difficulty with tin-bronzes. One might find difficulty if the pouring temperature was too low.

MEMBER: Was the casting of Fig. 19 made in dry sand?

MR. CARTWRIGHT: That is the annealing pot, though it looks like a bushing. The bottom of the pot replaces the core. It is of gray iron but is poured identically the same way as the 80-10-10 bronze bushing of Fig. 4, and was made in a dry sand mold. All these top poured molds are made in dry sand with the exception of one, (d) in Fig. 2, in which we put a dry sand core in the bottom of the mold similar to the method shown in Mr. Cassista's diagram of his discussion.

MEMBER: It occurs to me there still would be a cheaper method of making bronze bushings successfully in either dry or green sand mold by bottom gating and letting the metal come up quickly, as Mr. Ball suggested, and having a solid rim around the top as a solid feeding head, pouring it open, with no coke. Then that solid ring around the top would act as a feeding head which would be cut off in the machine operation. It seems to me the method described would be quite expensive for a casting of the 80-10-10 alloy.

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[3] General Supt., Edna Brass Mfg. Co., Cincinnati, O.

MR. CARTWRIGHT: We have used a similar method to that but there is also some uncertainty as to just what temperature and fluidity you get in that feeding head when it does come up, whereas, there is no doubt but that you have the hottest metal in the top when it is poured by this method. We have not had the trouble Mr. Ball suggested, the frothing. We have had good results by this method and we are sure we are getting good feeding by this method.

MEMBER: I can see where you would get good feeding with such alloys as Monel metal and high nickel alloys, where you would have high shrinkage, but I can't see where you would need it with an alloy of 80-10-10.

MR. CARTWRIGHT: We happened to run into difficulties on that particular bushing and we got our best results by the method we described.

CHAIRMAN ROAST: I would like to add my tribute to the carefulness with which these various trials have been made. I am not acquainted with all of them, but they were made as stated in the paper and they were very satisfactory not only to the foundry but to the customer who used them.

As to the silicon-bronze runner, which happens to be a P. M. G. runner I can tell Mr. Cartwright something that I am sure he has not heard because it was only at the close of last week that I heard that this runner was giving very excellent service. It would seem, therefore, that the burning that took place on that runner, performed with the knowledge of the customer, was entirely satisfactory. I do think that is important, that whenever we do something to repair a casting it should always be done with the knowledge of the customer, no matter how good and certain we may be that the results will be satisfactory. The question of bottom gating, as I say, has been brought up, and my experience coincides with the author's which is that in so many cases your cool metal gets into your risers and is adequate when called upon to fill the void caused by the shrinkage of the casting beneath. We have all had that trouble, I presume, especially in large castings.

Now the frothing proposition. P. M. G. is a metal with which we have been accustomed to take rather good care of in regard to the question of frothing. We used the horn gate, letting the metal in at the bottom, using extreme care. And it was with a good deal of trepidation that I stood by at the making of this 3,000 pound casting and saw the metal go in at the top when I expected to find rather sad results, but very fortunately I did not say so and, the outcome being good, I kept very quiet.

I would like to ask a direct question of Mr. Cartwright and that is, does he put all the metal in the basin and get his temperature or does he put part in, having gotten the temperature in the container from which the metal was poured?

MR. CARTWRIGHT: With regard to Mr. Roast's question, "Do we put the entire metal into the pouring basin"—we do not. We sometimes seem to get into rather extravagant proportions in pouring basins. We design the pouring basin so we can pour rapidly right into the

strainer cores and so choke them thoroughly to prevent dross from getting through. We are always successful in doing that but I would say sometimes we run to extravagances on pouring basins by having too great metal in the pouring basin at the finish. If we have it too small or shallow we are liable not to choke the strainer core, which would be very serious.

MEMBER: I would like to ask Mr. Cartwright if he would show the way he would gate by the direct method a bushing made from 80-10-10, 3½ in. with a 1-in. core?

MR. CARTWRIGHT: *(Note: Fig. 32 is supplied by Mr. Cartwright to represent the sketch he made to answer this question.)* With 80-10-10, apart from one instance I showed, which I explained was the case we got into trouble, we would probably not attempt to pour that, in the first place, by the direct top pouring. We would go to bottom pouring
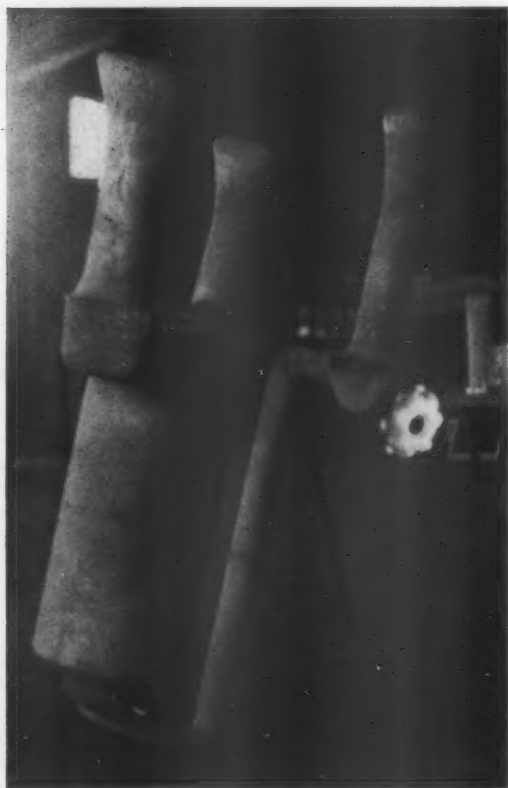
FIG. 32

and use the indirect filled feeding heads, as usual. If our riser was up at the top and our runner at the bottom, we would of course put an internal strainer at the juncture of two in-gates and run our metal into the feeding head as well as into the bottom. We usually put in the internal strainer when we do this because unless the gate entering the bottom of the riser is thoroughly clean, you are liable to get dross in the bushing. We usually put a strainer here, pouring down, so we can run clean metal simultaneously in both directions. But for 80-10-10, we would not, without good reason, direct-pour small bushings.

CHAIRMAN ROAST: Why couldn't that casting of Fig. 32 be poured with the direct method? I do not see any reason why it could not.

MR. CARTWRIGHT: I do not say it couldn't be so poured, but we wouldn't consider it necessary with 80-10-10 bronze because it is quite a moderate shrinkage alloy as a rule, as the gentleman over there suggested.

D. FRANK O'CONNER[4]: Has the author any experience pouring the 85-5-5-5 alloy through strainer feeding heads into castings such as large gate or globe valves where the core lies horizontal and directly in the path of the metal and if so, would the metal impinge the core, splatter throughout the mold creating oxides which might show on the surface of the casting?

Has the author had any experience pouring 85-5-5-5 through strainer feeding heads in green sand molds?

MR. CARTWRIGHT: We have not up until now tried very hard to make use of this for green sand work although we are beginning to do so in a tentative way. We have installed a sand slinger and by ramming the sand very hard and skin-drying, at the present time we are running some castings that way. But, we do not normally use this method for 85-5-5-5 or other alloys in green sand.

As far as small valves are concerned, I cannot say as we do not make valves. But on cored sections, we were, of course, very much afraid in the beginning to drop metal on delicate cores. The silicon bronze castings were made with exceedingly weak cores, about 30 to 1 sand to oil ratio, and with sawdust in the core as well to help to collapse it if necessary, so we do not think it is exceptionally severe dropping on the cores. One of the worst instances, I think, was shown in the cast iron section, Fig. 22, where the iron drops on the rather elaborate core of the refrigerator pump. You cannot see the core, of course, but it is a rather elaborate core and we had no trouble with dropping the metal on the cores.

MEMBER: Does Mr. Cartwright from his experience believe it is possible to get shrinkage from a head that is placed above the strainer core or if, when a head is necessary, must it be placed under the strainer?

MR. CARTWRIGHT: With a head placed above the strainer core, with the small holes connecting through and the interference of the strainer core, you do not get very much feeding benefit from metal

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[4] Walworth Co., South Boston, Mass.

above the strainer core. But as was mentioned in the paper in the case of Monel, where we ran into excess metal in the pouring basin, the feeding heads themselves were entirely solid. We cut them up and there was absolutely no pipe in them at all. They fed through the small, quarter-inch diameter holes, but we would not count on feeding through the strainer core. It would be much better, as Mr. Cassista mentioned, to break the strainer core.

C. C. BRISBOIS AND A. E. CARTWRIGHT *(Written Closure to Discussion)*: The authors desire to thank all those who contributed to the discussion of their paper, and particularly the chairman of the session, Harold J. Roast, for his comments and endorsement based on first hand observation of the practices described. The contributions to the discussion furnished in written form by Messrs. Cassista and Wood are much appreciated and are definitely additive to the data of the paper.

The tendency for the discussion at the non-ferrous session to focus upon these methods as applied to alloys of normally such moderate shrinkage and foundry difficulty as 80-10-10 and 85-5-5-5 bronzes was due, we believe, mainly to the proportion of the illustrations in the paper. These, while chosen to exemplify the range of alloys so handled probably left a false impression of the relative utilization of direct-pour methods for these bronzes in comparison with alloys of greater shrinkage and difficulty.

Generally, it is neither economical or sufficiently advantageous to direct-pour these bronzes. However, where ordinary methods are giving trouble, or designs are difficult from a feeding point of view, recourse to it may be profitable. It may also be considered that it is often possible to obtain, by the direct-pour method, greater density due to more effective feeding of a section which, gated and fed by ordinary methods, may be of a minimum density though satisfactory to visual inspection.

With regard to gray iron practice, the direct-pour method used in connection with a moderate shrinkage soft gray iron may be definitely unsuitable. This was illustrated during the cast iron shop course meeting by a casting so made. In that instance, the tendency of the large single direct-pour head to maintain high temperature in its locality produced an excessive precipitation of graphite to a point where normal shrinkage was overwhelmed by the expansion caused by kish graphite formation. The cooler condition provided in the hub of a duplicate casting made by conventional gating was sufficient to suppress the tendency toward an excessive precipitation of graphite with the result that the moderate shrinkage was unhampered.

The title and illustrations referring to the section of the paper dealing with iron did not tend to suggest that these methods were useful for soft gray irons of moderate shrinkage, but the authors are grateful, nevertheless, for the examples produced in the meeting that served to adequately emphasize a point which was only inferred by the paper.

A Study of Steel Molding Sands

By H. W. Dietert[1], E. E. Woodliff[2], and J. A. Schuch[3],

Detroit, Michigan

Summary

The authors collected and tested 33 steel molding sand samples used in various steel foundries throughout the United States. From data secured through a questionnaire submitted with each sand sample, it was possible to correlate certain foundry conditions with sand test figures. The average sand test figures for many sand properties for both green and dry sand molding are shown. These averages may be put to practical use when studying sand conditions in any individual steel foundry. In correlating casting defects and sintering temperatures reported, it was found that sand samples with high sintering temperatures caused the least casting defects. The expansion and contraction of the steel sands were determined. From the data collected, the authors draw the conclusions that a steel molding sand should have low expansion, preferably show no contraction, and have a sintering temperature above 2600° F.

1. The correlation of sand test figures derived from a study of sands used by a large group of foundries producing a similar type of metal is known to be of practical value. Any information so obtained allows conclusions to be drawn which have the two-fold purpose of being educational and of establishing limits by which sand control can be made both practical and profitable.

Object of the Study

2. Of all the metals that are cast, it has long been known that steel inflicts the most severe punishment on the molding sand. The determination of the properties which steel molding sand must have, is best made by correlating test figures of sands used in a large group of steel foundries. For that purpose, this investigation was undertaken and samples were secured of sand used for both dry sand and green sand molding in 17 foundries.

3. A total of 33 sand samples were obtained from the 17 steel foundries. With each sand sample, information was given as to the behavior of the sand, nature of the metal, casting weight, molding methods, and other foundry conditions. All sand samples

were received in sealed containers and were conditioned as used in the foundry. The questionnaire requesting information for each sand sample submitted is shown as Fig. 1.

METHOD OF TESTING

4. The equipment and procedure for testing the sand samples were in accordance with A.F.A. specifications. The equipment used in the various tests was as follows:

(a) Moisture Test—moisture teller and balance.

(b) Permeability Test—weights, sand rammer and permeability meter.

(c) Strength and Deformation Test — universal s a n d strength machine with deformation accessory.

(d) Fineness Test—oven, rapid sand washer, sieves and laboratory sifter.

(e) High temperature investigations of the sand samples were made with the sintering tester for the sintering test.

(f) Expansion and contraction at 2500°F. was measured by the dilatometer on 1⅛-in. dia. by 2-in. long double-end rammed specimens.

<div align="center">SAND SAMPLE NO....................</div>

(1) **Sand Contains:—**
Reclaimed Sand.................... Washed Silica.....................
Crude Silica.............. Clay.............. Bentonite..............
Cereal.............. Molasses.............. Silica Flour..............
Core Oil.............. Cement..............

(2) **Sand Used As:—**
Green Sand Facing.............. Dry Sand Facing..............

(3) **Sand Rammed By:—**
Hand.............. Jolt.............. Sand Slinger..............

(4) **Flask:—**
Sides of Flask vented? Yes.............. No..............

(5) **Mold Protection:—**
Mold wash used? Yes.............. No.............. Description of mold wash..

(6) **Weight of casting poured in sand:—**
Minimum...................Lbs. Maximum...................Lbs.

(7) **Thickness of casting wall:—**
Minimum..............inches. Maximum..............inches.

(8) **Metal:—**
Temperature of Metal Poured.............. degrees Fahr.
How measured...................... Corrected for Emissivity?
Yes.............. No..............
Type of metal..

(9) **Finish:—**
Smooth.............. Medium.............. Rough..............
Type of Casting Defects..
Samples of Sand Submitted by: ....

FIG. 1—FORM OF QUESTIONNAIRE SENT TO STEEL FOUNDRIES SUBMITTING SAND SAMPLES.

5. Sand samples specified as used for dry sand molding were tested for expansion and contraction after oven drying. The green sand samples were tested for expansion and contraction by placing a rammed green sand specimen in the dilatometer. The specimen suddenly immersed in the furnace was subjected to a temperature of 2500°F. and immediately expanded. Maximum expansion per in. of specimen was recorded. After the sand reached its maximum expansion, it would then start to contract. Contraction per in. was read in all cases after ten minutes immersion at 2500°F. Expansion shows growth of sand grains while contraction shows decrease of sand volume due to vitrification of clay bond. The sand test results and foundry conditions for green sand molding are tabulated in Table 1. The dry sand test results and foundry conditions are tabulated in Table 2.

6. The fineness test data are shown in Tables 3 and 4. The test data are shown in two groupings, namely, green sand and dry sand. This grouping permits a detailed study by which many correlations between sands and foundry methods used by the different foundries can be made.

SUMMARY OF SAND AND FOUNDRY CONDITIONS

7. A study of the types of sands and foundry conditions reported from the 17 steel foundries (Table 5) reveals many interesting facts and relationships worthy of brief discussion.

Kinds of Sands

8. All the sand samples submitted were of the synthetic type. Crude silica sand was used extensively for both green and dry sand molding. Crude and washed silica sand and reclaimed sand were used for green sand molding.

9. The binders used consist of clay, bentonite, cereal, molasses and pitch. Blends of these binders were reported to secure workability, the proper dry sand strength as well as an early breakdown. Clay, bentonite, molasses and silica flour apparently are preferred for dry sand molding while clay, bentonite and cereal are preferred for green sand molding.

Facing

10. Facing sands are used extensively in the steel foundry for both green and dry sand molding. The purpose of preparing these facing sands is to secure sands with maximum refractoriness.

Table 2

SAND TEST RESULTS AND FOUNDRY CONDITIONS FOR DRY SAND SAMPLES SUBMITTED

SAMPLE NO.	1	2	3	4	5	6	7	8	9
MOISTURE (Per Cent)	6.5	6.0	5.9	6.3	5.9	3.9	4.3	3.5	4.1
PERMEABILITY (A. F. A. No.)	78.0	82.5	63.1	113.0	81.3	261.0	66.5	82.0	83.0
GREEN COMPRESSION (Lb. Per Sq. In.)	5.9	7.4	5.5	4.3	6.3	7.2	5.1	7.8	7.8
GREEN SHEAR (Lb. Per Sq. In.)	1.6	1.9	1.6	1.2	1.9	2.3	1.4	2.1	2.4
DRY SHEAR (Lb. Per Sq. In.)	124.0	60.0	48.6	27.8	47.6	58.5	43.8	58.5	60.3
DEFORMATION (In. Per In.)	0.031	0.027	0.026	0.046	0.043	0.017	0.033	0.028	0.027
RESILIENCE (Lb.-In. x 1000)	183	200	148	198	270	128	163	204	210
FLOWABILITY	72	70	74	81	------	77	74	66	74
EXPANSION (In. Per In. at 2500°F.)	0.018	0.018	0.019	0.018	0.010	0.019	0.017	0.022	0.021
CONTRACTION (In. Per In. at 2500°F.)	—0.012	—0.015	—0.015	—0.010	—0.005	—0.005	—0.009	—0.009	—0.008
SINTERING POINT (°F.)	2650	2590	2680	2510	2492	2650	2572	2494	2540
SAND BINDER—(TYPE) C = Cereal B = Bentonite CL = Clay SF = Silica Flour MO = Molasses FC = Fire Clay SO = Silica Oil	CL B—C SF	B—MO SF	CL B—MO SF	FC C—MO SF	CL MO—SF	B—C	CL MO—SF		CL—B
MOLD PROTECTION (TYPE) B = Bentonite S = Silica SO = Silica Oil MO = Molasses	B—S MO	S—MO	S—MO	SO	SO	S—MO			SO
RAMMED BY	Hand	Air Rammer	Air Rammer	Hand Jolt	Hand	Air Rammer Sand Slinger	Hand Sand Slinger		Hand Jolt

Table 2 (Continued)

Sample No.	1 LC—LA	2 MC P & A OH	3 MC P & A OH	4 P & A	5 CS	6 MM	7 CS OH	8 SS	9 C & A
Metal (Type Steel)									
LC = Low Carbon									
LA = Low Alloy									
MC = Medium Carbon									
PC = Plain Carbon									
OH = Open Hearth									
MM = Medium Manganese									
SS = Stainless									
C = Carbon									
A = Alloy									
Pouring Temperature (°F.)	2850	2900	2900	2900	2700	2830	2850	——	2800
Weight of Castings Poured (lb.)	1200 to 50,000	500 to No Max	500 to No Max	300 to 20,000	300 to 10,000	400 to 700	480 Min.		50 to 1500
Casting Section Sizes (inches)	2 to 20	15 to No Max	1 to 15	1 to 12	1 to 4	1 to 8	⅜ to ¾		¾ to 2
Casting Finish	Smooth	Smooth	Smooth	Medium Tends to Burn In	Smooth	Smooth	Smooth		Smooth
Casting Defects					Castings not Smooth Enough	Dirt Spots	Sand Holes Cracks		

Table 3

SCREEN ANALYSIS OF GREEN SAND SAMPLES SUBMITTED

SAND (KIND)
WS = Washed Silica
CS = Crude Silica
RS = Reclaimed Sand
HS = Heap Sand
OS = Old Sand
NEV. S = Nevada Sand

Sample No.	1	2	3	4	5	6	7	8	9	10	11	12	13	14	15	16	17	18	19	20	21	22	23	24
Sand (Kind)	WS	WS &CS	CS	CS	RS	RS	CS	WS	RS &CS	RS &CS	WS &CS	RS &WS	RS WS &CS	HS	NevS	OS	RS &CS	RS &CS	RS &WS	WS		WS RS &CS	RS &WS	CS
Screen Analysis No. (Per Cent Material @ Remaining on Respective Sieve)	0.1	0.0	0.0	1.0	0.0	0.0	0.0	0.0	0.0	0.0	0.0	0.0	0.0	0.0	0.4	0.2	0.2	0.0	0.1	0.0	0.0	0.1	0.1	0.0
12	0.1	0.2	0.0	0.2	0.0	0.0	0.0	0.3	0.0	0.0	0.1	0.0	0.0	0.0	0.6	0.1	0.7	0.1	0.7	0.0	0.0	0.4	0.1	0.0
20	0.4	0.2	0.1	0.1	0.1	0.1	0.3	15.9	0.4	3.2	20.2	0.2	0.1	0.1	0.6	0.5	0.5	0.3	4.6	1.8	0.1	0.2	2.0	0.3
30	0.4	2.7	1.3	0.1	2.5	1.5	4.3	20.0	0.6	19.0	24.1	3.8	1.9	0.3	2.7	2.4	2.8	0.8	14.9	7.1	0.7	0.3	4.8	2.3
40	19.3	31.1	20.1	7.9	36.6	38.9	52.8	20.2	8.5	32.1	28.1	31.7	28.0	30.1	32.7	27.6	26.8	7.6	30.1	18.6	2.8	18.0	22.4	0.4
50	8.0	22.2	42.4	33.5	34.0	37.6	28.0	4.6	22.7	17.3	12.4	27.3	21.4	26.5	27.0	32.4	29.3	32.3	10.6	19.5	20.5	28.9	24.0	17.1
70	44.7	8.1	17.2	31.7	4.3	1.1	5.3	16.3	26.7	8.2	3.0	10.1	20.4	9.6	16.7	11.8	16.7	32.1	23.5	20.6	17.7	19.0	10.2	81.7
100	15.0	6.0	10.3	13.8	18.1	12.6	4.3	12.2	20.0	6.2	2.7	8.8	16.6	8.7	4.0	11.2	7.4	12.6	7.0	14.1	31.3	13.6	8.3	19.2
140	10.6	4.7	2.8	4.5	1.0	0.3	0.6	1.5	8.9	2.0	1.0	3.2	8.7	8.0	2.4	8.9	4.0	5.5	2.8	4.5	13.4	5.4	4.2	10.3
200	1.8	3.3	0.6	1.8	0.4	0.3	3.6	0.6	2.6	0.7	0.2	1.4	2.5	2.3	1.1	1.0	1.3	1.6	1.4	2.2	3.8	1.6	1.7	2.6
270	0.5	1.3	0.4	0.6	0.1	0.5	0.5	0.6	0.6	0.3	0.3	0.3	0.8	1.6	0.6	0.6	0.6	0.0	0.3	0.7	1.0	0.8	1.0	0.8
Pan	0.1	6.1	0.3	1.0	0.3	0.3	0.3	0.6	0.6	0.9	0.8	0.7	0.7	3.1	0.8	0.8	1.3	1.4	0.4	1.1	1.2	1.2	2.6	0.6
Fines (Per Cent)	2.4	10.7	1.3	3.4	0.7	0.6	4.4	1.7	4.3	1.9	1.3	2.3	4.0	7.0	2.5	2.4	3.2	3.9	2.0	4.0	5.5	3.6	5.8	4.0
Clay (Per Cent)	4.0	18.7	4.5	8.8	7.7	7.8	4.1	7.3	8.0	10.1	7.2	5.0	8.9	9.7	10.4	7.9	8.6	5.4	3.8	0.9	8.0	8.4	11.7	5.7
Fineness (A.F.A. No.)	56.7	66.8	46.5	34.7	42.0	40.8	41.2	38.1	58.4	39.4	29.9	46.2	49.5	55.2	44.6	48.0	49.1	57.5	42.6	52.6	69.0	54.2	54.8	60.3

Table 4

SCREEN ANALYSIS OF DRY SAND SAMPLE SUBMITTED

SAMPLE NUMBER		1	2	3	4	5	6	7	8	9
SAND (KIND)		Nev. S	CS	RS-CS	CS	WS	CS	CS	—	CS
Nev. S = Nevada Sand										
CS = Crude Silica										
RS = Reclaimed Sand										
WS = Washed Silica										
SCREEN ANALYSIS	Sieve No.									
(Per Cent Material	6	0.0	0.0	0.2	0.1	0.0	0.0	0.0	0.0	0.0
Remaining on Res-										
pective Sieve)	12	0.0	0.1	0.4	0.8	0.0	0.0	0.0	0.0	0.0
	20	0.3	0.1	0.1	2.9	1.8	0.2	0.3	0.2	0.2
	30	0.1	1.8	2.0	8.0	8.8	3.8	1.1	4.0	2.6
	40	29.6	23.6	26.0	33.4	18.5	31.7	11.0	10.0	10.5
	50	30.4	2.6	24.5	20.9	18.3	27.3	24.2	21.3	17.2
	70	5.0	28.3	13.2	5.9	19.3	16.1	11.6	5.5	29.7
	100	2.0	7.1	5.8	6.7	10.3	8.8	23.2	31.5	16.8
	140	2.6	5.4	3.6	1.9	4.0	3.3	11.2	6.2	8.0
	200	2.2	1.6	1.8	1.4	1.2	1.4	3.1	2.0	2.3
	270	2.8	2.1	1.6	0.6	0.6	0.2	1.3	1.5	0.6
	Pan	10.6	11.3	6.3	3.0	3.1	0.7	0.8	4.5	1.5
FINES (Per Cent)		15.6	15.0	9.7	5.0	4.9	2.3	5.2	8.0	4.4
CLAY (Per Cent)		14.4	16.0	14.8	14.4	14.1	6.5	7.2	13.3	10.6
FINENESS (A.F.A. No.)		79.5	73.0	66.5	49.1	55.5	46.2	64.0	67.1	60.6

Table 5

GENERAL SAND AND FOUNDRY CONDITIONS

Conditions	Green Sand	Dry Sand
SYNTHETIC SAND	All	All
Reclaimed sand	12	1
Washed sand	10	1
Crude silica sand	12	7
BINDERS (used alone or in combination)		
Clay	14	6
Bentonite	14	5
Cereal	15	3
Molasses	1	5
Silica flour	None	6
Fuel oil	1	None
Pitch	1	None
A. F. A. Fineness Number less than 50	13	2
A. F. A. Fineness Number greater than 50	11	7
Clay less than 7.0 per cent	9	1
Clay greater than 7.0 per cent	15	8
Less than 75% of sand grains on three consecutive screens	3	8
More than 75% of sand grains on three consecutive screens	11	1
Less than 75% of sand grains on four consecutive screens	5	7
More than 75% of sand grains on four consecutive screens	19	2
Less than 65% of sand grains on three consecutive screens		2
More than 65% of sand grains on three consecutive screens		7
Less than 65% of sand grains on four consecutive screens		1
More than 65% of sand grains on four consecutive screens		8
Mold surface not protected	16	None
Mold surface protected	7	All
Pouring temperature less than 2900°F.	6	5
Pouring temperature greater than 2900°F.	14	3
CASTING FINISH:		
Smooth	9	7
Medium	11	1
Rough	1	0
Number reporting casting defects due to sand	10	None
Number reporting no casting defects due to sand	13	None

The use of mold protection was reported in 100 per cent of the cases for dry sand molds while 30 per cent of the green sand molds were reported as protected.

Sintering

11. Sintering temperature is the measure of a sand's refractoriness. Sintering temperatures for steel sands range from 2450 to 2750°F. Sand possessing the higher sintering temperature consisted of crude silica or washed silica. The sintering temperature of the sand decreases rapidly as reclaimed sand is added to the facing. The maximum sintering temperature of any sand is obtained with a minimum clay content as shown in Fig. 2.

12. Of the foundries reporting defective castings due to sand, it was found that the probability of producing defective castings was 33 per cent when the sintering temperature was over 2700°F. while the probability of producing defective castings was increased to 70 per cent when the sintering temperature fell below 2450°F. This relationship is shown in Fig. 3.

13. The sintering test offers a most promising method to measure the suitability of a steel molding sand. The relation be-
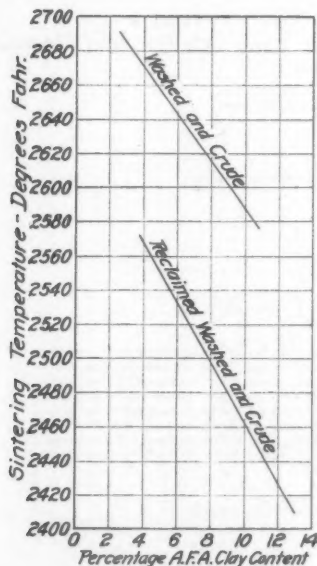
FIG. 2—EFFECT OF CLAY CONTENT ON SINTERING POINT OF GREEN SAND.
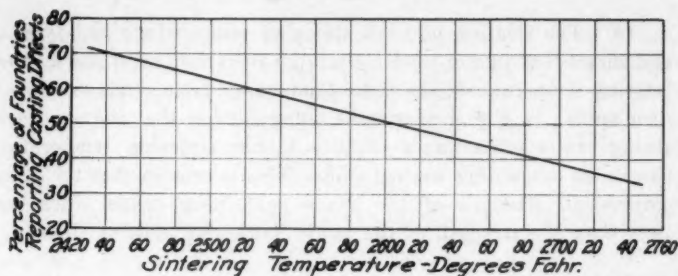
FIG. 3—PERCENTAGE OF FOUNDRIES REPORTING CASTING DEFECTS AS RELATED TO SINTERING TEMPERATURE.

tween the sintering temperature of a steel molding sand and the green compression strength and type of sand, such as, crude, washed or reclaimed sand, is shown in Fig. 4. The sintering temperature of this group of sands increases rapidly as the green compression strength decreases. As the strength decreases, the clay content decreases which reduces the amount of fluxing material present in the sand, thus giving a greater sintering temperature.
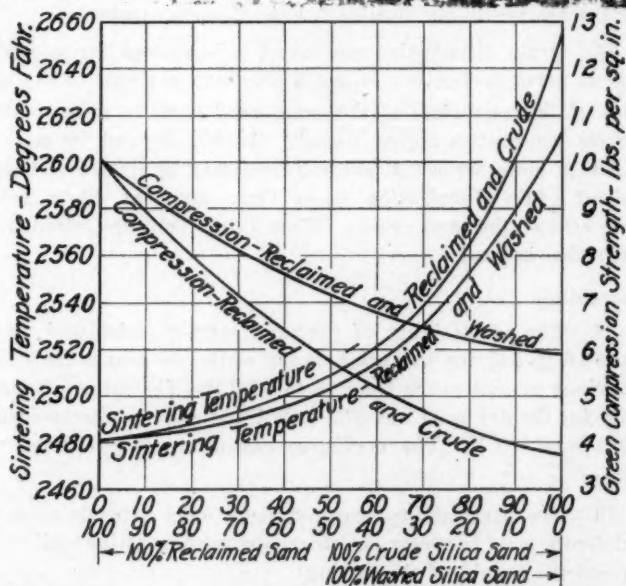
FIG. 4—THE RELATIONSHIP BETWEEN SINTERING POINT, GREEN COMPRESSION AND TYPE OF SAND FOR GREEN SAND.

14. The relation between sintering temperature and type of sand shows that facing sand containing reclaimed sand has a lower sintering temperature than those sands made from crude or washed silica sands. In Fig. 4, it may be noted further that the sand containing crude silica has a slightly higher sintering temperature than sands containing washed silica. This is true in that the green compression strength of the sands containing crude silica was lower than the strength of the sands containing washed silica.

Fineness

15. The fineness test data of Tables 3 and 4 show a remarkable uniformity of grain size, although it is well to note that the dry sands shown in Table 4 possess an average A.F.A. fineness number greater than 50. This explains why silica flour was used extensively in the dry molding sands to reduce fineness of the sand. Seventy-five per cent of these sands possessed more than 5 per cent fines, fines representing that material passing the 140 mesh sieve.

16. From Table 3, it may be noted that approximately 85 per cent of the green sands contain less than 5 per cent fines, with a correspondingly smaller A.F.A. fineness number.

17. Grain distribution was found to be narrow for most steel sands as noted in Tables 3, 4 and 5. Seventy per cent of the sand grains of three-quarters of the green sand samples were retained on three consecutive sieves, namely, the 40, 50, and 70 mesh of the A.F.A. sieve series. A molding sand may be classed as having excellent grain distribution when three adjacent sieves retain 75 per cent of the sand grains. When 70 per cent are retained the distribution is good.

Permeability

18. The permeability of the sand samples submitted ranges from 64.6 to 545 for the green sands while the maximum weight of castings poured range from 19 to 4000 lb. The range of permeability for the dry sand was 32.5 to 261 while the minimum casting weight was 300 lb. The maximum casting weight reported was 25 tons.

19. The permeability required for a mold depends upon the mold depth, mold hardness, moisture percentage, casting wall thickness and the weight of the casting.

20. The A.F.A. fineness numbers and permeability values as shown in Tables 1 and 2 may be plotted to show a fairly close re-

lationship, as shown in Fig. 5. This relationship is of practical importance in that the permeability of a steel molding sand may be determined approximately when the fineness number is known, by the use of the graph (Fig. 5) or formula, which is:

$$\text{Permeability} = 0.37S \times \left[\frac{714}{M-16} \right]^2$$

Where S=1 for round sand grains and M=A.F.A. fineness number.

Casting Finish

21. One request contained in the questionnaire was to indicate whether the casting surface was smooth, medium or rough. The casting sizes reported for green sand molding ranged less than 2000 lb. Thirty-nine per cent reported smooth and 55 per cent medium finish. For the green sands reported as giving a smooth finish, the average permeability was 208 and the permeability of the sands reported to give a medium finish was 218.
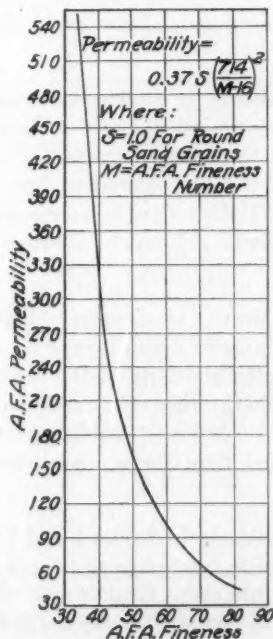
FIG. 5—THE RELATION BETWEEN PERMEABILITY AND A.F.A. FINENESS NUMBER.

22.   Eighty-three per cent of the castings made in dry sand total in excess of 2000 lb. while for the casting finishes for these castings 86 per cent were reported as smooth.

23.   The average green permeability for the dry sands submitted was 97. This would indicate that low permeability is possible when gas due to water vapor is not present. This lower permeability, together with mold wash, is conducive to producing a smooth casting finish.

Moisture

24.   The moisture content of steel sand must be controlled necessarily within narrow limits, since this variable can be traced as the cause of many casting defects and molding difficulties. Practical and suitable measuring equipment and moisture testing should become a rule for steel foundries.

25.   Moisture contents in the steel sands necessarily are kept low. Three per cent moisture was the average for the green sands while the dry sands averaged 1.5 per cent higher or 4.5 per cent moisture.

26.   The low moisture in steel sand is made possible by the coarseness of the sand grains, absence of fines and the extremely plastic bonds used in the synthetic mixes. The quantity of water required in any sand is that required to develop the proper plasticity or deformation. Deformation is a term applied to the amount the A.F.A. sand specimen deforms or shortens before it will rupture when loaded.

27.   Moisture in the dry sands must be higher for two reasons: to satisfy the slightly greater percentage of fines in these sands and also to develop a greater plasticity of the bond material, resulting in a higher dry strength when the free moisture is driven off. Thus, it may be said for dry sand molding that 1.0 to 2.0 per cent more water is required than for a corresponding sand used for green sand molding.

28.   It is shown clearly in Tables 1 and 2 that cereal binders, when used alone, require less water to temper than do sands containing bentonite and fire clay. Washed and crude silica sands also require less moisture to temper than do sands when reclaimed sand is added.

Green Strength

29. The green strength required of a sand depends largely upon the pattern construction and depth of draw. This information was not secured from the various foundries so no comparison of this nature can be made. The A.F.A. clay content bears a relationship to the green compression strength as shown in Fig. 6.

30. The average green compression strength of green sand was 6.3 lb. and 6.2 lb. for dry sand. The higher the green strength, the lower the sintering temperature in that clay bond may be considered as a fluxing element in with silica. This relation between green compression and sintering is shown in Fig. 7.

Dry Shear

31. As previously mentioned, the moisture content of steel sand bears a relationship to the dry strength of the green sands submitted. Thirty-eight per cent had dry shear readings less than 25 lb. per sq. in. For the dry sands, 13 per cent had dry shear strengths less than 25 lb.

32. The average dry shear strength for the green sands submitted was 35 lb. per sq. in. as compared to 58 lb. per sq. in. for
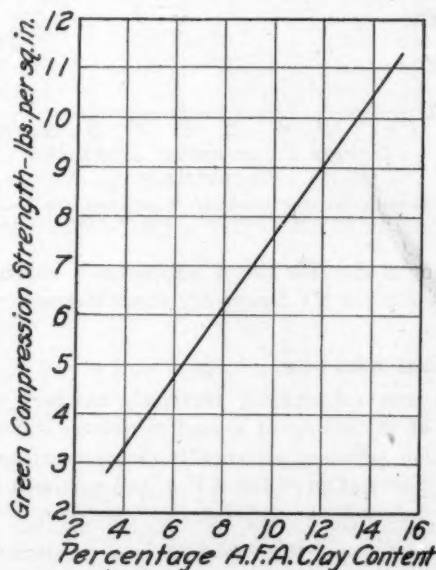
Fig. 6—Effect of Clay Content on Green Compression Strength of Green Sand.
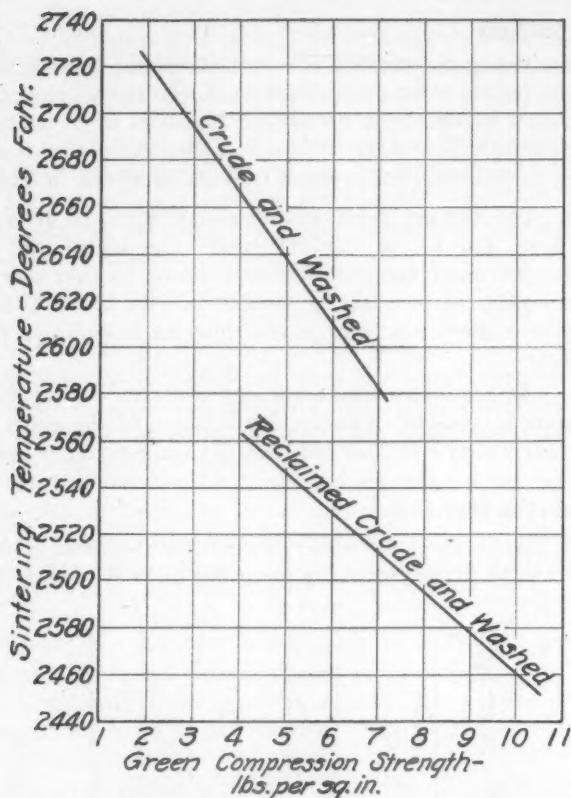
FIG. 7—THE RELATIONSHIP BETWEEN SINTERING TEMPERATURE AND GREEN COMPRESSION
STRENGTH FOR DIFFERENT TYPES OF SAND.

the dry molding sands. The use of molasses as a binder is, to some
extent, responsible for the higher dry shear strength readings for
the dry sand.

Deformation and Resilience

33. The term deformation previously has been defined as a
measurement of the ability of a sand to deform without rupture.
The deformation accessory attached to the sand strength machine
measures the deformation of the A.F.A. test specimen at the point
of rupture.

34. This important test reading is a measurement of the
sand's plasticity and is proportional to the percentage of moisture

required to develop the proper workability of the bonding material being used. It is expressed as linear change of length of the A.F.A. test specimen in thousandths of an inch. The product of deformation and green compression strength is expressed as resilience.

35. Referring to Tables 1 and 2, the resilience values for green sand molding, it may be noted that 75 per cent of the green sands show a resilience less than 160. The sands used for dry sand molding indicate that higher resilience values are required, for example, 75 per cent of these sands have resilience values greater than 160. As the resilience of either green or dry molding sand increases, the sintering temperature decreases rapidly, as graphically shown in Fig. 8.

Expansion and Contraction

36. The relationship between the factors giving rise to ex-
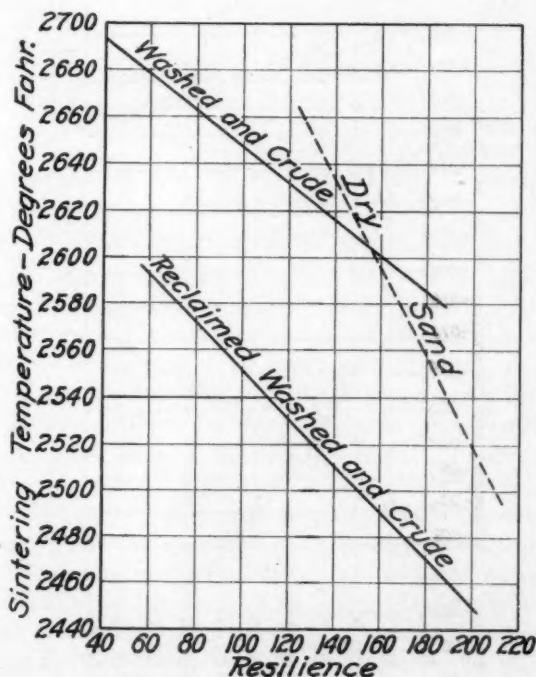
FIG. 8—THE RELATIONSHIP BETWEEN SINTERING TEMPERATURE AND RESILIENCE FOR GREEN AND DRY SAND.

pansion and contraction of steel molding sands are shown in Figs.
9 and 10.

37.   The expansion and contraction values represent the ac-
tual growth in length and shrinkage of the molding sand when this
sand is subjected to a shock temperature of 2500° F. The narrow
fineness range, grain distribution and absence of fines cause the
expansion of both the green and dry sand to remain a constant,
irrespective of the A.F.A. fineness number.

38.   The mechanics of the expansion and contraction of sands
are best understood by considering that only sand grains expand
and the clay or binders contract. The contraction of molding sand
takes place as the clay sinters or cereal binder is burned out.
Therefore, it is reasonable to expect that sands having higher per-
centages of clay should expand less than those with a lower per-
centage, as shown in Figs. 9 and 10.
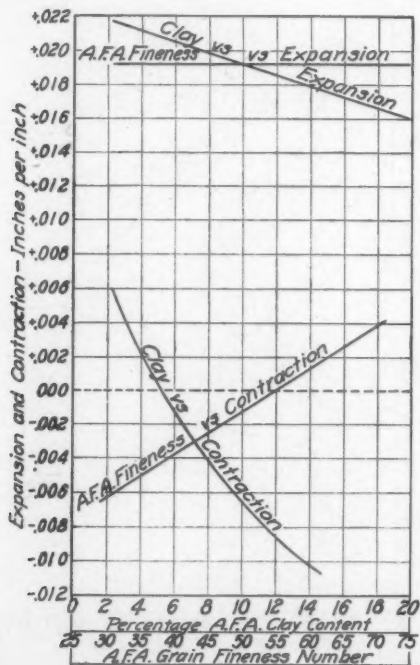
Fig. 9—The Relationship Between Expansion and Contraction, and Clay
Content and A.F. A. Grain Fineness Number for Green Sand.
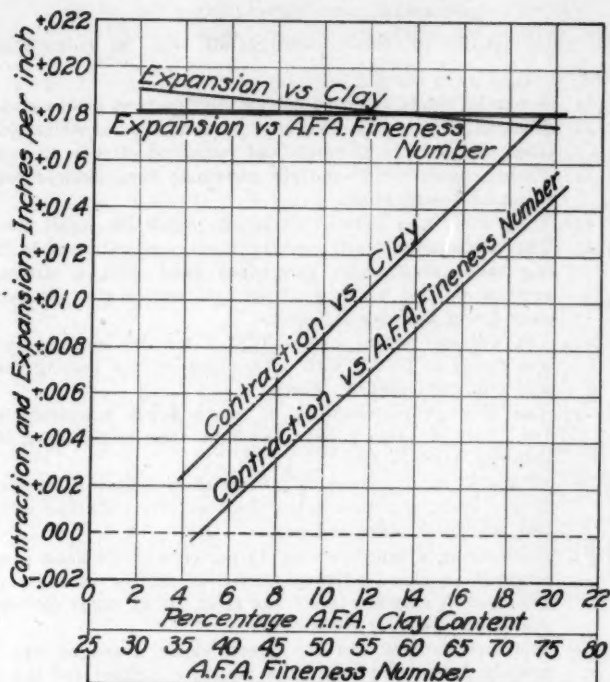
FIG. 10—THE RELATIONSHIP BETWEEN EXPANSION AND CONTRACTION AND A.F.A. CLAY
CONTENT AND A.F.A. FINENESS NUMBER FOR DRY SANDS.

39. Referring to Fig. 9, it may be noted that for green sand
the contraction increases rapidly with an increase in clay content.
Practically every green sand contracted to below its original vol-
ume. The contraction of the green sand decreased as the A.F.A.
fineness number increased in that the heat flow through the sand
was hindered with an increase of fineness (See Fig. 9.)

40. Sand samples which were specified as used for dry sand
molding were oven dried before testing. These sands showed less
contraction as the clay and fineness number both increased.

41. From the study of expansion and contraction, the ques-
tion "why dry sand for molds?" can be answered conclusively.
Molding sands which are dried, or have no free water present, have
greater dimensional stability. They show less expansion and con-
traction upon heating. This smaller volume change will produce a
mold surface where the least possibility exists of producing scabs
and buckles.

SUMMARY AND CONCLUSIONS

42. The results of this investigation will be summarized briefly:

(1) Synthetic sands are used extensively for steel molding sand.

(2) Washed silica sand was used in nine sands, while crude silica was used in 16 sands and reclaimed sand in 12 sands.

(3) Facing sands which contain reclaimed sand showed lower sintering temperature.

(4) Silica flour was used in six of the eight dry sands tested.

(5) The maximum sintering temperature was obtained in facing sands made with new silica sand with a minimum amount of bond which was just sufficient to give the necessary green and dry strength.

(6) The average green permeability of the dry molding sands was found to be 97, with a resulting smooth casting finish and minimum casting defects.

(7) The average permeability of green sands submitted was 196. This is a much lower average than would have been obtained 5 years ago.

(8) Of the 33 sands tested, only three showed fines in excess of 10 per cent; the fines being that material retained on the 200, 270 mesh sieve and pan.

(9) Considering a sand showing 75 per cent of the sand grains retained on three adjacent sieves as having excellent distribution, it may be stated that 12 of the 33 sands submitted possessed excellent distribution.

(10) Moisture content for the green sands averaged approximately 3 per cent while the moisture content for the dry sands was approximately 5 per cent.

(11) Green strength for steel sands averaged 6.3 lb. for green sand and 6.2 lb. compression for dry sand.

(12) The average dry shear strength for green sands and dry sands was 35 lb. and 58 lb. respectively.

(13) Resilience values for green sands are normally less than 160 while resilience values for dry sands are normally greater than 160.

(14) The use of green sand facing was reported as limited to castings below 5000 lb. and with one exception (No. 14, Table 1) were rammed by jolt, squeezer or sand slinger.

(15) The use of dry sand molding was reported as limited to castings over 300 lb. and were rammed by hand, air rammer, sand slinger and jolt.

(16) Molds, which are dried, have less tendency to expand and contract when subjected to high temperatures.

(17) Flowability of steel sands tested was between 75 and 80 per cent. A flowability below 75 per cent gives rise to voids in mold surface causing penetration.

(18) Ten foundries reported the use of optical pyrometers for measuring metal pouring temperatures, three estimated,

while one used a standard cup and measured the time from liquid to solid to arrive at the estimated pouring temperature.

(19) Temperature of metal reported averages 2840°F. for the castings made in dry sand, while 2890°F. was the average reported for castings made in green sand.

(20) Types of metal reported were electric and open-hearth steel.

(21) The probability of casting defects increases as sintering temperature decreases.

(22) A sintering temperature over 2700°F. should be the objective for all steel molding sands.

(23) The sintering temperature of a steel molding sand decreases as clay content and green strength increases.

(24) Sintering temperature of steel sands bears a relationship to the resilience value, as all steel sands, irrespective of type of sands, have higher sintering temperature when the resilience value is decreased.

(25) The group with the lowest sintering temperature consists entirely of reclaimed sands or sands containing some proportion of reclaimed sand in the mixture, while the other group of sands with the higher sintering temperature consisted entirely of crude or washed silica sands or crude and washed silica sand mixtures.

(26) A formula which may be used to calculate the permeability of steel molding sand is:

$$\text{Permeability} = 0.37 \ S \left[\frac{714}{M\text{-}16} \right]^2$$

Where $S = 1$ for round sand grains and $M = $ A.F.A. grain fineness number.

(27) Steel sands of reclaimed sand, or containing reclaimed sand, showed the highest green strengths and corresponding low deformations.

(28) Sands of washed and crude silica sand mixtures possessed the greatest deformation, followed by crude silica, reclaimed sand-washed silica sand mixtures, reclaimed sand-crude silica sand mixtures and reclaimed sand in the order named, which is the exact reverse of their green compression values.

(29) The expansion of the green sand decreases as clay content and green compression increases.

(30) The contraction of the green sand decreases as the clay content decreases and as the fines increase.

(31) The expansion of the dry sands decreases as clay content and green compression strength increase and is not affected by a small change in grain fineness.

(32) The contraction of the dry sands decreased as the clay content and fineness number decreased.

(33) In either green or dry sand the endeavor should be to com-
      pound sand so that the expansion would be a minimum and
      contraction be eliminated at 2500° Fahr., in which state the
      probability of casting surface defects are minimized.

DISCUSSION

Presiding: L. E. Everett, Key Company, East St. Louis, Ill.

F. A. Melmoth[1] *(written discussion)*: This paper constitutes a
very interesting and instructive collection of information on steel mold-
ing sands derived from sands in actual use. Its importance can be judged
from the acknowledged fact that future progress in steel foundries is
largely dependent upon developments in mold making materials.

It is an unfortunate fact that apart from silica sands, there exist
very few materials in nature in sufficient quantity, or regularity, which
exhibit the necessary qualities, both mechanical and chemical, to with-
stand the extreme heat and searching action of molten steel. It is all
the more necessary, therefore, that such investigations as the one now
being discussed should take place, so as to give us the maximum oppor-
tunity of getting the best out of such materials as are available, and a
frank interchange of experiences represents to my mind the best way
known at present.

There does not exist much controversial material in a paper of
this description but I think it is open to question as to whether
sintering temperature should be looked upon as of prime importance
right through the range of steel casting sizes. It has not been our
experience that a sand of very great refractoriness is as satisfactory
for light castings as one appreciably more fusible. We have found
that too high a refractoriness for the very light castings usually
carries with it a greater degree of penetration, at the abnormally
high temperature used to pour steel into this type of work. It is
customary, therefore, to work with a high green bond, and such a
sand is generally of lower refractoriness, on account of the increased
amount of more fusible binding material. I would say, therefore, that
sintering temperature was largely a matter of relative importance,
when viewed from the standpoint of size and mass of castings.

Most of us use silica flour, with a view to preventing penetration,
as part of our heavy dry facing sand. I will be interested in an ex-
pression of opinion from the authors as to whether a much finer
grade of sand would be equally satisfactory. Silica flour is compara-
tively expensive, increases health hazards, not only on the molding
floor but more particularly in the shake-out, and if economies could
be effected in this direction safely, they would be distinctly worth
while from both viewpoints.

I would also like to hear the author's opinions as to whether we
do not place considerably more insistence than necessary on the
permeability of facing sand. Assuming that adequate control is

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[1] Vice President and Works Manager, Detroit Steel Casting Co., Detroit, Mich.

exerted over backing sand, and also over the amount of new facing used upon the pattern, it is rather surprising to find how very low one can go on the facing permeability, without creating the slightest surface unsoundness.

I was glad to note the confirmation by the authors of the opinion that a higher moisture content was necessary in the making up of facing sand mixtures for dry sand work, in order to develop dry sand strength. This has been a fruitful cause of serious trouble with weak molds after drying, and we ourselves have found it very necessary to keep our moisture between 5 and 6 per cent at the mill.

H. L. McKINNON:[2] Permeability as high as 700 was encountered in some of the sands in this investigation. Is there any reason for believing that permeabilities above 400 are really necessary in the heavy steel casting industry and is there any line of demarcation that would clear up that point?

A few months ago in a plant doing heavy railroad work it was stated to me that they could not do satisfactory work unless the permeabilities were above 450. On similar work in another foundry permeabilities of 225 were entirely satisfactory. I have been trying to reconcile the two permeabilities. The only discrepancy I found between the sands in the two places was that in one case they were using an initial sand of a fineness No. 28, in the other a fineness No. 40 for the new sand. I do not know why permeabilities as high as 600 or 700 should be desirable.

MR. DIETERT: If we have a permeability around 250 we can make any casting whatsoever.

I do appreciate that the flask design has a great deal to do with the permeability required. If you have a flask that is well vented and the mold is not put in a pit, I see no reason to go into high permeabilities. If you are working in a pit where you are required to vent the mold from the top, higher permeabilities such as 500 may be desirable.

At this time may I answer Mr. Melmoth's question on permeabilities? I have made light steel castings in sand of 45 permeability. If you work a green sand which is high in permeability, then you would probably need a low sintering temperature sand to stop penetration which will glaze causing pores of the sand to close and stop penetration. This glaze would act very much like facing the mold with silica wash.

Where the sand permeability is correctly proportioned for the size of the job, a higher sintering temperature sand is the more desirable one. In times past, some foundrymen have felt they wanted to glaze the mold surface. That is one way of doing it, but there is a better way of doing it. That is to use a high refractory sand with just enough permeability to make it a safe molding job and producing a clean cast surface.

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[2] Vice President, The C. O. Bartlett & Snow Company, Cleveland, O.

MEMBER: Conclusion No. 17 says "A flowability below 75 per cent gives rise to voids in mold surface, causing penetration." I would like an explanation of that.

MR. DIETERT: A flowability below 75 per cent gives rise to mold voids in mold surface, causing penetration. If a sand has a low flowability and that sand is rammed in the mold, in certain parts of this mold you will find the mold surface to be spongy containing voids. In other words, the sand grains have not rammed together to create a continuous mold surface. A sand with low flowability will have on the mold surface voids in which you could stick the lead of a pencil. Naturally, when you pour metal against such mold surfaces, the iron will run into the mold voids and cause penetration defects.

A sand that is low in flowability is, as a rule, a stiff sand. It is putty-like or two rich in bond which is high in green strength. If you get excess bond in your sand, then it takes a large amount of energy to ram the sand grains in to place because excess clay means a stiff sand which is high in strength thus resisting flowing.

MEMBER: In connection with permeability for skin dried, normally torch dried molds, it has been my experience that we require considerably higher permeability in a mold that is to be skin dried than in a green sand mold and the reason for it is the kick-back from the sudden line of demarcation between dry sand and green sand just under the surface of the mold.

MR. DIETERT: Where you torch-dry a mold, you have a layer of the mold which is dry and right behind it you have tempered sand. The dry sand, when heated, changes in volume quickly whereas the green sand (that is, the next layer back) is cooled by the water that is present and you certainly have to use an open sand to prevent what we call a kick-back action.

I do not know what permeability figure the member proposing this question has in mind but I feel that a 250 permeability or possibly a 300 permeability is entirely satisfactory for such work. We find such permeability figures to hold true also for air-dried molds where we take a green sand mold and let them set over night and pour them the next day instead of putting them in a dry sand mold.

As we increase the permeability of dry molding sand, we are able to dry the mold faster and more thoroughly then molds made with sand permeability of one-half the value. In many shops where they have to use high permeability sands, they are not equipped to dry molds thoroughly and do not use the correct time of drying.

We have made heavy castings in dry sand molds with permeability as low as 40 to stop penetration. If the mold is perfectly dry, there will not be any blows or kicks. If all the water is out of the sand, very little permeability is required. It takes a longer time to dry such molds.

MEMBER: A term used in this paper that I would like to have Mr.

Dietert further explain is this factor of resilience that is reported
and commented upon. Resilience of molding sands is a comparatively
new term and the practical aspect or application of that term might
be of interest to us.

MR. DIETERT: Pick up a handful of sand and squeeze it in the
hand. A certain sense of feel of this sand will be gotten as to its
softness or what we call its plasticity, or its richness. Then break the
sand and feel the bond strength. That is the compression strength.
But this plasticity, or softness, or richness that you feel is the
deformation factor. To measure that, just measure the shortening of
the sand specimen under the compression test. On the green sands
that were submitted, there is one sample that had a deformation of
only 0.009 per linear inch. That sand is brittle or low in plasticity.
Another sand in the green sand group read as high as 0.038. The
sand of 0.009 deformation was a very rich or plastic molding sand.
The sand of 0.038 deformation was almost gummy or pasty. It would
almost spring back when you squeezed it thus showing a high
deformation.

Both sands, the 0.009 and the 0.038 deformations could have been
of the same strength. You can make two steel castings in your shop
both of the same tensile strength but of different ductility. One could
be brittle steel and the other could be a soft ductile steel. The brittle
steel should not be subjected to shock loads; it would have entirely
different load applications than the ductile steel casting.

The same applies to molding sands. The same strength sands
with one sand brittle and the other real ductile or plastic may be had.
This difference is called deformation.

The brittle sand is the one foundrymen call rotten sand. The
testing machine showed it possessed good strength. The strength
reading was correct but the ductility or deformation was different
and should have been determined. Instead of speaking of deforma-
tion and strength, these two terms may be combined. Multiply the
deformation by the compression strength and obtain a figure which
is called sand resilience. This is the workable strength of your sand.

Resilience is a better criterion of bond strength of the sand than
just saying seven and a half pounds compression strength. If steel
sand is worked whose resilience would be down to a hundred, that
sand would drop very easily. It would be a brittle sand whereas a
sand with a hundred and sixty resilience were worked, that sand
would feel fine and would have a good strength. If worked up to two
hundred or two hundred fifty resilience, the sand will have good
strength but would be too rich. It would tend to make a very rigid
mold surface when dried. It would make a hard, dense mold mass
since there was an excess of plastic clay present.

There are two extremes. The rotten sand and the very plastic sand.
To get in between those two and get this moderate sand resilience,
evaluate the strength of your sand in pounds compression and defor-
mation, or, evaluate it in the one term "sand resilience."

Report of Committee on Radiography—1938

To the Steel Division, American Foundrymen's Association:

In January of this year, your Radiographic Committee sent out a questionnaire on "Radiography in the Steel Foundry." The questionnaire was mailed by the Society to 251 steel foundry organizations in the United States and Canada that were actively engaged in the production of steel castings. Organizations which operated more than one foundry were sent only one questionnaire on which they tabulated information from all their plants. It was stated in the questionnaire and the accompanying letter that if the organization answering the questionnaire wished to remain anonymous, it was at liberty to do so.

The committee received 49 replies from the 251 questionnaires sent out, or a return of about 20 per cent. The committee feels that this is not what might be considered as a poor showing in view of the usual returns that are made to questionnaires. It also feels that the percent returns were lower than expected, considering the importance of the subject and the fact that a producer could remain anonymous and still express his opinions.

Of those that replied to the questionnaire, 19 stated that they had used radiography. This means that approximately 39 percent of those that made returns had made some study of the subject. The committee is inclined to believe, however, that practically all those that have experienced the use of radiography made returns. If this condition is true, then only 7.5 percent of all the steel casting producers in the United States and Canada have actual radiographic experience, and only 2 percent, or 5 organizations, use radiographic procedure regularly.

The committee has not discounted the replies from the 30 organizations which had not used radiography. In fact, some very interesting information was made available by their replies, for they not only show the trend that they desire radiography to take, but they also show a live interest in the subject.

NOTE: Presented before Steel Session of the 42nd Annual Convention, Cleveland, O., May 17, 1938.

QUESTIONNAIRE REPLIES

In the following pages the committee reports in detail the questions as submitted and the answers received. The committee has also taken the liberty to comment on these answers.

Question 1.

a. Has your organization used x-rays or gamma ray non-destructive testing in the manufacture or inspection of steel castings?

$$\text{Yes} \dots\dots\dots\dots\dots\dots\dots 19$$
$$\text{No} \dots\dots\dots\dots\dots\dots\dots 30$$

b. If the answer to (a) is yes, please check one of the following:

1. Rarely for special examination..........10
2. Occasionally 4
3. Regular procedure 5

c. If so, has it been instrumental in causing beneficial improvements in manufacturing methods?

$$\text{Yes} \dots\dots\dots\dots\dots\dots\dots 12$$
$$\text{No} \dots\dots\dots\dots\dots\dots\dots 4$$
$$\text{Failed to answer}\dots\dots\dots 3$$

In answering b3, all 5 organizations stated that it was regular procedure on certain types of castings, such as airplane and high pressure and high temperature equipment.

It may be noted that 75 percent of those using radiography believe that it has produced beneficial results. This is called to the attention of those with no experience in radiography.

Question 2.

a. Have your customers used x-ray or gamma ray radiography for inspection?

$$\text{Yes} \dots\dots\dots\dots\dots\dots\dots 22$$
$$\text{No} \dots\dots\dots\dots\dots\dots\dots 17$$
$$\text{Failed to answer}\dots\dots\dots 10$$

b. Have the results of this inspection been mutually satisfactory?

$$\text{Yes} \dots\dots\dots\dots\dots\dots\dots 13$$
$$\text{No} \dots\dots\dots\dots\dots\dots\dots 9$$
$$\text{Failed to answer}\dots\dots\dots 0$$

c. Do you hesitate to use radiographic examination?

    Yes 31
    No 11
    Failed to answer.......... 7

d. If so is it because of:

1. Unfamiliarity with procedures..........10
2. Fear of incorrect interpretation of results.21
3. Fear of injury to personnel.............. 7
4. Cost of radiographic equipment..........27
5. Cost of radiographic examination (person-
   nel) 21

More information may be obtained if the answers are broken down into two groups, those that have used radiography, and those that have not, as shown by the answer to question 1a. These groups will be termed "Yes Radiography" and "No Radiography." Answers to question 2 break down as follows:

	Yes Radiography			No Radiography		
	Yes	No	Failed to Answer	Yes	No	Failed to Answer
2a.	10	6	3	12	11	7
2b.	6	4	0	7	5	0
2c.	9	9	1	22	2	6

2d.	Number		Number
1.	1	1.	9
2.	9	2.	12
3.	2	3.	5
4.	6	4.	21
5.	5	5.	16

In considering the composite answers, as well as the breakdown studies, it may be said that 2a and 2b results were as might be expected in that opinion was fairly divided with a slight advantage on the yes side.

The answer to 2c showed that two-thirds of those replying hesitated to use radiographic examination. If, however, this is broken down into the two groups, it is found that those who have not used radiography are considerably more fearful of using radiography than those that have used it.

In the composite tabulation for 2d the "cost of radiographic equipment" is given as the greatest reason for hesitating to use radiographic examination with the "cost of radiographic examination (personnel)" and "fear of incorrect interpretation of results" followed very closely. This is similar to what was recorded by those who have not used radiography whereas with

those that have used radiography, the "fear of incorrect inter-
pretation" received the most consideration. There were very few
comments made by producers to this section. One manufacturer
said that he hesitated to use radiography but that he did not
object to its use. Another producer stated that he believed the
hesitancy in using radiography can be the result of reasons other
than those given—namely, failure of various personnel of the
organization to see advantages to radiography, fear that it will
increase production costs and fear of using and depending on a
tool, the familiarity of which is not widespread.

Question 3.

Will you take orders for castings that are subject to radio-
graphic examination?

Yes 26
No 12
Failed to answer..........11

A study of the returns shows that those who have not used
radiography are about divided in their opinion, whereas those
that have used radiography are decidedly favorable to taking
orders which require radiographic examination.

The following is how the two groups answered on question 3.

Group	Yes	No	Failed to answer
Yes Radiography	14	2	3
No Radiography	12	10	8

In commenting on this question, one producer claimed that
they would take orders subject to radiographic examination, pro-
viding there was a mutual rejection agreement. Three producers
made the exception that if the design did not lend itself to the
production of a sound casting in their opinion, they would not
take orders which involved radiographic inspection.

Question 4.

a. Would you use radiographic examination in production
work to improve casting technique?

Yes 37
No 6
Failed to answer.......... 6

b. If the answer is No to (a) are you fearful that because

of such equipment, the customer will demand more than you are
capable of producing?

<div align="center">

Yes 3

No 3

</div>

c. Would you be willing to use radiography if your customers
did not know you had such equipment?

<div align="center">

Yes 18

No 2

Failed to answer..........22

</div>

It was indeed interesting to note that so many producers
would use radiographic examination in connection with studies in
their casting technique. It is evident that they believe that radio-
graphic inspection is a benefit to the foundryman. Only 6 pro-
ducers felt that they would not use it and 3 of these were afraid
that the presence of radiographic equipment would give their
customers erroneous ideas. The other 3 either were not interested
in improving their castings, or they did not think much of radiog-
raphy as a tool by which improvement could be obtained.

In answering 4c we assume that only those that answered 4a
would be interested. The two that answer "no" on 4c also
answered "no" on 4a. The 22 which failed to answer this
question, were part of the 37 that answered "yes" on 4a. These
producers evidently felt that they would use radiographic methods
regardless of the consumer's knowledge as to the use being made
of radiography.

Of the 37 that answered "yes" to question 4a, 20 of them
were producers that had not used radiography. In general, the
comment received on this question was that the producers would
want to use any device, if commercially practical, to better their
product. However, several thought that the price of such equip-
ment was not within reason, and it had prevented them from
using it.

Question 5.

Do you believe that there are cases where castings should be
rejected because of internal unsoundness as shown by radio-
graphic examination?

<div align="center">

Yes 41

No 1

Failed to answer.......... 7

</div>

Comments on this question are as follows:

1. Yes, if after radiographic examination the estimated cost of repair exceeds the cost of replacement.

2. Yes, if the unsoundness affects the designed strengths or serviceability.

3. In these days of high quality welding, a defect revealed by radiography can sometimes be removed and the casting repaired by welding, followed by heat treatment and re-examination if desired.

4. An answer of "yes" applies to those cases where the defect is obviously harmful and would weaken the casting. An answer "no" should be applied to cases where there was uncertainty concerning the extent of the defect and its influence on the weakening of the casting.

Question 6.

Who should have authority for such rejection?

a. Producer 8

b. Consumer 1

c. Umpire (radiographic expert).................... 9

d. By mutual agreement of producer and consumer. . 33

Some producers checked more than one answer to this question. All answers were, however, used in the tabulation. The outstanding feature is that the steel casting manufacture is not impartial but that rejection can be settled by mutual agreement between the producer and the consumer.

There was only one producer who cast his vote for consumer authority. In doing this, the producer stated: "Authority for rejection should rest with the consumer provided the consumer is competent to judge what should constitute a reasonably sound steel casting. It is believed that in any case the best authority for rejection would be a competent consumer with advice of a radiographic expert." In all fairness it should be added that this producer is also a consumer and has had considerable experience with radiography.

Question 7.

  a. Do you regard radiographic findings completely reliable?

                    Yes.......................12
                    No 18
                    Failed to answer..........19

  b. If answer is no, are there large uncertainties due to doubtful interpretations of films?

                    Yes 18
                    No 0

The (a) portion of this question is further broken down to show how the groups lined up in their answers.

	Group	Yes	No	Failed to answer
Yes	Radiography	10	8	1
No	Radiography	2	10	18

There is nearly an equal division of answers from those that have used radiographic examination with a slight favor for the belief that radiographic findings are completely reliable. The decided "no" reply given by those producers that have not used radiographic inspection would tend to indicate that they were slightly prejudiced and did not have sufficient experience to answer the question.

For such a highly controversial subject as denoted by the answers submitted there was received very little comment on this question. In general, however, the replies indicated that with competent operators, excellent radiographic technique and expert interpretation, radiographic findings could be considered completely reliable.

Question 8.

Would you use radiography for the exploration of defects such as to fully uncover the defect and determine the serviceability of the casting?

                    Yes 35
                    No 4
                    Failed to answer..........10

This question is very similar to 4a except that it is a little more definite and tangible. It was included not only to obtain the necessary information but to check on the character of the

replies obtained. The almost identical check of question 4a as obtained by the answers to this question affirms the sincerity with which the questionnaire was given by those producers who so kindly submitted replies.

Question 9.

a. Where radiographic inspection is used in acceptance tests, should this inspection be covered by specifications?

> Yes 33
> No 5
> Failed to answer..........11

b. What type of specifications do you feel should be used?

> Number of answers received................10
> Failed to answer..........................39

The answers to this question present an interesting point. A very large majority definitely decided that radiographic inspection should be covered by specifications but on the other hand they do not know what type of specifications should be used. Of the 10 comments received, 7 of them were submitted by producers that have used radiography.

In general, the 10 replies suggested that specifications should cover:

1. Generally accepted standards of radiographic technique.

2. Type of defect that will be permitted; type of defect that may be repaired; type of defect that will be the basis for rejection.

3. Points to be radiographed because of major stresses acting on the section.

4. Points to be radiographed because of possible defects which may result from molding technique or normal casting hazards.

One producer, who has used radiography, stated: "It is believed that specifications covering radiography of steel castings should never be drawn up in a standard form such as to preclude considerable detailed requirements for each and every individual casting under consideration. One criticism of most radiographic requirements for steel castings is that they are too general and subject to various interpretations.

On castings that are sufficiently important to require such exacting inspection, the purchaser should give the matter sufficient thought and individual attention to clearly indicate what is required on the individual casting. Such specifications which have been given individual intelligent consideration could be accepted by foundrymen without undue hazard.''

Another producer replied: "It is our belief that questions 9 to 12 inclusive, are almost impossible to answer at present. The greatest hindrance is a fear, widely held, that radiographic interpretation may be so much a matter of individual opinion as to constitute an intolerable weapon in some hands. Welding is known and understood and yet, individually, there occur cases of complete rejection of steel castings for one small local defect. The same individual probably inspects and approves a welded assembly the same day.''

It is felt that the answers to this question point out that certain specifications should be used in radiographic inspection but that considerable more study of, and experience with, radiographic inspection is necessary before these specifications can be defined.

Question 10.

Do you think that radiographic specifications should specify details of radiographic technique?

Yes 25
No 9
Failed to answer..........15

This reply should be broken down a little further to show the disposition of the answers, especially from the group of producers that have used radiography.

	Group	Yes	No	Failed to answer
Yes	Radiography	10	7	2
No	Radiography	15	2	13

It is evident from the answers that there is almost an equal division on this point by producers that have used radiography. While the replies tend to show that the subject is somewhat controversial, there were very few comments supplied.

One producer claimed that: "Specifications should specify details of radiographic technique in respect to sensitivity gauges

only." Another stated: "No technique specifications other than the requirements of a satisfactory film." A producer of opposite views wrote: "Technique should be specified as to film densities desired, type of penetrameters that are to be used, sensitivity desired, and a statement of the film to source distance to accompany each radiograph."

Question 11.

Do you believe that a mutual benefit would result to producers and consumers if a standard specification on details of technique would be prepared by some technical society, such as the American Society for Testing Materials?

Yes 34
No 2
Failed to answer.......... 13

The response definitely pointed out that there was a need for a study of technique standards. It would most certainly seem that the ideal organization to make such a study would be the American Society for Testing Materials.

In commenting on his answer, one producer made the following statement: "Standard specifications such as those prepared by the A. S. T. M., should largely consist of description and definition of technique and establishing broad standards to define different types of defects, but avoiding, insofar as possible the stipulation of acceptable or unacceptable conditions as they apply to the actual product."

Question 12.

In inspection and acceptance tests, should reference be made as to what parts or sections of the casting are to be radiographed?

Yes 31
No 3
Failed to answer.......... 15

It thus seems to be the concensus of opinion that, especially in large castings that can not be economically radiographed in their entirety, a specific reference should be carried as to the parts that are subject to extreme conditions and which should be inspected.

Question 13.

Is radiographic inspection of economic value to the:

<div align="center">

(a) Producer? 26

(b) Consumer? 22

</div>

Many of those that answered this question were of the opinion that it was of economic value to both. Three producers claimed that radiographic inspection was of no value to the producer and two claimed that it was of no value to the consumer. In general, it is believed that the answers to this question point to the fact that steel casting producers believe that radiographic inspection is of economic value to both the producer and the consumer.

A comment received stated the following: "Radiographic inspection will probably increase the producer's expenses. Where a consumer is having failures in service because of sub-surface defects, it is possible that an extra charge for the proven casting will result in final savings for him. It is quite likely also that the producer adjusting his technique following radiographic examinations may prove a saving finally in the foundry. At the beginning it is believed there will be confusion, and long time study can only determine the proper answer."

Question 14.

What do you think that the Radiographic Committee of the American Foundrymen's Association could best do to broadcast the benefits and disadvantages of radiography?

<div align="center">

Number of answers   23

Failed to answer   26

</div>

1. Inaugurate and carry on by publishing papers, by conferences, or by prepared technical discussions an educational program calculated to make the producer and the consumer radiographic conscious and to inform them both as to the advantages and disadvantages of radiographic inspection. Also act as a clearing house for specific information requested by the foundrymen on specific problems or cases . . . .*5 replies.*

2. Make a study of the existing commercial foundry installations and by a series of articles give the experience of these organizations with radiography. Cost data should also be included if possible . . . .*4 replies.*

3. Develop adequate standards and technique. Show types of work to which radiographic inspection is applicable..*3 replies.*

4. Summarize the work that has been done by technical societies; summarize the technical papers that have been presented on the subject, and report regularly on the progress that is being made in radiography.........................*2 replies.*

5. Publication of the results of the questionnaire in engineering publications and A.F.A. Transactions ......*2 replies.*

6. Investigate and recommend to producers the use of radiography as a useful foundry control instrument as a means of improving their product ....*2 replies.*

7. Stay out and do nothing ....*2 replies.*

8. Refrain from broadcasting, exploiting or advertising radiographic inspection. Emphasize the cost of this kind of examination, including the cost of replacing rejected castings. ....*2 replies.*

9. Gather more data and withhold definite conclusions until such time as results are uniformly conclusive........*1 reply.*

Question 15.

Please give some indication of your:
   a. Monthly tonnage output.
   b. An estimate of size (weight) of castings with the amount of tonnage represented.
   c. The types of castings produced.
   d. And, if desired, the name of your organization.

a. Monthly tonnage output:

Tons—Range	Number of Foundries
0 — 99	10
100 — 499	16
500 — 999	13
Over 1000	10

b. An estimate of size (weight) of castings with the amount of tonnage represented.

Size of Largest Casting	Number of Foundries
½ lb. — 100 lbs.	6
100 lb. — 1 ton	13
1 ton— 5 tons	14
5 ton— 10 tons	6
10 ton— 25 tons	6
Above 25 tons	4

Only three or four producers supplied the information de-
sired. In general, the question was answered by stating the
smallest and the largest size (weight) casting that they produced.
It was, therefore, necessary that the results be tabulated in the
above manner.

c. The types of castings produced:

Type	Number of Foundries
Pressure	13
Miscellaneous	17
Railroad	12
Machine	13
Automotive	2
Marine	2
Oil	3
Heat and Corrosion Resis.	4

A number of producers recorded one or more types of cast-
ings produced all of which are tabulated above.

FUTURE PLANS OF COMMITTEE

The committee believes that the above report should be re-
ceived by the members of the Society as an indication of the
initial period of development of radiography in the steel foundry.
It is the hope of the committee that another questionnaire em-
bodying perhaps the same set of questions may be sent out after
a period of five years. In this manner, the Society may be help-
ful to the steel casting manufacturer by showing him the trend
of the times.

RADIOGRAPHY DURING 1937

The most outstanding event in radiography during the past
year was the formation of Committee E-7 on Radiographic Test-
ing by the American Society of Testing Materials. An initial
meeting of those that might be interested in membership in such
a committee was held at Atlantic City in October 1937. At this
time the sphere of activity of the committee was outlined. It
was decided that all matters relating to the development and use
of radiography should be of concern to the committee. The broad
lines of policy as outlined by the temporary chairman were as
follows:

1. E-7 should not duplicate work done by other radio-

graphic committees. The policy should be to cooperate with other groups and to correlate all findings.

2. The sphere of activity of E-7 should be related to the improvement of the radiographic testing methods, to the proper use of the testing methods, and to the establishment of standards whereby radiographic results may be properly evaluated.

3. E-7 must have important men in its membership that its work will carry weight with industry. Also, all industrial groups actively engaged in radiography should be represented on the committee.

The organization meeting was held in New York City, February 14, 1938. Dr. H. H. Lester of Watertown Arsenal was elected chairman. Several sub-committees were established and chairmen were appointed. These sub-committees and their chairmen are as follows:

I. Radiography of Cast Metals.. C. W. Briggs
II. Technical Research .... H. E. Seeman
III. Radiography of Welding.... H. H. Lester
IV. Correlated Abstracts ...... W. P. Davey
V. Safety .... E. W. Page
VI. Papers and Publicity........ L. Thomassen

The chairmanship of the sub-committee on cast metals was accepted by your A.F.A. Radiographic Committee chairman because of the desire of the A.S.T.M. to cooperate with existing radiographic committees. Since the A.F.A. had designated your chairman to represent the A.F.A. on Committee E-7, this arrangement was acceptable to all concerned.

At a meeting held in Rochester in March, the organization of Committee E-7 was completed and a discussion of the activities of the various sub-committees was held.

Sub-Committee I on Cast Metals has a membership of 17 men, representing commercial casting organizations throughout the country. Four of these men are also members of the A.F.A. Radiographic Committee.

The present reported questionnaire pointed out that the industry thought it desirable that if some technical society such as the American Society for Testing Materials would study details on technique of radiography, that a mutual benefit would

result to both producers and consumers. It now appears that such
a wish has come to pass.

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Respectfully submitted

COMMITTEE ON RADIOGRAPHY

C. W. Briggs, *Chairman*

</div>

E. J. Ash	L. C. Wilson
L. E. Everett	P. E. McKinney
Fred Grotts	J. J. Curran

<div style="text-align: center">

E. W. Page

</div>

DISCUSSION

Presiding: JOHN HOWE HALL, Philadelphia, with W. C. Hartman,
Bethlehem Steel Co. as vice chairman.

F: A. MELMOTH[1]: The question of radiography is one of which
we all know so little that frankly, where I am concerned, I rather
hesitate to enter into a discussion. Actually, what we are after is
education, and we do not necessarily get that by advancing our own
opinion when it is, at the best, only half formed.

Our firm tried to answer the questionnaire. It was an extremely
difficult questionnaire, to my mind, to answer. I feel that Mr. Briggs
should not be discouraged if in some cases he did not get just the
information he wanted. As I say, we have no basis of fact and real
experience in most cases. There are a few foundries who have that
and who are going ahead and getting quite a lot of information, but
most of us can only do our best to answer any questionnaire and hope
that we will help the movement along.

Personally, I think that the use of radiography is coming, and
we should not attempt to hinder it by any prejudice. We ought to
help it along and in a very few years' time the x-ray and other modi-
fications will probably be among our best and most useful foundry
tools. If we look upon it as that and do not become afraid that any-
one using the methods is doing it to scare the foundryman, I believe
everything will be perfectly all right. In most cases, in conversations
in which I have joined, there is a fear that radiographic inspection
will be used as a whip with which to scourge the foundrymen, rather
than a tool to help them. I do not believe that is the correct mental
attitude to take, and I am perfectly sure that if we keep our minds
open, we are going to use this new tool with very great advantage
to ourselves.

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1 Vice President and Works Manager, Detroit Steel Castings Co., Detroit, Mich.

WM. H. WALTZ[2]: We use the x-ray exclusively. I would like to say that on the last order going through the foundry, the foundry insisted on the x-ray and not the customer. Future orders of intricate castings that do not call for x-ray will be x-rayed for our own benefit. That has been our experience. We have made thousands of x-rays and I do not believe one of them was not worth the price that we paid for it.

JAMES R. HEWITT, JR.[3]: The usefulness of radiographic inspection, either by x-ray or gamma radiation, as a foundry tool is becoming more generally recognized, but general acceptance in the foundry industry has been relatively slow, primarily because executives have found it difficult to justify the seemingly large expenditure for x-ray or gamma ray equipment.

Approximately three and one-half years ago, the firm with which I am associated, started an intensive program of process and quality control. A very complete x-ray laboratory was installed as an adjunct to existing testing equipment. During this period the x-ray laboratory, supplemented by other laboratory testing equipment, has served not as a routine inspection instrument, but as a pilot inspection and control tool, and most effectively as an educational instrument for instruction of the foundry personnel in improved heading and gating practices, molding technique and sand control.

The economic value and effectiveness of radiography as a foundry instrument has been amply justified in our own case, as testified by a material reduction in returned materials and claims. This reduction has partially offset the initial investment in the x-ray laboratory and covered its operating expenses during the three and one-half year period.

I would like to transmit the thought that while reserves for defective castings are often considered a necessary expenditure in foundry operations, the reduction of such reserves may justify the investment in radiographic equipment.

We do not lay that entirely to radiography but we do insist that radiography has been a very material aid in reducing that loss that we have, or expense that we have had heretofore, in return materials and claims.

I would like to interject the idea that there is a spot where you may be able to figure some means of getting your x-ray equipment in without it costing you an exorbitant amount on your cost of production. Our installation cost us in the neighborhood of thirty thousand dollars and it costs us, roughly, twenty-five hundred dollars a year to operate it but the reduction we have made in our return material allowances over a three-year period has almost paid for that installation.

2 Assistant Superintendent of Foundries, General Electric Co., Schenectady, N. Y.
3 Metallurgist, American Manganese Steel Div., American Brake Shoe & Foundry Co., Chicago Heights, Ill.

W. J. JEFFRIES[4]: This very excellent survey conducted by Mr. Briggs'
committee deserves a lot of favorable consideration, particularly in view
of the development that has taken place since Dr. Lester made a similar
survey two years ago. The advancement in the short space of two years
is little short of marvelous, and I hope that, as Mr. Briggs proposed, his
committee will not wait for five years to make a second survey, because
in my opinion development will be far more rapid in the next two years
than it has been in the past two years.

4 Navy Department, Bureau of Construction, Washington, D.C.

The Formation of Graphite in Gray Iron

ALFRED BOYLES*, COLUMBUS, OHIO

1. The structure of gray iron is a subject of increasing interest to foundrymen, especially those phases of it which have to do with the size and distribution of the graphite flakes. Castings usually go into service with the structure which originated during solidification and a knowledge of the mechanism of freezing therefore is useful to those who are attempting to control physical properties. In a recent publication, the writer described some experiments relating to this subject. The present paper is an attempt to formulate a theory on the basis of that work, explaining the variation in the size of the graphite flakes in terms of composition gradients set up during the freezing process, and to relate such gradients to variations in sulphur and manganese together with the conditions under which the iron is melted.

THE MECHANISM OF FREEZING IN HYPOEUTECTIC ALLOYS

2. The following conclusions were drawn from a study of small melts quenched at various temperatures during slow cooling from the molten state:[1]

(1) Primary austenite freezes out in the form of dendrites which continue to grow down to the eutectic temperature.

(2) Crystallization of the eutectic liquid begins at centers which grow equally in all directions, forming a cell-like structure.

(3) Segregation takes place in two stages: (a) between the primary dendrites and the liquid, (b) from the crystallization centers of the eutectic outward into the boundaries of the cells.

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(4) Constituents formed during the freezing of the eutectic occupy the interstices of the dendrites. The graphite flakes and the phosphide eutectic thus are restricted by the size and distribution of the dendrites.

(5) Graphite flakes do not begin to form until the eutectic begins to freeze. As soon as the eutectic is completely frozen, the flake structure essentially is complete. The flakes grow radially from the crystallization centers of the eutectic outward into the surrounding liquid, resulting in a "rosette" or "whorl" formation as shown in Figs. 1 and 2.

3. These conclusions were based on a study of alloys containing from 3.00 to 3.50 per cent carbon and about 2.00 per cent silicon, cooled at the rate shown in Fig. 3. The number of primary dendrites formed obviously depends on the carbon content, i. e. the position of the alloy in the iron-carbon-silicon system determines the amount of primary austenite precipitated during freezing (Fig. 4). The size of the dendrites depends entirely on the rate of cooling. As the carbon is lowered, the dendrites become more numerous and lie closer together but the spacing of the branches remains the same.

4. This is an important consideration because in low carbon irons the spacing of the dendritic branches determines the spacing of the graphite flakes, inasmuch as these are confined to the interstices. Additions of the minor elements, phosphorus, manganese and sulphur, to pure alloys of iron, carbon and silicon have no effect on the dendritic spacing. Superheating over the range from 2650 to 3050°F. has no observable effect.

Primary Dendrite Formation

5. Primary dendrites form in families having numerous parallel branches (Fig. 5). Each family constitutes the skeleton of a crystal which might be called the original "grain" in the casting. In Fig. 5, the boundaries of the various dendritic families are sufficiently distinct to reveal the "primary grain size" of the test bar. Such dendritic skeletons are never completed because a new mechanism intervenes, namely, the freezing of the eutectic.

Freezing Mechanism of Eutectic

6. This is apparently an entirely independent phenomenon
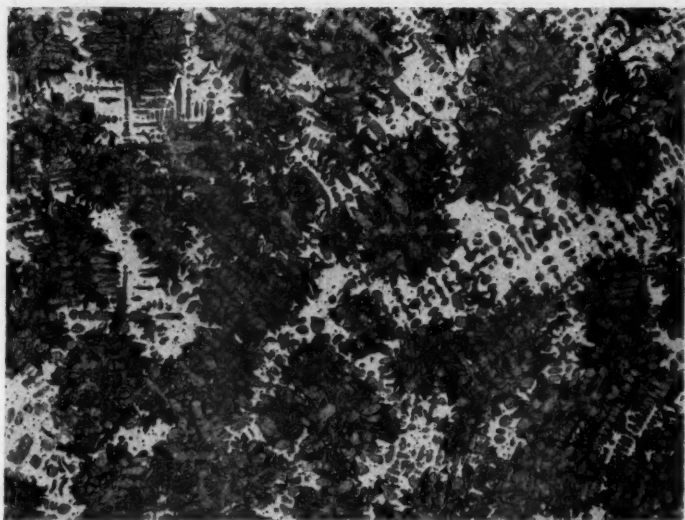
FIG. 1—HYPOEUTECTIC ALLOY QUENCHED FROM 2025°F. HEAT-TINTED, 20x.

FIG. 2—DETAIL OF STRUCTURE SHOWN IN FIG. 1. HEAT-TINTED, 100x.
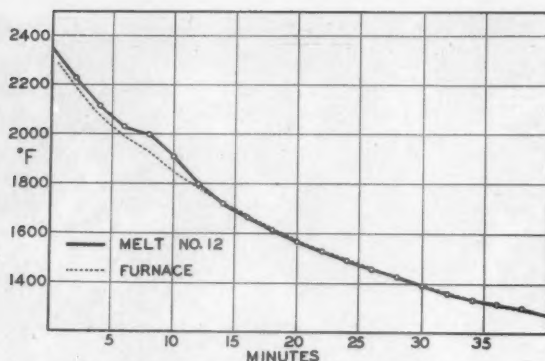
FIG. 3—COOLING CURVE OF THE GLOBAR FURNACE.

which begins at centers having no particular relation to the
dendritic families, or to their size. Each center grows equally
in all directions insofar as obstruction by the dendritic skeletons
will permit (Figs. 1 and 2). The result is a cell-like formation,
into the boundaries of which are concentrated those impurities
which segregate during the freezing of the eutectic. In hypo-
eutectic gray iron, the phosphide eutectic commonly is found in
the cell boundaries. In iron-carbon-silicon alloys containing
sulphur but no manganese, iron sulphide is likewise concentrated
into the cell boundaries. (Fig. 31).

Cell Formation

7. The size of the cells is determined by the total number
of crystallization centers formed in the eutectic. The rate at
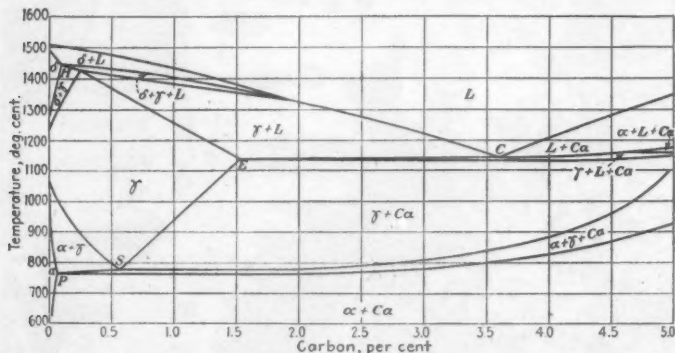
FIG. 4—IRON-CARBON DIAGRAM IN PRESENCE OF 2.00 PER CENT SILICON (GREINER, MARSH
AND STOUGHTON).

which new centers form during freezing obviously is influenced by the rate of cooling. The same cooling rate, however, does not always produce the same cell size in a given alloy when it is melted under different atmospheres. Ladle additions also seem to increase the number of crystallization centers in the eutectic and thus produce a small cell size. These eutectic cells
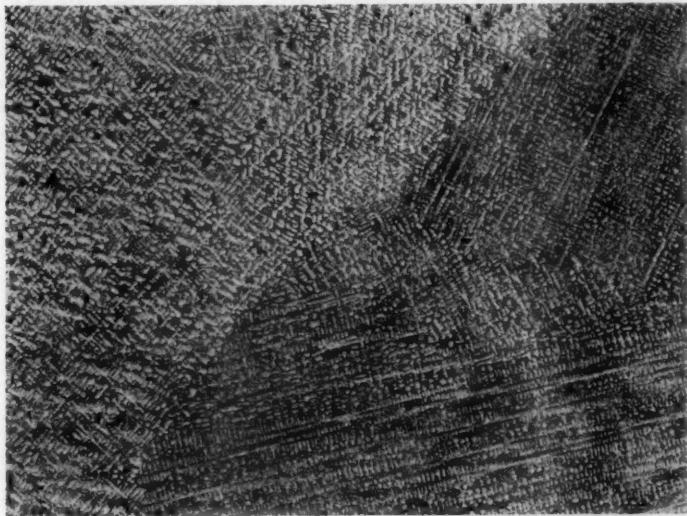
FIG. 5—TRANSVERSE TEST BAR OF HYPOEUTECTIC CAST IRON. ANALYSIS: CARBON 3.08 PER CENT, SILICON 2.16 PER CENT, MANGANESE 0.65 PER CENT, SULPHUR 0.029 PER CENT, PHOSPHORUS 0.06 PER CENT, NICKEL 2.09 PER CENT, CHROMIUM 0.17 PER CENT, MOLYBDENUM 1.05 PER CENT. ETCHED IN HNO_3, 20x.

in cast iron cannot be called "grains" in the strict sense because at no time in the thermal history do they consist of single crystals.

Flake Growth

8. On the contrary, each cell is developed by a colony of graphite flakes growing radially from a crystallization center. As each flake grows, it forms about itself a layer of solid austenite, which is one of the constituents of the eutectic. Fig. 2 illustrates this stage of development. Any carbon added to the sides of the flake must diffuse through the layer of solid austenite. Thus, we see that most of the growth must occur at one end of the flake, the end that projects out into the remaining liquid.

Under such conditions the growth is dependent on two factors: (1) the rate of solidification and (2) the rate at which carbon becomes available.

9.   The growth of a flake will terminate whenever its supply of carbon is insufficient to keep pace with the advancing front of solidification. As the flakes, shown in Fig. 2, grow out into the remaining liquid and become more widely separated, new centers of graphitization inevitably must spring up in the spaces between them. The availability of carbon to any particular flake is limited by the total number of graphitization centers so formed along the solid-liquid interface. A strong urge to graphitize therefore, will favor the formation of small flakes by increasing the rate of nucleization. The ultimate condition of this sort would be a true "graphite eutectic" which might be defined as a system in which the rate of dissociation is "infinitely rapid."

10.   On the other hand, the availability of carbon may be limited by a slow rate of dissociation. The growth of the flakes then lags behind the advancing front of solidification. The flakes are pinched off, so to speak, by the freezing of the metal and as a result small flakes are formed. Such cases are approaching graphitization in the solid state and the fine graphite so produced differs in appearance from that formed by rapid graphitization. Between the two extremes of very rapid and very slow carbide dissociation, a great range of flake sizes may be produced with the same rate of cooling. Somewhere in this series, the flakes will have a maximum size.

THE EFFECT OF SULPHUR

Melting Procedure

11.   The mechanism postulated above, was based on the effect of sulphur additions to alloys of iron, carbon and silicon, which briefly may be reviewed as follows: Electrolytic iron was melted with the proper amount of graphite and silicon carbide to give alloys of the composition listed in Table 1. These were cast into standard transverse test bars from which 50 gram

Table 1
COMPOSITION OF IRON, CARBON, SILICON ALLOYS

Alloy No.	Total C.	Si.	S.	Mn.	P.
7	3.06	1.91	0.007	Nil	0.010
8	3.49	2.01	0.007	Nil	0.008

charges were machined to fit the small crucibles used in re-
melting. A vertical-tube, globar furnace was employed and the
heating and cooling cycle standardized as follows: Each melt
was brought up to 2500°F. in a period of 70 min. and held at
this temperature for 10 min. after which the power was cut off
and the furnace allowed to cool at its natural rate down to
below 500°F. before the melt was removed. Fig. 3 shows the
cooling rate.

12. Additions were made in the following manner: Several
small holes were drilled into the interior of the charge and the
element to be added was placed in the bottom of these holes.
Chips taken from the holes were then placed above the added
element and driven in very tightly until the holes were com-
pletely filled.

Structures

13. The structure of the bars of alloy no. 8 as cast was
highly abnormal and this persisted after remelting in the globar
furnace in a nitrogen atmosphere, as shown in Fig. 6, which
consists mainly of ferrite containing very fine graphite flakes.
A series of quenching tests showed that this alloy froze in pre-
cisely the same manner as cast iron. It differed, however, in
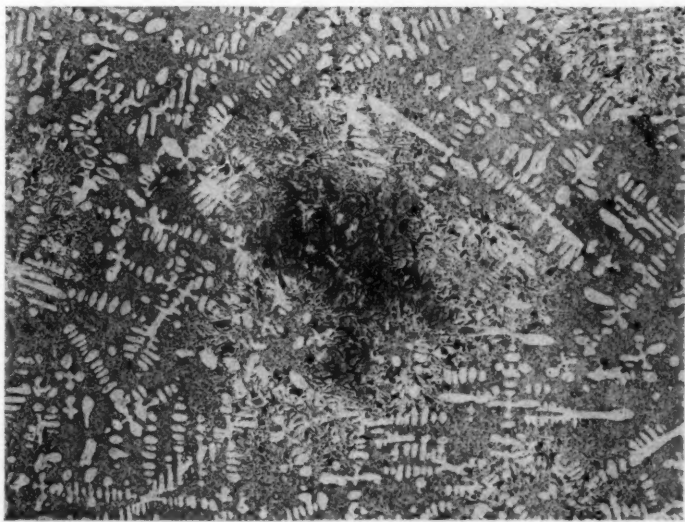
FIG. 6—IRON-CARBON-SILICON ALLOY NO. 8 ETCHED IN HNO₃, 20X.

the extent to which it graphitized during the quench, indicating that the urge to graphitize was very much greater than in cast iron.

14. An addition of sulphur produced a very marked increase in the size of the graphite flakes and rendered the matrix fully pearlitic, as shown in Fig. 7. A graded series of sulphur additions was made giving the compositions listed in Table 2. These were melted in nitrogen. The resulting structures of alloy no. 8 are shown in Figs. 8 to 13. The size of the flakes increases as the sulphur is raised up to a certain point after which they became smaller again. Alloy No. 7 showed a similar behavior. All the melts were fully pearlitic but no trace of mottle could be found in any of them.

Table 2

IRON-CARBON-SILICON ALLOYS WITH SULPHUR ADDITIONS

Alloy No. 8; Carbon, 3.49 Per Cent, Alloy No. 7; Carbon, 3.06 Per Cent, Silicon 2.01 Per Cent   Silicon 1.91 Per Cent

Melt No.	Sulphur, Per Cent	Brinell Hardness	Melt No.	Sulphur, Per Cent	Brinell Hardness
27	0.018	200	46	0.026	216
28	0.024	195	47	0.100	225
24	0.076	190	48	0.146	228
52	0.161	200	49	0.194	228
53	0.250	205	50	0.282	222
54	0.294	205			

Quenching Experiments

15. A series of quenching experiments was then made on alloy no. 8 plus an addition of 0.1 per cent sulphur, which demonstrated that the mechanism of freezing was the same as that previously observed. The urge to graphitize, however, was much less than in the original alloy, as shown by the small amount of graphite formed during the quench. From the well known action of sulphur in increasing the stability of the carbide in cast iron, it was reasoned that a similar effect had occurred in the iron-carbon-silicon alloy and that graphitization during freezing became slower as the sulphur was raised. A relationship between carbide stability and graphite flake size therefore, was, postulated as outlined above. This is most readily shown in the form of a diagram such as Fig. 14.
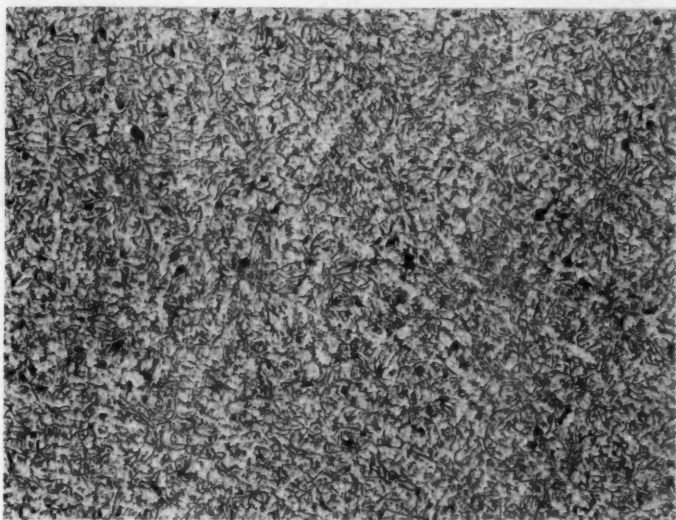
FIG. 7—SAME ALLOY SHOWN IN FIG. 6 WITH AN ADDITION OF 0.076 PER CENT SULPHUR. ETCHED IN HNO₃, 20X.
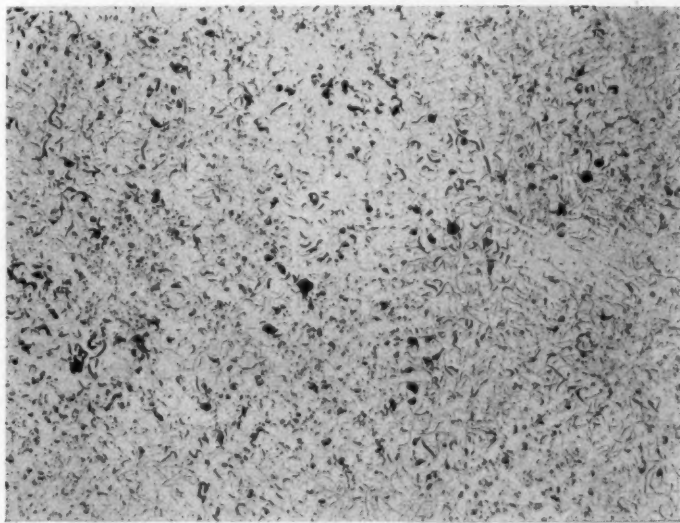
FIG. 8—ALLOY NO. 8 + 0.018 PER CENT SULPHUR. MELTED IN NITROGEN. UNETCHED, 20X.
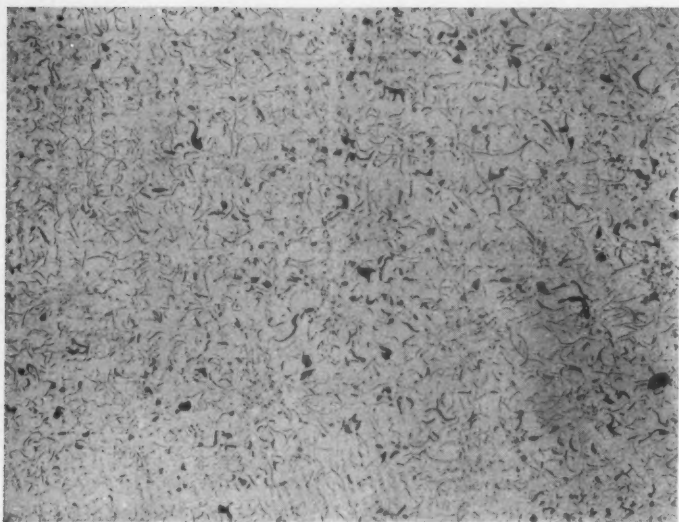
FIG. 9—ALLOY NO. 8 + 0.024 PER CENT SULPHUR. MELTED IN NITROGEN. UNETCHED, 20X.
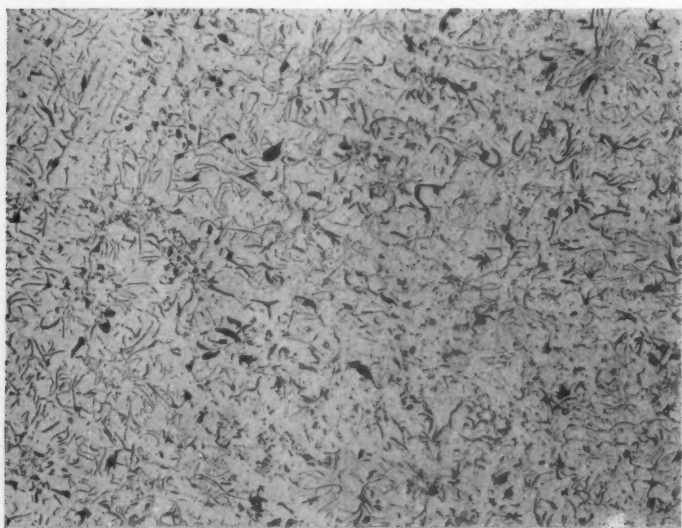
FIG. 10—ALLOY NO. 8 + 0.076 PER CENT SULPHUR. MELTED IN NITROGEN. UNETCHED, 20X.
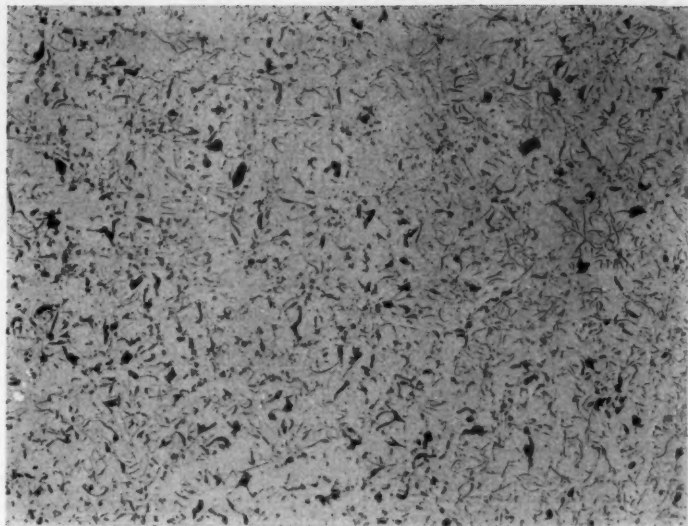
FIG. 11—ALLOY NO. 8 + 0.161 PER CENT SULPHUR. MELTED IN NITROGEN. UNETCHED, 20X.
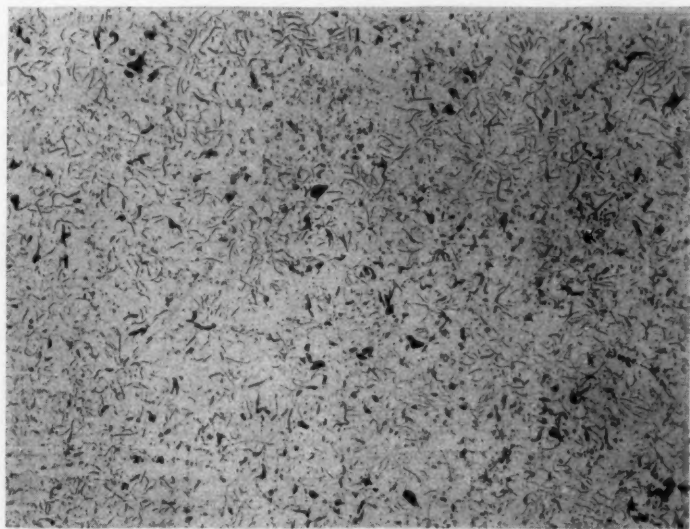
FIG. 12—ALLOY NO. 8 + 0.250 PER CENT SULPHUR. MELTED IN NITROGEN. UNETCHED, 20X.
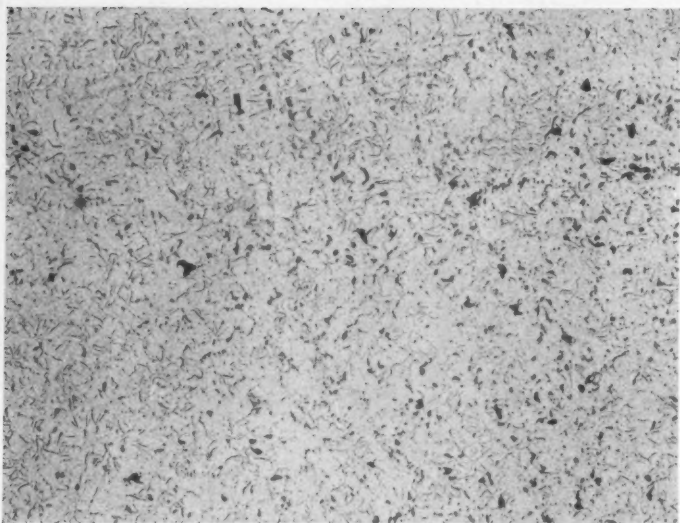
FIG. 13—ALLOY NO. 8 + 0.294 PER CENT SULPHUR. MELTED IN NITROGEN. UNETCHED, 20X.
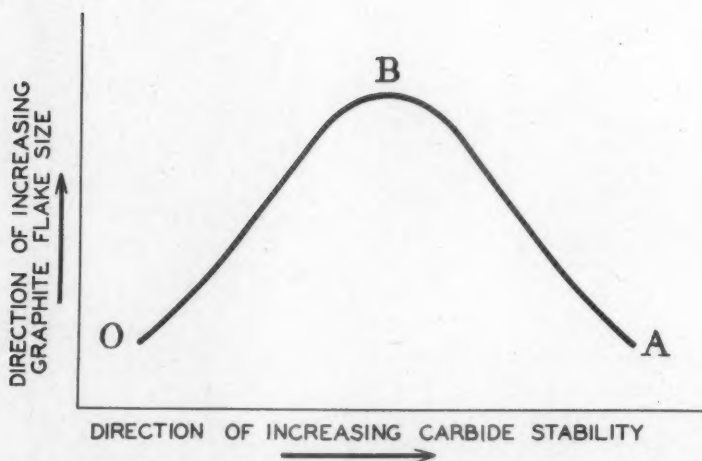
FIG. 14—DIAGRAMMATIC REPRESENTATION OF THE RELATIONSHIP BETWEEN FLAKE SIZE
AND RATE OF GRAPHITIZATION.

16. In the original iron-carbon-silicon alloy, the urge to graphitize was very intense and it might be placed on Fig. 14 near "O." As the sulphur was raised, the carbide stability increased and the flake size reached a maximum at point "B," after which still greater amounts produced a decrease in flake size as the system approached point "A." Carried sufficiently far this would result in a white iron. The sulphur additions were not increased to a point sufficient to do so but the same thing was accomplished in another way.

THE EFFECT OF HYDROGEN ON CARBIDE STABILITY

17. In the course of other experiments, it was observed that certain alloys which are gray normally, freeze white when they are melted in hydrogen and slowly cooled. Resulting white iron is restored to its original condition by melting in air or rendered still softer by melting under reduced pressure. Such changes are attributed to an absorption of hydrogen by the iron, the extent of which is governed by the amount of hydrogen in contact with the molten metal, by the pressure in the furnace and by the temperature, inasmuch as the solubility of hydrogen increases as the temperature is raised.

Effects of Sulphur Plus Hydrogen

18. In order to investigate the combined effects of sulphur and hydrogen, each of the series of alloys listed in Table 2 was remelted in hydrogen at atmospheric pressure, held ten minutes at 2500° F. and slow cooled as usual. The resulting structures are shown in Figs. 15 to 20. The maximum flake size now occurs in melt no. 27 which contained 0.018 per cent sulphur (Fig. 15). With increasing sulphur, the flakes rapidly diminish in size and in melt no. 24 white areas begin to appear (A in Fig. 17). These increase in prominence until they occupy about one third of the entire structure as shown in Fig. 21. The graphite in the last two melts is very finely divided and has lost its flakelike form (Figs. 19 and 20). No free ferrite occurs in any of these melts. All of them contain numerous cavities such as those in Figs. 15 and 21.

19. Referring to Fig. 14, it is seen that melting in hydrogen has caused the whole series of alloys to shift along the curve so that all the structures now lie between "B" and "A." In the original condition, a maximum flake size occurred with 0.076 per cent sulphur. The maximum is now found in a sample which contained 0.018 per cent sulphur, part of which was undoubtedly lost in

FIG. 15—SAME METAL SHOWN IN FIG. 8 REMELTED IN HYDROGEN. UNETCHED, 20X.
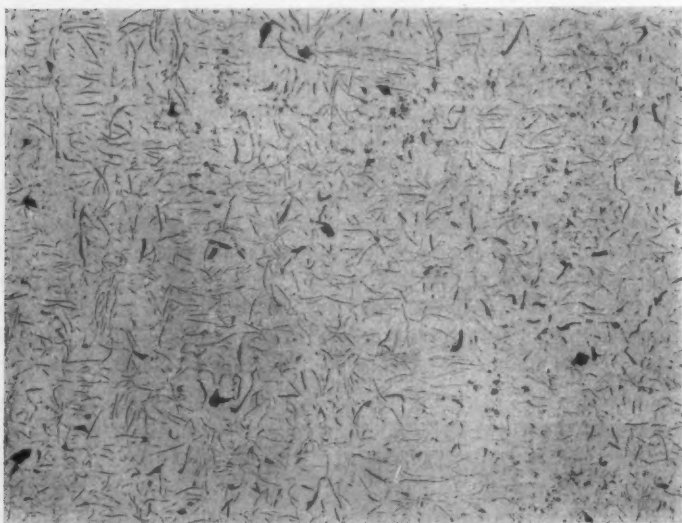
FIG. 16—SAME METAL SHOWN IN FIG. 9 REMELTED IN HYDROGEN. UNETCHED, 20X.
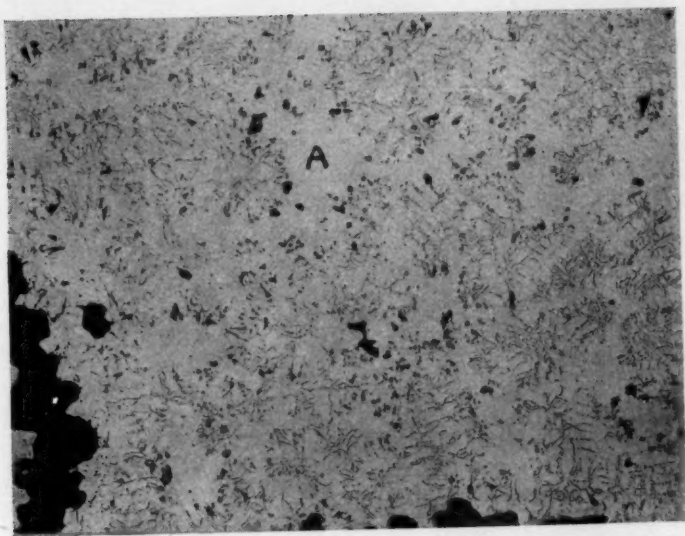
FIG. 17—SAME METAL SHOWN IN FIG. 10 REMELTED IN HYDROGEN. UNETCHED, 20X.
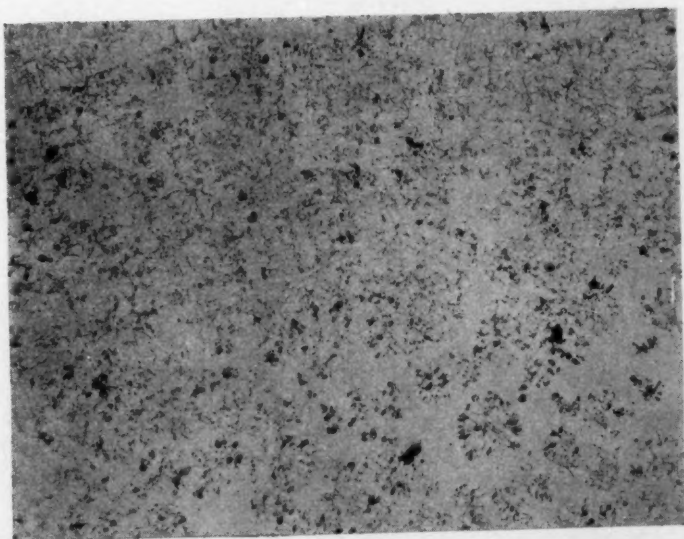
FIG. 18—SAME METAL SHOWN IN FIG. 11 REMELTED IN HYDROGEN. UNETCHED, 20X.
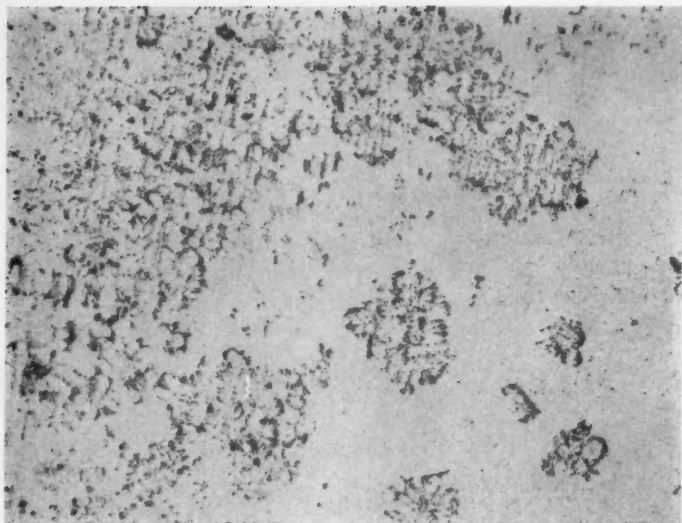
FIG. 19—SAME METAL SHOWN IN FIG. 12 REMELTED IN HYDROGEN. UNETCHED, 20X.
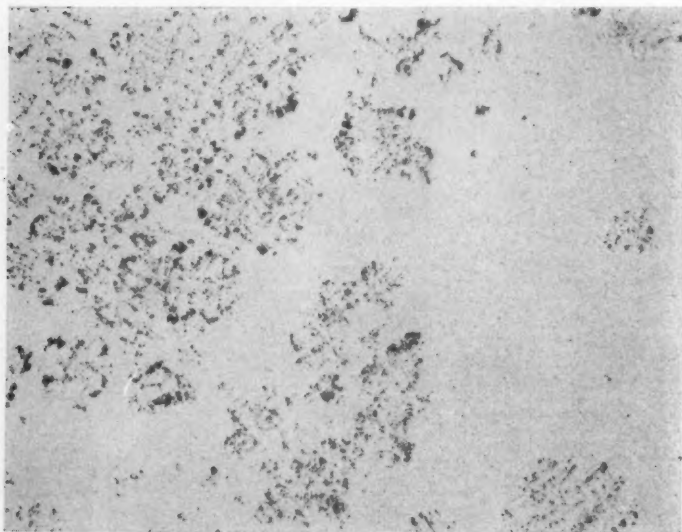
FIG. 20—SAME METAL SHOWN IN FIG. 13 REMELTED IN HYDROGEN. UNETCHED, 20X.
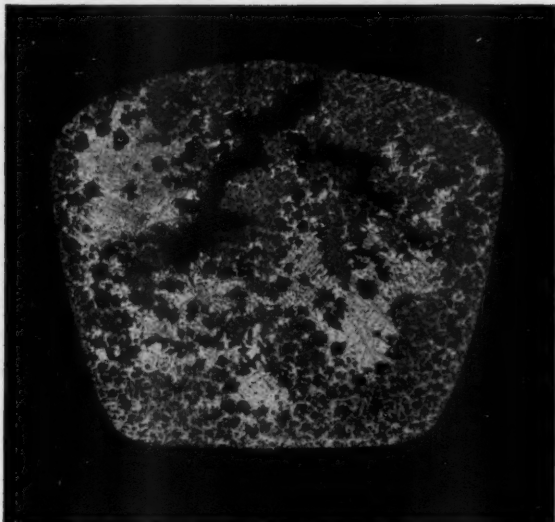
FIG. 21—ALLOY NO. 8 + 0.294 PER CENT SULPHUR MELTED IN HYDROGEN. ETCHED IN HNO₃, 3X.
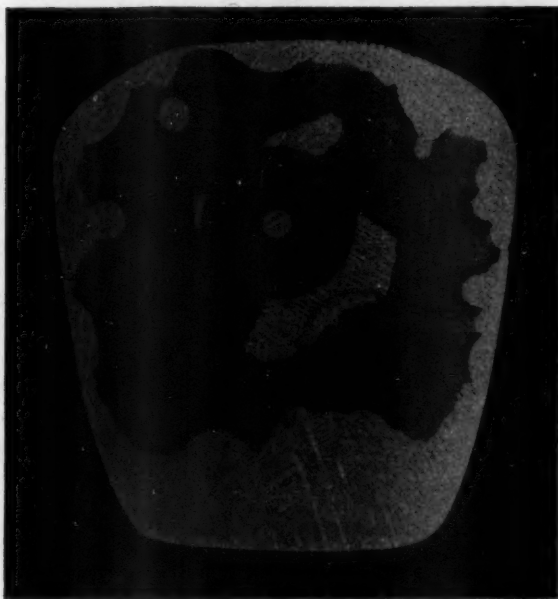
FIG. 22—ALLOY NO. 8 MELTED IN HYDROGEN. THE OUTER PORTION IS ALL FERRITE. ETCHED IN HNO₃, 3X.

remelting in hydrogen. An increase in carbide stability is indicated
by the fact that all the melts above 0.076 per cent sulphur are now
mottled, whereas before they were gray. Hydrogen and sulphur
therefore, may be looked upon as acting in the same direction, both
tending to stabilize the carbide.

20. By remelting the samples under reduced pressure, it
should be possibile to remove some of the hydrogen and restore the
original condition or something analogous to it. This was done,
using a pressure of about 2 mm. of mercury, the melts being held
10 min. at 2500° F. and slow cooled as usual. Analyses were then
made to determine the loss of sulphur caused by the double melting,
with the results given in Table 3. The structures are shown in
Figs. 23 to 28.

Table 3

ALLOY No. 8 WITH SULPHUR ADDITIONS

Analysis, Per Cent S	Melted in Nitrogen, See Fig. No.	Melted in Hydrogen, See Fig. No.	Melted in Vacuo, See Fig. No.	Final Analysis, Per Cent S
0.018	8	15*	23	0.012
0.024	9	16	24	0.016
0.076	10*	17	25	0.046
0.161	11	18	26	0.099
0.250	12	19	27*	0.139
0.294	13	20	28	0.172

* Indicates maximum flake size for each series.

Microstructures

21. Melt no. 27, with 0.012 per cent sulphur, shows a struc-
ture resembling the original iron—carbon—silicon alloy (compare
Fig. 23 with Fig. 6). Only a few traces of pearlite occur in the
matrix, which consists of ferrite with very fine graphite flakes. It
should be noted that primary dendrites are more numerous than
before, due to a loss of carbon in the double melting. The spacing
of the dendritic branches is exactly the same as before because the
melts were cooled at the same rate.

22. In melt no. 28, with 0.016 per cent sulphur, considerable
pearlite occurs at the boundaries of the eutectic cells (Fig. 24).
A few large flakes are also present in the cell boundaries.

23. Melt no. 24, with 0.046 per cent sulphur, shows a marked
increase both in the amount of pearlite and the number of large
flakes occurring at the cell boundaries (Fig. 25).

24. In melt no. 52, with 0.099 per cent sulphur, only a small

FIG. 23—SAME METAL SHOWN IN FIG. 15 REMELTED IN VACUO (0.012 PER CENT SULPHUR). ETCHED IN HNO₃, 20X.
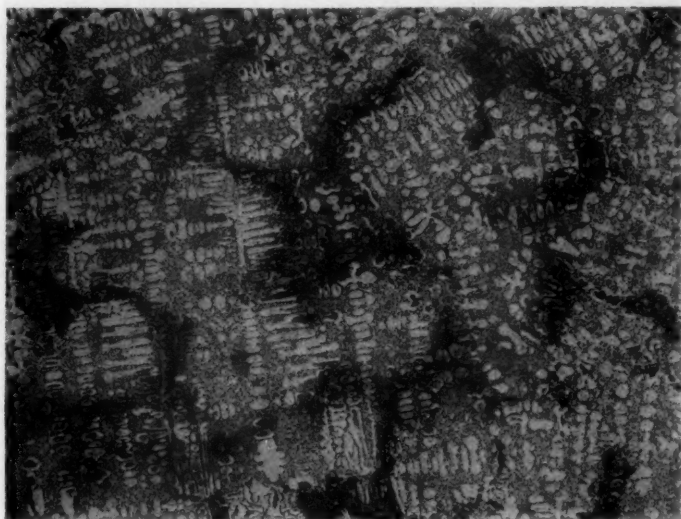
FIG. 24—SAME METAL SHOWN IN FIG. 16 REMELTED IN VACUO (0.016 PER CENT SULPHUR). ETCHED IN HNO₃, 20X.
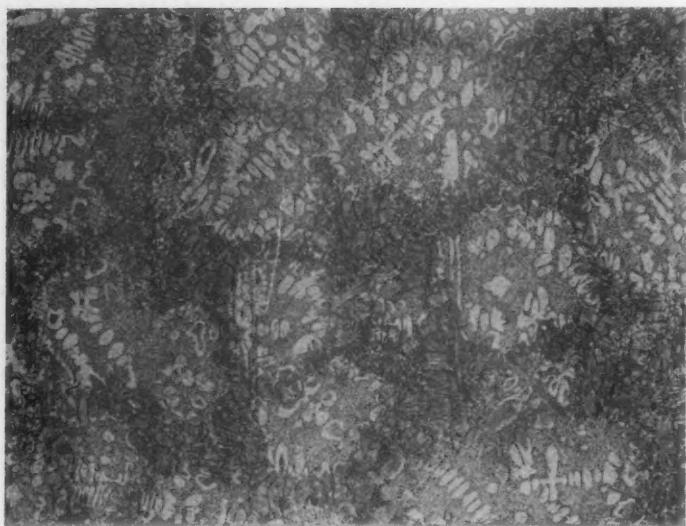
FIG. 25—SAME METAL SHOWN IN FIG. 17 REMELTED IN VACUO (0.046 PER CENT SULPHUR).
ETCHED IN HNO₃, 20X.
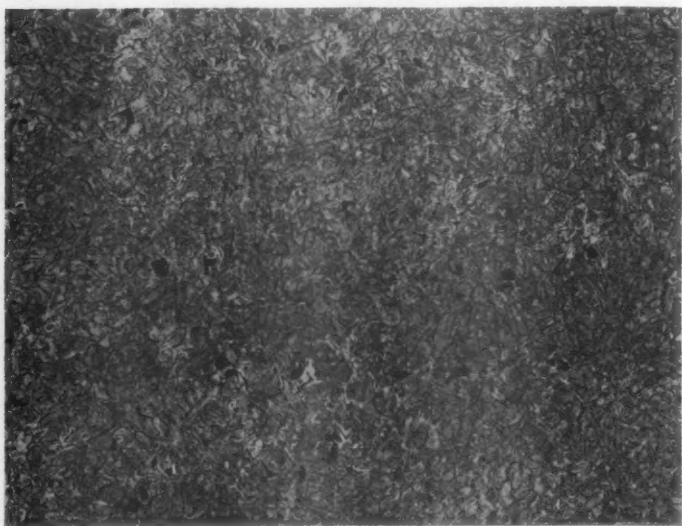
FIG. 26—SAME METAL SHOWN IN FIG. 18 REMELTED IN VACUO (0.099 PER CENT SULPHUR).
ETCHED IN HNO₃, 20X.

FIG. 27—SAME METAL SHOWN IN FIG. 19 REMELTED IN VACUO (0.139 PER CENT SULPHUR). ETCHED IN HNO₃, 20X.
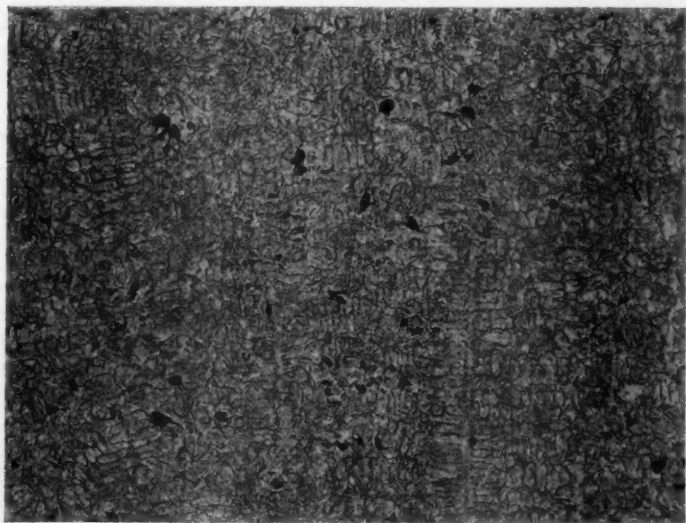
FIG. 28—SAME METAL SHOWN IN FIG. 20 REMELTED IN VACUO (0.172 PER CENT SULPHUR). ETCHED IN HNO₃, 20X.

amount of ferrite occurs and the cell boundaries are no longer visible. Most of the graphite is in the form of medium sized flakes (Fig. 26).

25. In the last two melts, containing 0.139 per cent and 0.172 per cent sulphur respectively, the structure is practically normal. The maximum flake size occurs in melt no. 53 (Fig. 27) while the least amount of ferrite is found in no. 54 (Fig. 28).

Dendritism and Flake Size

26. This series of specimens shows very clearly how the visibility of the primary dendrites is dependent on the size of the graphite which lies between them. In all cases, dendrites of the same size formed during solidification. The extent to which they are visible in the final structure, *i. e.* the completeness with which they are outlined, increases as the graphite becomes finer. Dendritism is least noticeable where the flake size is a maximum (Fig. 27).

27. Referring again to Fig. 14 and Table 3, it will be seen that the maximum flake size now occurs with 0.139 per cent sulphur in a melt that was previously mottled. Removing hydrogen therefore, has decreased the stability of the carbide and shifted the whole series on the curve toward the point "*O*." What happened in this whole experiment may be recapitulated as follows:

28. The original melts, made in nitrogen, picked up a certain amount of hydrogen from the furnace atmosphere, the most likely source being moisture. This, together with the sulphur added, produced varying degrees of carbide stability, giving rise to a series of structures lying on either side of point "*B*" in Fig. 14. When the same pieces were remelted in hydrogen, the carbide stability was increased sufficiently to throw the whole series toward the side of the curve marked "*A*", the effect still being proportional to the sulphur content.

29. Upon remelting and freezing in vacuo, the carbide stability was decreased by a removal of hydrogen and the series was shifted over the curve toward the side "*O*." Although some sulphur was lost in the double melting, the structural change still is proportional to the sulphur content. The manner in which hydrogen and sulphur interact in these cases is a matter of considerable interest and merits some discussion.

THE SEGREGATION OF SULPHUR

30. In the absence of manganese, sulphur segregates in a

very definite manner. The solubility of iron sulphide in iron changes abruptly at the freezing point. In liquid iron the sulphide is quite soluble, whereas in solid iron its solubility is low[2]. A major precipitation of iron sulphide therefore, will occur during solidification.

31. As the primary dendrites grow, sulphur is concentrated into the remaining liquid which is thereby enriched to an extent governed by the amount of primary austenite formed, in other words, by the carbon content. When the freezing of the eutectic begins, the sulphide in solution is further concentrated into the liquid remaining around the rosettes. A precipitation of inclusions of iron sulphide finally occurs in the boundaries of the cells. Such a distribution of sulphur is readily demonstrated by means of sulphur printing (Figs. 30 and 31).

Interpretation of Structures

32. With this mechanism in mind, some interpretation of the structures shown in Figs. 23 to 28 may be made. Here the sulphur may be considered as acting in its own right, the hydrogen having been removed by vacuum melting.

33. It will be assumed that a certain concentration of iron sulphide is needed along the solid-liquid interface to convert fine graphite, like that in Fig. 6, into normal flakes. As the eutectic in Fig. 23 froze, segregation failed to produce a concentration
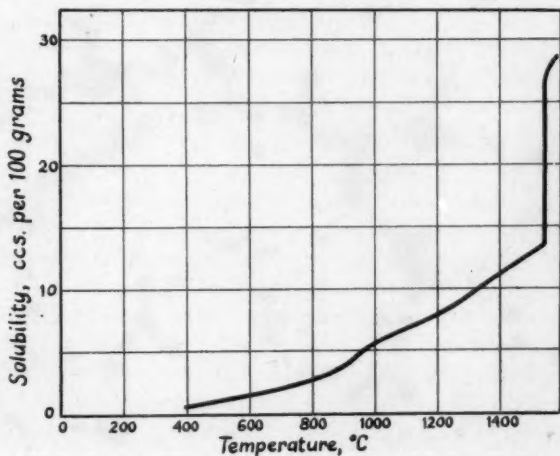
FIG. 29—SOLUBILITY OF HYDROGEN IN IRON (AFTER SIEVERTS).
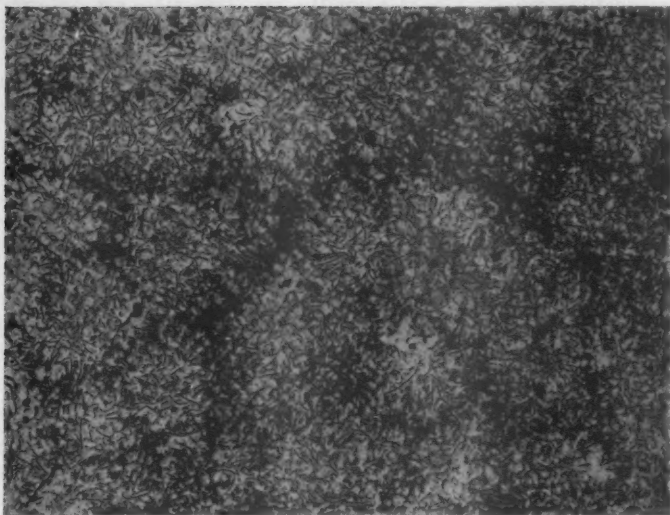
FIG. 30—IDENTICAL FIELD SHOWN IN FIG. 7. HEAT-TINTED TO SHOW THE CELL BOUNDARIES.
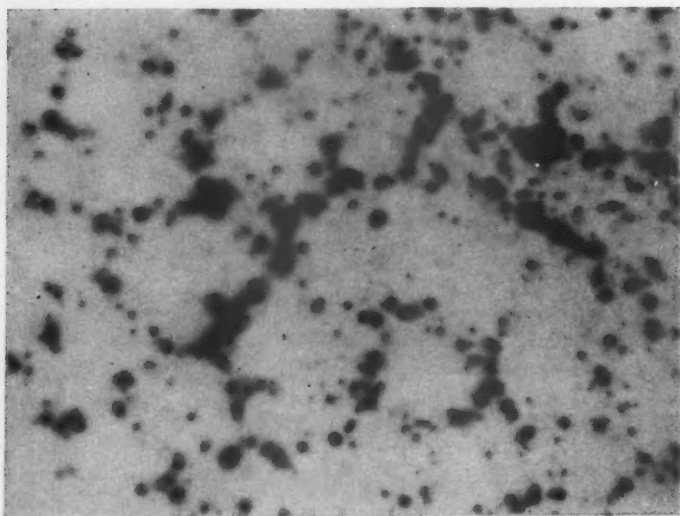
FIG. 31—SULPHUR PRINT FROM THE SPOT SHOWN IN FIG. 30. ENLARGED 20 DIAMETERS.

at any point sufficient to bring about this change because the total sulphur content was too low to begin with. In Fig. 24, where the sulphur is slightly higher, a noticeable change has occurred in the boundaries of the cells, the points of maximum segregation. With still higher sulphur, in Fig. 25, the boundary regions containing the normal flakes are much wider, as would be expected. In Fig. 26, they have become so broad as to embrace most of the structure. The same conditions exist in Fig. 27 and Fig. 28. Here the sulphur concentration along the solid-liquid interface was sufficient to produce a normal flake structure over the whole melt.

34. The behavior of hydrogen during freezing only can be surmized from what is known of its action with pure iron. According to Sieverts[3], the solubility of hydrogen in iron changes with temperature as shown in Fig. 29. An abrupt change occurs at the melting point. Therefore, hydrogen would be expected to segregate during freezing in the same manner as sulphur because its solubility in the solid is less than in the liquid state.

Hydrogen During Solidification

35. As primary dendrites form, the hydrogen in solution is no doubt concentrated into the eutectic liquid and as the eutectic freezes a liberation of the gas must occur whenever its limit of solubility in the remaining liquid is exceeded. If the initial concentration is high enough, a liberation may begin before the eutectic starts to freeze. This is evidenced by the presence of numerous interdendritic cavities in melts made under hydrogen (Fig. 15). A button of eutectic liquid is often extruded on top or, in cases where this does not happen, the whole melt may burst (Fig. 21).

36. Several things should be noted in Fig. 21. The white areas occur at the boundaries of the eutectic cells and, in addition, there is a general concentration of eutectic carbide in the central part of the melt. The first circumstance may be looked upon as a result of the segregation of sulphur into the cell boundaries plus the stabilizing influence of hydrogen. Sulphur printing, however, does not show any general concentration of sulphur into the large white areas. These may possibly be due to a higher hydrogen content in the central part of the melt as a whole, brought about by the change in solubility during freezing. After a solid crust had formed, the hydrogen was trapped and finally developed enough pressure to burst the melt as shown.

Sulphur and Hydrogen Effects Not Equivalent

37. The fundamental cause of the stabilizing effect of hydrogen is not known. Pressure effects, acting through the mechanism just described, are a possibility. Apparently its action is closely linked with that of sulphur but nothing can be said as to the exact manner in which the two act on the carbide.

38. It should be noted that, in the melts just described, the effect of the two elements is not precisely equivalent. For example, the structure produced by a low sulphur and high hydrogen (Fig. 15) is not exactly like that produced by a high sulphur and low hydrogen (Fig. 27). It also should be mentioned that hydrogen, in the absence of sulphur, does not have any great effect on the iron—carbon—silicon alloy (Fig. 22) but that sulphur does have an effect in the absence of hydrogen, as shown in the melts made in vacuo. It is not certain, however, that the vacuum melts are entirely free from hydrogen. The amount held in the metal under reduced pressure may be proportional to the sulphur content.

39. In other experiments, where the amount of sulphur was held constant and the hydrogen varied by melting under various pressures and temperatures, it was found that the graphite flakes could be varied in a manner precisely analogous to that shown in Fig. 14[1].

THE EFFECTS OF MANGANESE

40. When manganese is added to an iron—carbon—silicon alloy containing both sulphur and hydrogen, the system becomes very complex. The writer will not venture to describe all that transpires during the freezing of such an alloy, but the following series of melts will illustrate some of the possible variations in structure.

41. A series of melts was made under hydrogen using alloy no. 7 (Table 1). A constant amount of sulphur (0.15 per cent) and a varying amount of manganese (Baker's C. P. Metal) was added, and the alloys were held 10 min. at 2500° F. and slow cooled as usual. To interpret the resulting structures, it is necessary to consider the manner in which the sulphur and manganese were added to the charge.

42. Four small holes were drilled in the charge, reaching nearly to the bottom. Sulphur was placed in these holes and pow-

dered manganese metal was put on top of the sulphur, after which the holes were rammed tightly with chips.

43. Long before the alloy melted, a reaction must have occurred between the iron and sulphur and also between the manganese and sulphur. After fusion, some of the manganese sulphide floated to the top of the melt leaving the lower portion relatively rich in iron sulphide. This fact is shown very clearly by the color and shape of the inclusions. Those near the bottom are pale, yellowish-gray and have rounded forms, while near the top the inclusions are dove gray and have crystalline outlines.

44. Wohrman[2] has pointed out that FeS and MnS form a continuous series of solid solutions in which the melting point increases with increasing manganese content, ranging from about 2150 to 2950° F. FeS+MnS inclusions, which freeze earlier than the surrounding metal, are able to form idiomorphic crystals while those which freeze later are not able to do so. Wohrman also made melts of electrolytic iron which exhibited a variation in inclusions from top to bottom analogous to that found in the melts described below.

STRUCTURE OF THE MELTS

Melt No. 134 (Addition: No Mn., 0.15 per cent S.)

45. Fig. 32 shows a vertical section of the entire melt. The structure is uniform from top to bottom and consists of pearlite and very fine graphite, with eutectic carbide at the cell boundaries. Numerous interdendritic cavities occur.

Melt No. 135 (Additions: Mn., 0.20 per cent; S., 0.15 per cent)

46. Fig. 33 shows the entire melt. The bottom third has a structure like the preceding melt. Eutectic carbide is not present in the middle and upper parts. Near the top graphite appears as fine flakes. Several zones have been indicated alongside the photograph but these, of course, are not sharply defined, the transition from top to bottom being gradual.

Melt No. 136 (Additions: Mn., 0.40 per cent; S., 0.15 per cent)

47. Fig. 34 shows the whole melt, in which several entirely different types of structure occur as indicated alongside the photograph. A ferritic layer has appeared at the very top of the melt.

VERY FINE GRAPHITE,
PEARLITE AND
EUTECTIC CARBIDE

FIG. 32—MELT NO. 184—ADDITIONS: SULPHUR, 0.15 PER CENT; MANGANESE, NIL.
ETCHED IN HNO₃, 3X.
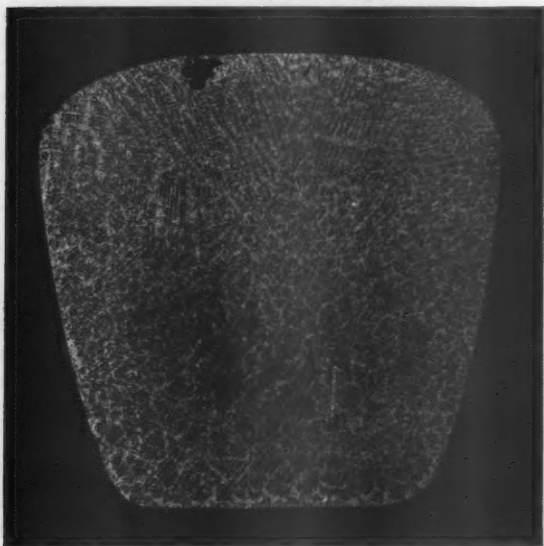
FINE FLAKES
AND PEARLITE

VERY FINE GRAPHITE
AND PEARLITE

VERY FINE GRAPHITE,
PEARLITE AND
EUTECTIC CARBIDE

FIG. 33—MELT NO. 185—ADDITIONS: SULPHUR, 0.15 PER CENT; MANGANESE, 0.20 PER CENT.
ETCHED IN HNO₃, 3X.
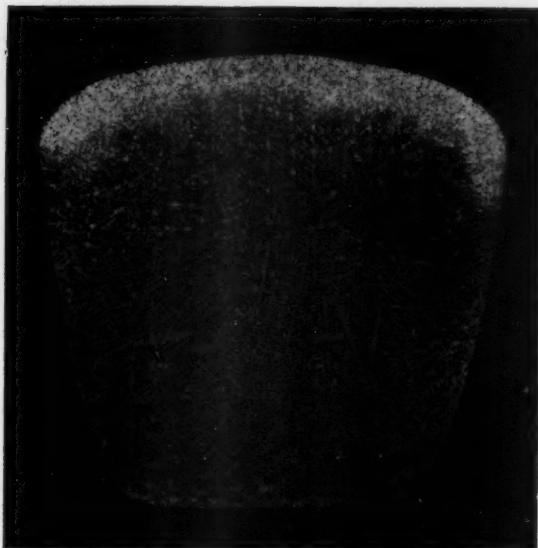
LARGE FLAKES,
PEARLITE AND
FERRITE

SMALL FLAKES
AND PEARLITE

FINE FLAKES
AND PEARLITE

VERY FINE GRAPHITE,
PEARLITE AND
EUTECTIC CARBIDE

FIG. 34—MELT NO. 136—ADDITIONS: SULPHUR, 0.15 PER CENT; MANGANESE, 0.40 PER CENT. ETCHED IN HNO₃, 3X.
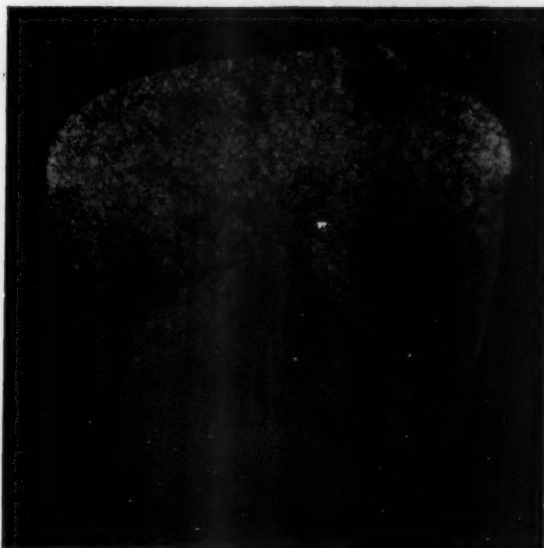
LARGE FLAKES,
PEARLITE
AND FERRITE

LARGE FLAKES
AND PEARLITE

SMALL FLAKES
AND PEARLITE

FIG. 35—MELT NO. 137—ADDITIONS: SULPHUR, 0.15 PER CENT; MANGANESE, 0.60 PER CENT. ETCHED IN HNO₃, 3X.

Melt No. 137 (Additions: Mn., 0.60 per cent; S., 0.15 per cent)

48.   Fig. 35 shows the entire melt and lists the various constituents present at different levels. There is no eutectic carbide in this melt. The flakes in the ferritic layer are quite large and it extends further down into the melt than previously.

Melt No. 138 (Additions: Mn., 0.90 per cent; S., 0.15 per cent)

49.   Fig. 36 shows the entire melt. The ferritic region is now near the middle of the melt. At the top is a fully pearlitic layer in which the flakes are somewhat smaller. The lower pearlitic layer is like that in melt no. 137 but the upper pearlitic layer is quite different, the pearlite lamellae being much finer. This is probably due to a higher manganese content in the top layer which lowered the transformation temperature.

Melt No. 139 (Additions: Mn., 1.20 per cent; S., 0.15 per cent)

50.   Fig. 37 shows the entire melt. The upper pearlitic layer is much thicker than in the preceding melt. The ferritic zone is quite narrow and sharply defined. A difference in the appearance of the pearlite is noticeable in Fig. 37. The upper layer has much finer pearlite lamellae than the lower. After examination was

MEDIUM FLAKES
AND PEARLITE

LARGE FLAKES,
PEARLITE AND
FERRITE

SMALL FLAKES
AND PEARLITE
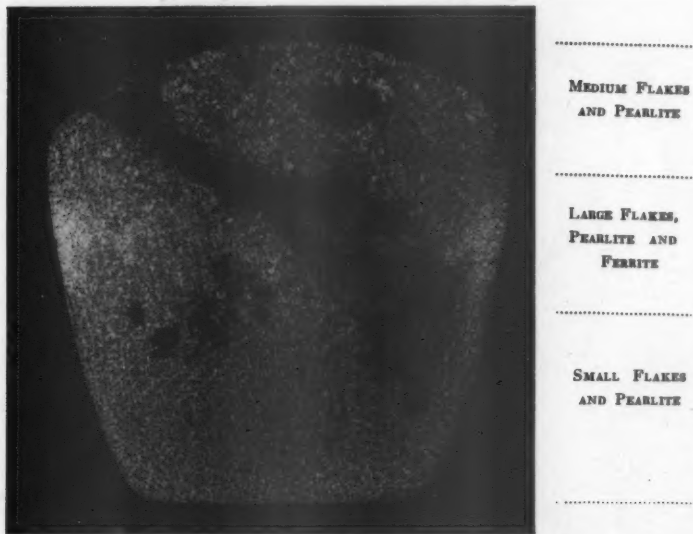
FIG. 36—MELT NO. 138—ADDITIONS: SULPHUR, 0.15 PER CENT; MANGANESE, 0.90 PER CENT. ETCHED IN HNO₃, 3X.
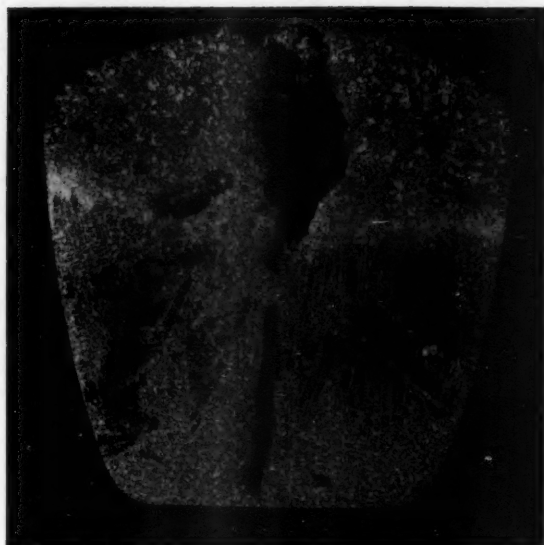
MEDIUM FLAKES
AND PEARLITE

LARGE FLAKES,
PEARLITE
AND FERRITE

SMALL FLAKES
AND PEARLITE

FIG. 37—MELT No. 189—ADDITIONS: SULPHUR, 0.15 PER CENT; MANGANESE, 1.20 PER CENT.
ETCHED IN HNO₃, 3X.

completed this melt was cut in half horizontally just below the
ferritic layer and samples for analysis were prepared from the
upper and lower parts with the following results:

Location	Per Cent Manganese	Per Cent Sulphur
Upper Half	1.27	0.16
Lower Half	0.14	0.06

DISTRIBUTION OF THE SULPHIDE

51. Sulphur prints were made on this series of melts in an
attempt to relate the distribution of sulphide to the variations
in structure. The resulting prints were enlarged three diameters
and reversed in copying so as to give a direct comparison with
the photographs showing the macrostructure. A number of prints
were made in succession on each specimen. A reproduction of
one of the sulphur prints will suffice for purposes of discussion
(Fig. 38).

Melt No. 134 (Addition: No Mn.; 0.15 per cent S.)

52. The iron sulphide is concentrated into the cell bound-
aries in the manner previously observed in melts containing no

manganese (Fig. 31). The white areas in the melt correspond quite accurately to the darker portions of the sulphur print. There is no difference from top to bottom.

Melt No. 135 (Additions: Mn.; 0.20 per cent; S., 0.15 per cent)

53.   In the lower part of the melt, the distribution is cellular, as in the preceding case. Dark portions in the sulphur print correspond to white areas in the melt. In the upper half, the sulphur distribution is random.

Melt No. 136 (Additions: Mn., 0.40 per cent; S., 0.15 per cent)

54.   A cellular distribution occurs near the bottom, which corresponds to the white areas present there. In the ferritic layer at the top, the distribution resembles the upper half of Fig. 38 and in the central part it is random.
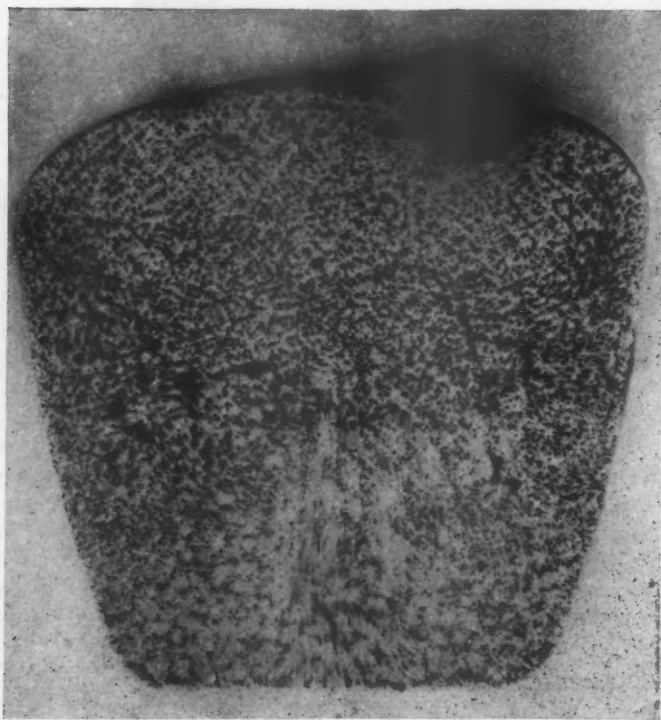
FIG. 38—SULPHUR PRINT FROM THE MELT SHOWN IN FIG. 35. ENLARGED 4 DIAMETERS.

Melt No. 137 (Additions: Mn., 0.60 per cent; S., 0.15 per cent)

55. A copy of this print is shown in Fig. 38. There is a very marked difference in the sulphur distribution in the upper and lower parts. Near the bottom, the sulphide has been concentrated into the boundaries of the cells, although to a lesser degree than previously. In the upper part, the distribution has been influenced by the primary dendrites. This difference is doubtless related to the variation in manganese from top to bottom and is of great interest because it offers a method for studying the effect of manganese on the temperature at which the segregation or precipitation of sulphide takes place in the melt during freezing. The black region at the top in Fig. 38 was caused by a flotation of sulphide rich in manganese.

Melt No. 138 (Additions: Mn., 0.90 per cent; S., 0.15 per cent)

56. This print resembles Fig. 38 except that the cellular distribution in the lower part is less distinct.

Melt No. 139 (Additions: Mn., 1.20 per cent; S., 0.15 per cent)

57. The sulphide distribution is very similar to melt no. 138. Traces of a cellular arrangement still are visible near the bottom. In the upper part, the primary dendrites were dominant.

58. The mode of distribution of the sulphide in these specimens suggests a relationship between the manganese content and the temperature at which sulphide inclusions are precipitated during cooling. In a melt containing no manganese, a major precipitation of FeS occurs during solidification. The formation of primary dendrites concentrates sulphur into the remaining liquid which is approaching eutectic composition and, as the eutectic freezes, sulphur is concentrated into the last part to solidify, *i. e.* into the cell boundaries, where inclusions of FeS finally are precipitated.

59. The presence of manganese in the melt changes the whole picture. Wohrman[2] has thrown much light on the relationship between manganese, sulphur and iron in systems containing mostly iron. Although it is not certain that his conclusions apply precisely in the presence of much silicon and carbon, they are certainly worth careful consideration.

Sulphur-Manganese Relations

60. A partition of sulphur occurs between the manganese and the iron resulting in the formation of inclusions of a solid

solution FeS + MnS, the composition of which is a function of
the relative concentration of manganese. Pure MnS can form
only in the presence of a large excess of manganese. Wohrman
says: "The solubility in iron of the remaining sulphide is a func-
tion of the FeS content of the sulphide. The higher the FeS
content, the more sulphide is held in solution at a given
temperature."

61. In the case of cast iron, the temperature of precipita-
tion of FeS + MnS may be far above the freezing point of the
eutectic or even above the temperature at which primary austen-
ite begins to form in hypo-eutectic alloys. If time is allowed,
some of the inclusions float to the top, resulting in the well known
phenomenon of desulphurization in ladles. As the temperature
falls, precipitation of FeS + MnS may continue during the
growth of primary dendrites. Inclusions formed at this time are
trapped and regimented by the dendritic growth, resulting in
formations like that in the upper part of Fig. 38. This is a com-
mon mode of distribution of the sulphide inclusions in gray iron[4].

62. In the presence of much manganese, most of the sul-
phide may be precipitated before the freezing of the eutectic
begins so that the cell formation does not influence the distribu-
tion observed by sulphur printing. In view of the effects of sul-
phur on graphitization, such a mechanism may play an important
part in determining the structure because it limits the effective
sulphur content of the eutectic liquid and influences the composi-
tion gradients set up during the growth of each rosette.

63. It will be recalled that in Fig. 27 a sulphur content of
0.139 per cent was required to convert the abnormal alloy no. 8
into a material resembling cast iron. This was in the absence of
both manganese and hydrogen. Where hydrogen is present and
manganese is absent a sulphur content of 0.018 per cent is able
to do the same thing (Fig. 15).

64. If manganese is added to the metal shown in Fig. 27,
remelting in vacuo produces structures like Figs. 23 or 24. A
percentage of manganese equal to the percentage of sulphur is
sufficient to produce a structure in alloy no. 8 like Fig. 23 when
such material is melted at a pressure of 2 mm. of mercury. This
same response has been shown by numerous commercial gray
irons when these were melted under the same conditions, such

material invariably containing sufficient manganese, even though
some is lost by distillation under reduced pressure. If such
vacuum treated material is remelted in hydrogen, a normal struc-
ture is restored and the change can be repeated as often as de-
sired by melting successively in vacuo and in hydrogen. There-
fore, we may consider that manganese, in the absence of hydro-
gen, neutralizes the flake changing propensities of sulphur.

65. The close connection between hydrogen and sulphur
already has been pointed out. In the theory of graphitization
outlined above, the effective sulphide, as far as flake size is con-
cerned, is that remaining in solution in the liquid along the solid-
liquid interface during the growth of the rosette. This, in a man-
ner not understood, determines the effectiveness of the hydrogen.
Three general postulates may be stated:

(1) If hydrogen is present, a very low sulphur con-
centration in the eutectic liquid is sufficient to produce nor-
mal flakes along the solid-liquid interface (Fig. 15).

(2) If hydrogen is absent, a much higher sulphur con-
centration is needed to produce the same effect (Fig. 27).

(3) Manganese limits the amount of sulphide remain-
ing in solution in the liquid by changing the temperature at
which FeS + MnS is precipitated during solidification. Un-
der condition (1), manganese may leave sufficient sulphide
in solution to produce normal flakes (Fig. 40). Under con-
dition (2), manganese may prevent normal flake develop-
ment entirely (Fig. 39). There are, of course, all degrees of
variation between these extremes.

The Occurrence of Ferrite

66. Some comment should be made on the ferritic layers
in the melts made under hydrogen, although changes at the eutec-
toid temperature are not involved directly in the mechanism of
graphitization outlined.

67. It is reasonably certain that the manganese and sulphur
content of these melts changes gradually from top to bottom.
In melt no. 136, a ferritic layer appears at the very top (Fig. 34).
As the total manganese addition is increased, the ferritic region
gradually descends in the melt as a horizontal band, above which
a fully pearlitic region occurs. With a high manganese addition,
the ferritic band is quite narrow and sharply defined (Fig. 37).

FIG. 39—ALLOY NO. 8 PLUS 0.09 PER CENT MANGANESE AND 0.032 PER CENT SULPHUR.
MELTED IN VACUO. ETCHED IN HNO₃, 20X.
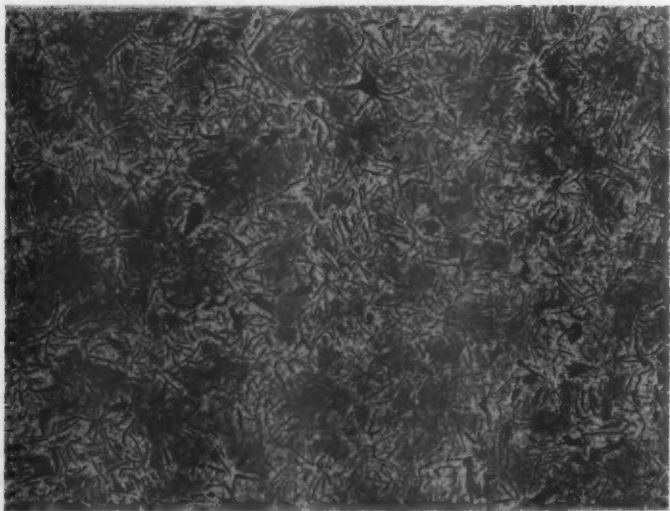
FIG. 40—SAME METAL SHOWN IN FIG. 39 REMELTED IN HYDROGEN. ETCHED IN HNO₃, 20X.

This is taken to mean that a certain ratio of sulphur and manganese produces a ferritic structure when this alloy is melted in hydrogen. The steeper the concentration gradient from top to bottom, the narrower the zone of suitable composition becomes. For example, Fig. 37 has a steeper gradient than Fig. 36.

68. It should be borne in mind that the exact ratio involved in this case applies only to the carbon and silicon content of this particular alloy and to the rate of cooling and maximum temperature used in the experiment.

69. In the study of alloys melted in the globar furnace, no simple relationship has been found between the changes in flake size and the amount of ferrite appearing in the final structure. With the pure iron-carbon-silicon alloys, melting in hydrogen increases the amount of pearlite. Alloy no. 7 becomes fully pearlitic while alloy No. 8 is about half ferrite (Fig. 22). Remelting in vacuo restores the original structures. There is some resemblance between the distribution of the eutectic carbide in Fig. 21 and that of the eutectoid carbide in Fig. 22, which may be related to the segregation of hydrogen previously mentioned.

70. In alloys containing sulphur but no manganese, melting in hydrogen invariably gives a fully pearlitic structure. Remelting in vacuo produces ferrite, the amount formed decreasing as the sulphur is raised (Figs. 23, 24, 25). When manganese is present as well as sulphur, no such consistency is found. An alloy, which is partly ferritic when melted in air or nitrogen, generally will show an increase in pearlite when melted in hydrogen and an increase in ferrite when melted in vacuo.

71. In some cases, however, melting in hydrogen produces a fully pearlitic core with an outer rim containing ferrite patches at the center of the rosettes. The ranges of composition in which this occurs cannot be stated from the data in hand. Aside from variations in manganese and sulphur, it is obvious that both carbon and silicon are very important factors in the response to hydrogen.

SUMMARY

72. In hypoeutectic gray iron, the graphite flakes form during the freezing of the eutectic and grow radially from the crystallization centers of the eutectic liquid. With a given rate of cooling, the size of the flakes is determined by the rate of

graphitization along the solid-liquid interface. Sulphur and hydrogen in solution in the eutectic liquid decrease the rate of graphitization, and the segregation of these elements produces composition gradients around each crystallization center.

73. The amount of sulphide remaining in solution in the eutectic liquid is determined by the extent to which sulphide inclusions have been precipitated in the melt above the eutectic temperature.

74. Manganese, by changing the solubility of the sulphide, limits the amount remaining in solution along the solid-liquid interface during the growth of the graphite flakes. An alloy of constant silicon and carbon content can be made to assume a great variety of structures by varying the manganese and sulphur contents and melting under circumstances conducive to a change in the amount of hydrogen in solution in the molten metal.

References

1.—Boyles, A., *"The Freezing of Cast Iron."* TRANS. A. I. M. E., (1937), vol. 125, pp. 141-203.

2. Wohrman, C. R., TRANS. A. S. S. T., (1928), vol. 14, pp. 255-299.

3. Sieverts, Z. phys. Chem., (1911), vol. 77, p. 591.

4. Hanemann and Schrader, *"Atlas Metallographicus, Band II,"* Lieferung 2, Tafel 11-12-13.

DISCUSSION

Presiding: J. W. BOLTON, Lunkenheimer Co., Cincinnati, Ohio.

P. C. ROSENTHAL[1] *(written discussion):* Although little can be added to enhance the excellent paper of the author, there is one minor point concerning the ferritic layer observed in the manganese series of melts that seem worthy of some comment.

Mr. Boyles found that, by adding increasing amounts of manganese to a melt containing 0.15 per cent sulphur, he obtained a variation in microstructure depending upon the manganese content. With 0.40 per cent manganese (Fig. 34), a layer composed of large graphite flakes, pearlite and ferrite made its appearance at the top of the melt. Increasing the manganese (Figs. 35, 36 and 37) shifted this layer down towards the center of the melt. It was suggested that this ferritic structure is linked up with a certain ratio of sulphur and manganese which is established by a concentration gradient of these elements from the top to the bottom of the melt.

[1] Dept. of mining and metallurgy, The University of Wisconsin, Madison, Wis.

Continuing from this point, it appears that a plausible explanation for the change in microstructure from the top to the bottom of the melt, which, in the case of Melt No. 138 and 139, Figs. 36 and 37, places a ferritic layer between pearlitic layers can be based not only upon the effects of the reaction between manganese and sulphur, but also upon the recognized ability of manganese to stabilize the carbides when present in amounts in excess of that necessary to form manganese sulphite. For instance, in Fig. 37, the upper layer of medium flakes and pearlite could be attributed to the carbide stabilizing effect of the excessive manganese in this region; the central ferritic layer could be indicative of a composition ratio of manganese and sulphur whereby they tend to counteract the carbide stabilizing effects of each other through the formation of manganese sulphite; and the lower layer of small flakes and pearlite could be due to the carbide stabilizing effect of sulphur in the absence of sufficient manganese to counteract this reaction as in the case of the low manganese melt No. 135, Fig. 33. That there seems to be some justification for the assumption that a composition gradient exists is substantiated by the author's data which include a sulphur print (Fig. 38) showing a concentration of sulphur in the upper part of the melt and also the results of a chemical analysis (page 31) showing a difference in the manganese and sulphur contents between the top and bottom of Melt 139.

ALFRED BOYLES *(written reply to Mr. Rosenthal's discussion)*:. The writer is indebted to Mr. Rosenthal for his interesting discussion of the ferritic layer. Some caution is necessary, however, in attempting an exact or quantitative generalization concerning the relation between
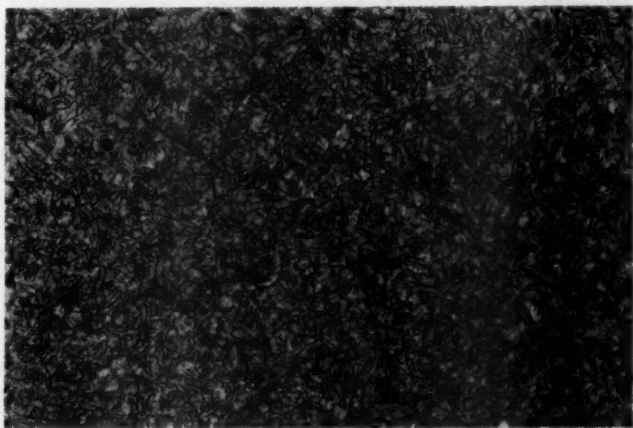
FIG. 41—ALLOY NO. 8 PLUS 0.08 PER CENT MANGANESE AND 0.061 PER CENT SULPHUR. MELTED IN NITROGEN. ETCHED IN HNO₂. — X20.

manganese and sulphur in these alloys. An addition of the two elements in the ratio corresponding to the compound MnS will not produce a ferritic structure.

As an example, Fig. 41 shows alloy No. 8 with 0.08 per cent manganese and 0.061 per cent sulphur melted in nitrogen at 2500°F. and slow cooled in the usual manner. The matrix is fully pearlitic and has large graphite flakes. When the same metal was remelted in vacuo, it closely resembled Fig. 24. Upon remelting in hydrogen, the alloy became fully pearlitic and contained fine graphite flakes (Fig. 42). The final analysis shows 0.08 per cent manganese and 0.050 per cent sulphur, which is not far from the ratio MnS. In Fig. 40, which shows considerable ferrite, the amount of manganese is considerably above the ratio MnS. The ferritic structure is probably related, not only to the ratio between manganese and sulphur, but also to the total amount of both elements present.

In any case it is unlikely that inclusions of pure MnS form in cast iron, even in the presence of a large excess of manganese. What we term "manganese sulphite" inclusions should not be looked upon as so much inert matter mechanically mixed with the metal. On the contrary, such inclusions represent the by-product of a complex chemical relationship between iron, manganese, and sulphur, acting during the freezing period and intimately bound up with the whole mechanism of selective solidification.
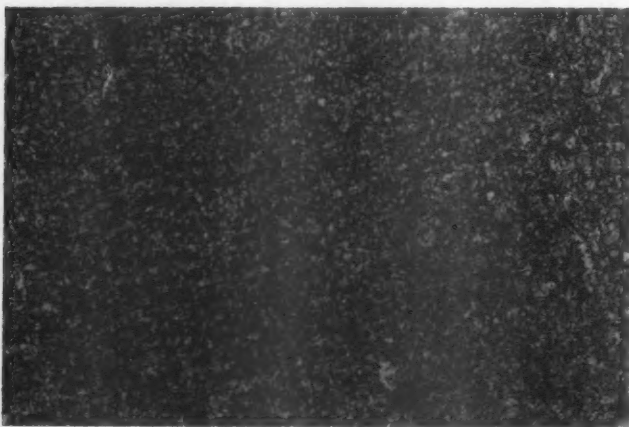
FIG. 42—SAME ALLOY AS IN FIG. 41, REMELTED IN HYDROGEN. FINAL ANALYSIS 0.08 PER CENT MANGANESE AND 0.05 PER CENT SULPHUR. ETCHED IN HNO₃. — X20.

J. S. VANICK[2]: I am much interested in Mr. Boyles' extension of the work he did a year ago, illustrating that in getting closer to the foundry type of product which carries manganese, sulphur, and perhaps gases in considerable quantity, the quality of the crystallization is becoming somewhat better understood. It is also appreciated, that it is possible to get both the eutectic graphite or the pseudo-eutectic graphite structure as well as the flake graphite structure in an ordinary gray iron.

G. P. PHILLIPS[3]: This paper is certainly of value in extending our understanding of the fundamentals of gray iron metallurgy and Mr. Boyles is to be complimented on the work. We, of industry, should certainly support this type of paper, to the end that we all have a better understanding and better control over our melting procedure and practice.

HARRY B. SWAN[4]: I have always felt that we had a very substantial lack of knowledge on the effect of gases in our metals, and in practice, we have to contend with a good many of these things and often we only guess at them. A paper of this character I feel is getting down more to the fundamentals of the causes and the theories and I hope that we will have many more like it. As a matter of fact, I happened two or three years ago to discuss this matter with Dr. Gillett who agreed with me that more knowledge on the gas content of iron would be most helpful. The Association should be congratulated on getting papers of this character.

H. A. SCHWARTZ[5]: We talk about the gas content of metals and we just assume something about it. It seems to me we are getting a little closer to doing more than assume in the present case, by at least having a technique which suggests that when the speaker thinks there is hydrogen present, he has really done something that makes it very reasonable that hydrogen should be present.

The thought comes to me whether anything has been done to actually determine the hydrogen content. I know that the determination of hydrogen in these metals is somewhat of a doubtful matter, if for no other reason than that the hydrogen may be is not there any longer when you start making the analysis. It has become somewhat simple to burn iron in the same general manner as in the determination of carbon, filter out sulphur on lead chromate, and get hydrogen values that are somewhat in line with what we get by vacuum fusion. Then we do not need quite all the equipment that the vacuum fusion determination requires. We have been interested in this and with the help of the Bureau of Standards for a year or so I have thought perhaps some of us could learn something as to the actual amount of hydrogen that causes these effects, with some information being obtained by this or similar methods.

[2] Metallurgist, International Nickel Co., New York, N. Y.
[3] Metallurgist, Tractor Works, International Harvester Co., Chicago, Ill.
[4] Foundry Supt., Cadillac Motor Car Division, General Motor Sales Corp., Detroit.
[5] Manager of Research, National Malleable & Steel Castings Co., Cleveland, Ohio.

MR. BOYLES: In that connection, if you will refer to Fig. 29 in the paper which is a curve taken from Sieverts, who has done most of the work on hydrogen solubility in iron, it will be noticed that the amount in 100 grams is, say, 20 cc. of gas. In weight per cent, that would be very low. I do not know what it would run but, it would be very difficult to determine, I should think.

We have thought about the combustion method as brought up by Dr. Schwartz, but the experimental difficulties are so great that we have not gone into it.

Another consideration is this. The curve indicates that as the metal cools, the hydrogen diffuses out, so you might not find in this metal melted in hydrogen, after it has cooled to room temperature, an amount at all proportional to the quantity absorbed in the liquid state, I believe the amount in solution at the time of freezing is the significant thing.

A. H. DIERKER[6]: When Mr. Boyles showed Fig. 2 on the screen, he made the statement that the white structure was a liquid when quenched and the rosette structure which was already formed was the eutectic. I cannot quite follow that the eutectic was formed and you still had a liquid.

MR. BOYLES: The freezing is gradual. The dark part represents what was solid at the time of quenching. The freezing of the eutectic is not instantaneous but would require some time. The liquid remaining was still of eutectic composition but froze white during the quench.

F. J. DOST[7]: A question for the benefit of the practical or operating men. I know there are several who have had the same problem that we have had. In all the papers given yesterday and today, papers on late additions or ladle treatments for the purpose of breaking up dendritic structures and getting a more normal graphite distribution, have been given in terms related to tensile strength. Nothing has been said about the type of structure which would give the best wear. I happen to be in the machine tool business and I know there are many like myself, who have been striving for the last year or year and one-half to find some means of ladle treatment to give the best wearing type of structure. I would like to ask if any of the research men have made any analysis or have drawn any conclusions as to what type of structure, which, as a result of these treatments, would give the best wear resistance for machine tool castings.

CHAIRMAN BOLTON: You refer, I presume, to the formation of pseudo-eutectic graphite in small rosettes in a ferrite background. In other words, you had an effect similar to that produced by undercooling. Instead of getting flake graphite, you got very fine graphite that was so intimately associated with the carbon in the austenite that there was such little distance between flakes left there, the carbon migration was rapid and you got little soft spots of ferrite when cooling below the transformation range.

MR. DOST: That is right. We actually had trouble with the scoring of the machine beds and our determination was that it was due to the dislodgment of these soft, mushy areas. That is what we concluded they were. We have used ladle treatments of different types and we are striving to get a pearlitic structure. Now, are we right in trying to get a pearlitic structure as the best type of structure for machine tool castings? In other words, that is the point we are striving for, to get a pearlitic structure. We feel that with a normal graphite and a truly pearlitic matrix, so far as our knowledge is concerned, we have the best type of material.

H. BORNSTEIN[8]: These questions on wear, I think, can only be solved by results from actual experience. In our particular case, which was on wear on automotive cylinders, we found that the pearlitic structure was desirable, but that does not mean that that would work out on machine tools. I think you have to correlate the results from actual service with the structures and then say that you want your matrix of a certain structure and you want your graphite of a certain pattern. If you have already found that the so-called dendritic structure is objectionable, of course, you try to do away with that. Probably, the next step would be to produce a so-called pearlitic iron and then see what you get.

There is another thing that will come into the problem very soon, I think, in your particular case. Do you want a pearlitic iron with 3.50 per cent total carbon or do you want a pearlitic iron with 3.25 per cent total carbon, or should it be 3 per cent? The tendency seems for some of these applications to be away from the high-strength, low-carbon irons, but the actual proof is in actual experience; and I do not believe anybody here, unless he has actually tied in analysis, microstructure and all that sort of thing with results in the field, under similar conditions, can give you an accurate answer to your problem.

MR. DOST: Mr. Lane gave a very nice paper last year at the Milwaukee convention and it was my hope that he would continue to do more work or possibly someone else would follow up his good work in that connection. I know with us in the machine tool industry and the machinery industry, where wear is the vital problem, it is a study very much worth while considering.

CHAIRMAN BOLTON: What actually has happened apparently, in the machine tool industry, is that in the "good old days" years ago, when little attention was paid to structure, you got large graphite flakes surrounded by ferrite. The machine tool people have tried and tried to get smaller graphite flake size and they have been rather successful in doing that. Then, in this urge for finer and finer graphite, they got down to a point where they got this undercooled structure, and although they have extremely fine graphite, it is found in a ferritic network, and this soft network has caused them trouble.

[8] Director of Laboratories, Deere & Co., Moline, Ill.

F. G. SEFING[9]: I have investigated quite a few irons that have shown both excellent and bad service in various kinds of heavy and light work, and in all cases where there was a dendritic pattern, they have shown bad wear service. I have also found that all those that showed bad service had some form of ferrite associated with it.

Let me exemplify that just a little further. This matter of ferrite is rather relative, but I have found a few cases where the pearlite bands were so broad that the size of the ferrite areas in the pearlite approached the size of the ferrite that you get in a dendritic pattern, and when that occurred, there was some trouble with regard to galling in some of the machines. So that there is not only an optimum of graphite size, which I believe Mr. Bornstein mentioned, but also an optimum of pearlite size. What this is, I certainly am unable to say, but we are going to go ahead investigating structures related to galling and wear in our laboratory, and we hope to have something more to say about it.

CHAIRMAN BOLTON: I would like to make a critical comment upon this paper by Mr. Boyles. Fig. 14, in connection with the question asked by Mr. Dost, may lead us a little bit off the right track. In Fig. 14, Mr. Boyles, indicates by the line O-B that as you increase the carbide stability you increase the graphite flake size. That does apply within the limitations of his experimentation. It does not apply in general and, an exception to this curve O-B-A is where these machine tool men are getting into trouble. If one followed this curve, one would think that is the direction in which the machine tool people should go. It is not.

I would like to ask Mr. Boyles, for example, to explain the mechanism and the result of superheating as related to this curve. It seems to me that you have a stabilization of carbide by superheating and you have a decrease in the flake size. I am referring to Fig. 14.

MR. BOYLES: I have no answer to the chairman's question except to say this, that we should, in speaking of carbide stability, distinguish between the stability of the eutectic and the stability of the eutectoid. What he is speaking of is the amount of pearlite. My contention is that the stability, or, to put it otherwise, the rate of graphitization of the eutectic liquid determines the flake size. This is not necessarily related to the amount of pearlite that will appear in the iron. I have not found any simple relationship between the two. You may increase the chill of iron and also increase the amount of ferrite. You men are all familiar with that, I am sure. It should have been stated that Fig. 14 applies only to the eutectic.

* Research Metallurgist, International Nickel Co., Bayonne, N. J.

Deoxidation and Graphitization of Cast Iron

By R. G. McElwee[*], Detroit, Mich.

Abstract

This paper suggests means for the practical foundryman to experiment logically with those jobs which may be giving trouble due to abrupt changes in structure. It assumes that the foundryman is interested in making cast iron a more versatile material by reducing the effect of different cooling rates in the same casting or other castings from the same pattern. It suggests the use of a test casting (wedge) which assists logical study of effects of experiments.

1. For a number of years, and particularly in the past 2 or 3 years, a number of papers have been contributed to the literature on cast iron dealing with the theory of deoxidation and graphitization. Naturally, any consideration of the freezing of cast iron also covers this subject. For the most part, these papers have been attempting to explain, and have very satisfactorily discussed, the physical changes which take place in cast iron composition before, during and after its solidification. To assist the production foundryman in applying some of these principles, this paper is presented, with the fact in mind that a discussion of the various changes in composition which might be effectively used to eliminate certain difficulties should be presented in a practical manner.

2. In making hundreds of test bars as a practical study of this problem, it has become apparent that certain alloying additions, which affect the form of carbon in the final casting, have two somewhat different effects that may be at times confused, because in each case the final result may be the increase in graphitic carbon and a decrease of combined carbon. One of the fundamental things that we are taught in dealing with silicon in cast iron is very graphically expressed by the Maurer diagram and relates to the intersolubility of carbon and silicon. In this respect, the graphitizing alloys may be assumed to change the carbon form because of their ability to remove carbon from solution by reducing the solubility of the melt for that material. Practically all commonly used alloy additions, including silicon, may have, in addition to this effect, an indirect influence on the combined carbon present by reactions with certain elements which enter the mixture with the blast or by other uncalculated means.
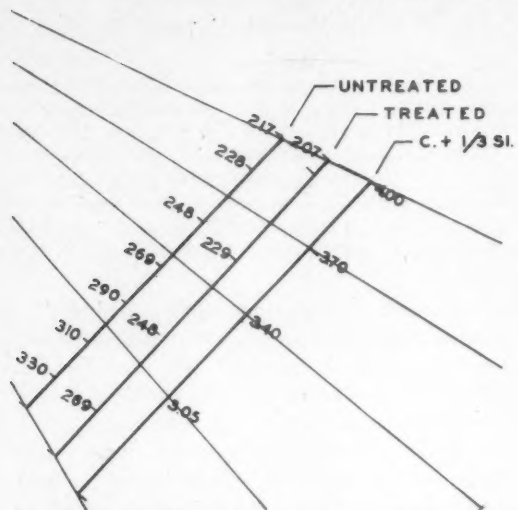
FIG. 1—SOFTENING EFFECT OF A COMMERCIAL DEOXIDIZER CONTAINING TITANIUM, ALUMINUM
AND SILICON IN RATIO OF 1:2:4, RESPECTIVELY.
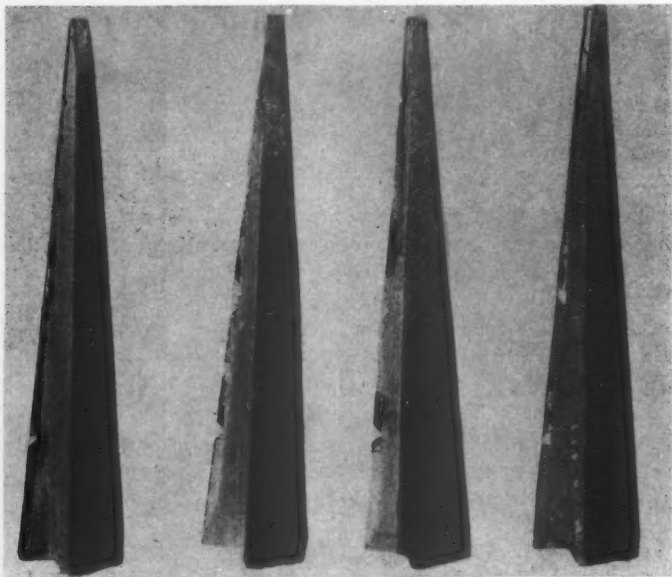
FIG. 2—TEST WEDGES SHOWING WHITE IRON PRESENT AND THE CHARACTER OF TRANSITION
FROM WHITE TO GRAY SECTIONS.

3. For the convenience of this discussion, we are assuming that the oxygen present in any mixture may have very different influence on the structure of the iron after its solidification, and it tends to increase the chill, that is, to retain the carbon in a combined form, and that it also tends to promote eutectiform graphite and primary ferrite in those sections of a casting which are so proportioned to the mixture under discussion as to be susceptible to undercooling. It is obvious that the amount of oxygen which can remain in the charge to produce the effects referred to, will be inversely proportional to the carbon present and, to illustrate how this effect may be practically measured, we have prepared a graph, Fig. 1, which indicates the experience over a large number of irons of varying composition, some of which were untreated and others of which were treated with a constant amount of a commercial deoxidizer containing titanium, aluminum and silicon. For the convenience of those who prefer to calculate on a basis of carbon plus one-third silicon, this scale is also shown against the Brinell hardness experience.

4. Fig. 2 shows pictures of test wedges, in which it is desired to note not alone the amount of white iron present, but also the character of the transition from white to gray. Wedges of this type are highly recommended for the study of graphitizing conditions with relation to the particular casting job. They are much more effectual than step bars because the smooth gradient of cooling rate can give readily translated information from one test to another. The test wedge, if a proper size is chosen, represents a complete range of cooling conditions and structures found in the casting, and more conveniently located for examination.

5. In making a study of this kind, it is well also to take into consideration the change in Brinell hardness from heat to heat and this hardness test should be taken at several places in the bar and always to a point equal in hardness to the softest part of the casting. What this wedge indicates, which is not clearly shown by a standard chill test or the Brinell test on the arbitration bar, is the ratio of graphitizing rate to cooling time. Graphitization in the heaviest section may cause a softening of that section which is not proportional to the effect of alloy treatment in the lightest section. In other words, it is possible, by certain combinations of material, to produce greater lengths of white iron in the same bar in which the Brinell test may be noticeably lower at the heavy end.

6. The narrow transition from white to gray, especially if the transition occurs at some distance from the end, indicates that the iron is very susceptible to slight changes in cooling rate and that castings produced therefrom may have abrupt changes in hardness from one part of the casting to another. It is highly desirable, therefore, to so proportion the charge, or the alloy additions, if any, so that the wedge will show the greatest approach to strict proportionality of cooling rate to Brinell hardness. This naturally indicates a wider transition zone.

7. There is also the effect to take into consideration which is often referred to as dendritic pattern or eutectiform graphite, usually but not always associated with primary ferrite. Our own experience has shown that for many hundred tests the susceptibility to this undesirable structure is definitely related to the degree of sensitivity of the iron, as indicated by the wedge. The narrower is the zone, the higher the frequency of undercooled results. There may be some dispute as to the cause of this condition but there are certain facts which have been developed which indicate that it can be produced practically at will by certain methods of treatment including excessive superheat, holding at high temperatures for long intervals of time or by the addition of metallic oxides. Fig. 3, shows (a) the structure of a plain iron (b) the same iron to which a definite quantity of titanium oxide was added, calculating 0.20 per cent titanium, and (c) the same iron to which subsequently another addition was made of Ti-Al-Si, containing approximately 0.01 per cent Ti, 0.02 per cent Al and 0.04 per cent Si in the same amount (0.01; 0.02; 0.04) as mentioned in paragraph 3. This latter addition not only restored the random pattern of graphite but also improved the physical properties over the bars poured from the ladle which was not so treated. For instance the addition of deoxidizing alloy Ti-Si-Al changed the properties as follows:

	Transverse, Lb.	Deflection, In.	Tensile, Lb. Per Sq. In.	B.H.N.
Oxidized	2265	0.176	35000	225
Deoxidized	2365	0.320	39650	185

8. Working on the assumption that this eutectification graphite occurs as a result of undercooling and mold annealing, we would expect to find this structure in the wedge between the white and gray in those irons in which the transition is narrow. From a control standpoint, if this condition also exists in the casting which
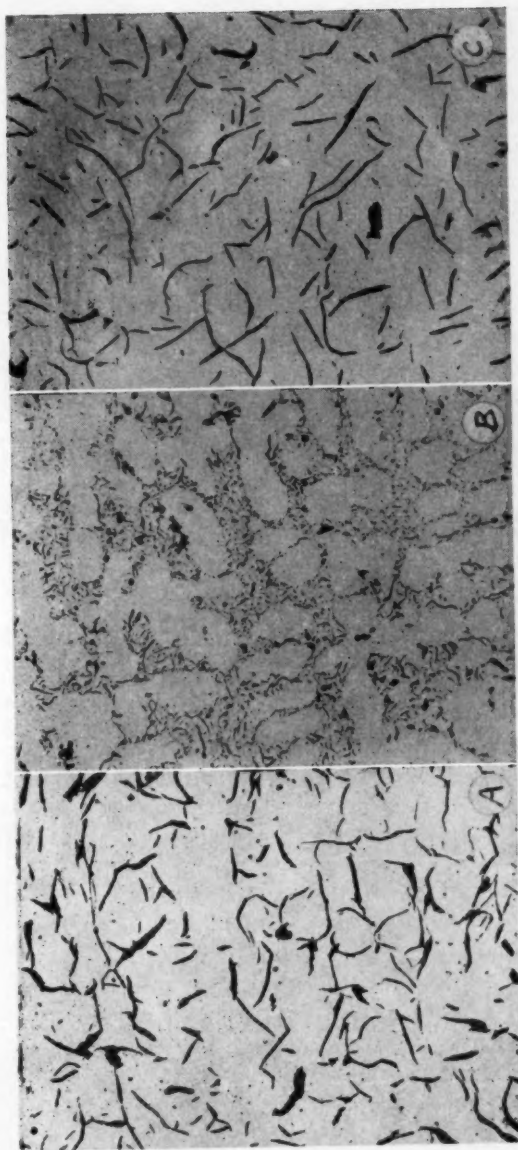
Fig. 3.—A—Plain Iron—X100. B—TiO₂ added. C—Ti-Al-Si added.
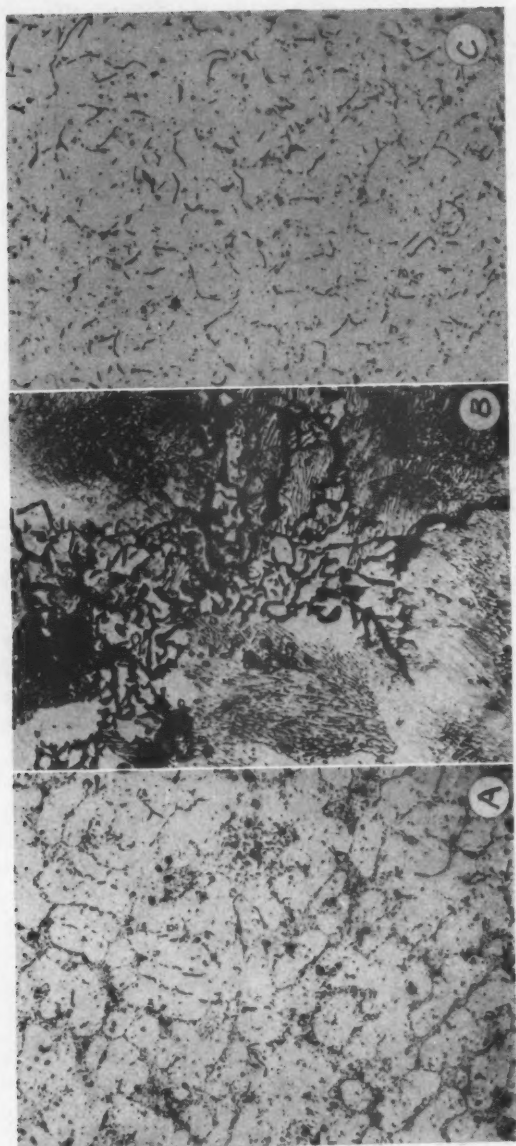
Fig. 4.—A—Plain Iron No. 7, Unetched—X100. B—Plain Iron No. 7, Etched—X1000. C—Iron No. 6, Treated with Separate Alloys, Unetched—X100.
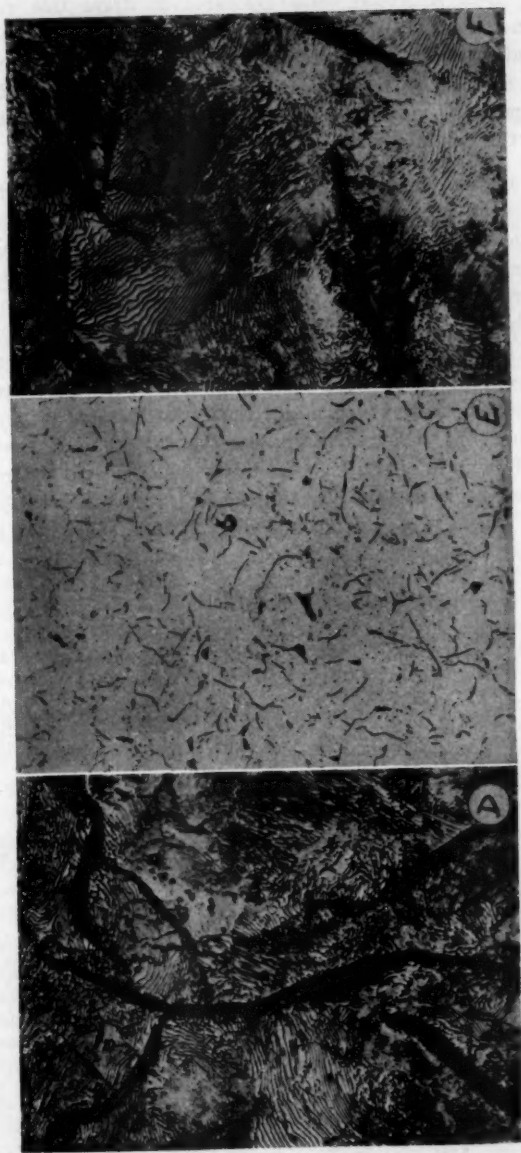
FIG. 4—D—IRON NO. 6, TREATED WITH SEPARATE ALLOYS, ETCHED—X1000. E—IRON NO. 5, TREATED WITH COMBINATION ALLOYS—UNETCHED—X100. F—IRON—NO. 5, TREATED WITH COMBINATION ALLOYS—ETCHED—X1000.

is being studied, it would seem desirable to drive this cooling phenomenon towards one end or the other, which of course can be accomplished by definitely softening or hardening the mixture. In many cases, this can be done with unidirectional control to a degree which will permit the entire elimination of this particular cooling effect from the casting.

9. In such castings, however, as seem impractical to change by hardening or softening, due to specific hardness specifications or requirements, it is entirely practical to spread the zone, thereby reducing the tendency to undesirable structure, and to produce, in this manner, castings which are of controlled hardness in any specified point, and in which the hardness limits can be definitely narrowed. This involves the simultaneous addition of alloys working in different directions, for instance, silicon and chromium, or any other combinations which might theoretically seem to neutralize each other but which actually can be so balanced that the hardener will be most effective in the heavy section, with the graphitizers or deoxidizers most effective in the light section. Examples of this control are shown in Fig. 4. Tables 1 and 2 also are given

Table 1

COMPOSITION OF IRONS

Iron No.	T. C., %	Si., %	Mn., %	S., %	P., %	Ti., %	Cr., %
5..........	2.51	2.57	0.90	0.032	0.050		0.27
6..........	2.53	2.70	0.89	0.037	0.058		0.28
7..........	2.55	2.39	0.89	0.039	0.058	0.01	0.06

Table 2

DATA FOR STRESS—STRAIN DIAGRAM

Iron No. 5				Iron No. 6				Iron No. 7			
A		B		A		B		A		B	
Transv. Strength lb.	Defl., in in.	Transv. Strength lb.	Defl., in in.	Transv. Strength lb.	Defl., in in.	Transv. Strength lb.	Defl., in in.	Transv. Strength lb.	Defl., in in.	Transv. Strength lb.	Defl., in in.
0	0	0	0	0	0	0	0	0	0	0	0
200	0.014	200	0.018	200	0.016	300	0.020	200	0.014	200	0.012
400	0.028	400	0.028	400	0.028	400	0.029	400	0.027	400	0.026
600	0.042	600	0.044	600	0.044	600	0.046	600	0.042	600	0.036
800	0.058	800	0.059	800	0.058	800	0.060	800	0.055	800	0.050
1000	0.073	1000	0.073	1000	0.074	1000	0.074	1000	0.066	1000	0.065
1200	0.088	1200	0.088	1200	0.088	1200	0.090	1200	0.079	1200	0.078
1400	0.094	1400	0.104	1400	0.105	1400	0.106	1400	0.095	1400	0.091
1600	0.100	1600	0.121	1600	0.123	1600	0.123	1600	0.108	1600	0.104
1800	0.138	1800	0.137	1800	0.140	1800	0.140	1800	0.122	1800	0.120
2000	0.155	2000	0.157	2000	0.157	2000	0.157	2000	0.138	2000	0.134
2200	0.176	2200	0.176	2200	0.178	2200	0.178	2200	0.156	2140	0.146
2500	0.209	2400	0.196	2400	0.200	2400	0.200				
2600	0.222	2600	0.220	2500	0.211	2600	0.226				
2700	0.235	2800	0.244	2800		2700	0.241				
2800	0.250	3000	0.264								
2900	0.267	3110	0.308								

showing the composition of the plain iron and the alloy additions
and the working data for a stress strain diagram which very clearly
indicates the desirability of this type of treatment. Attention is
especially directed to the breaking load of the transverse bars and
to the relative rigidity of the irons at 2200 lb. transverse loading.

CONCLUSION

10. In conclusion, this procedure is suggested as a method of
approach and not as a cure. There are many combinations of com-
mercial alloys which might be used in a program of this kind, the
ones cited are examples only. If the hardener and graphitizer are
correctly proportioned to take care of the heaviest section of the
casting, the light sections, which often give trouble due to chilling
or undercooling, can then be treated by late additions giving a
deoxidizing effect which tends to reduce this chill without seriously
reducing the hardness in the heavy section.

DISCUSSION

Presiding: J. W. BOLTON, The Lunkenheimer Co., Cincinnati, O.

G. F. COMSTOCK[1]: Does the author feel that the addition of aluminum
in his graphitizing alloy had any distinctive effect as compared with
the titanium? He used both and in one case showed that after the addi-
tion of this deoxidizer, the graphite was changed from the very fine
eutectic form to a coarser form. Might not that effect have been caused
by the aluminum in the deoxidizer? Has the author any evidence one
way or the other, either for or against that idea?

MR. McELWEE: Unfortunately, I do not happen to have the answer
to that question. I think we have tests that will give some interesting
stories about that later on. If I interpret the question correctly, what
Mr. Comstock would like to know is whether we could not accomplish
the same result of deoxidation with a small amount of titanium.

MR. COMSTOCK: What I had in mind was the difference in effect
between aluminum and titanium, whether the coarsening of the graphite
was not due to the aluminum rather than to the titanium in the
deoxidizer.

MR. McELWEE: We have tried to use small amounts of titanium as
a comparison against this particular alloy of titanium, aluminum and
silicon, and we have found that the amount of titanium used is rather
critical. If we add too much titanium, using titanium alone we have not
obtained quite the desirable deoxidizing effect we think we get with the
combination of aluminum-titanium.

R. SCHNEIDEWIND[2]: We have tried deoxidation using straight
titanium, with ferro-silicon, and Dr. Murphy has done work with alumi-

num. Any one of these materials will straighten out and coarsen the graphite if properly used. Titanium is a little sensitive to temperature. Added to a material at a high temperature, it loses its effect rapidly and we do not get quite as much correction of the graphite. Aluminum, of course, is critical, in that if you use too much aluminum, you will end up with pin holes. The effect of ferro-silicon was described in the paper[3] presented yesterday, and is pretty well known to foundrymen. Ferro-silicon will correct this condition somewhat. Each one of these deoxidizing alloys seems to have a temperature range or a composition range in which it works best.

A. BOYLES[4]: Mr. McElwee, were these irons cupola melted or were they made in an induction furnace? We, at Battelle, have found a very marked difference in response between irons melted in different furnaces. We have no data on cupola melts but in the electric furnace, there is quite a marked difference in their response to various treatments. Do you also find that?

MR. McELWEE: Yes, that is probably true. I think Mr. Crosby brought out in the paper presented yesterday that that effect is related closely in composition. The ranges of our cupola irons are normally softer irons. The electric furnace is a tool for making low carbon irons and most of our electric furnace irons are low carbon irons. I think if we made low carbon irons in the cupola, we would have about the same problem to deal with. I believe there is not much question but that it is a matter of composition more than a method of melting. That is indicated by our first chart, where we have found that the effectiveness of deoxidation is related to the carbon-silicon balance rather than to the method of melting, although the method of melting may have some influence.

MEMBER: What are the dimensions of those wedge blocks and how do you break them without breaking the tip of the wedge block?

MR. McELWEE: These wedges are about 2 in. x 2 in. at the heavy end and are approximately 12 in. in length. We make a saw cut through about one-third of the long axis, as shown by the light section of the bars in Fig. 2, then fracture the bar. They are very easy to break. There may be some other way of doing this, but we found this method a convenient one.

L. H. RUDESILL[5]: Mr. McElwee found that he must control the hardness in the heavy section of the wedge and at the same time could control the chill in the light section. Also he mentioned there is a function of the addition and of the temperature. Would Mr. McElwee elaborate on that.

[3] E. R. Young, V. A. Crosby and A. J. Herzig, "Physical Properties of Cast Iron in Heavy Sections."
[4] Metallurgist, Battelle Memorial Institute, Columbus, O.
[5] Metallurgist, Griffin Wheel Co., Chicago, Ill.

MR. McELWEE: We found that some alloys which were considered to be very potent graphitizer-deoxidizer types of alloys, seemed to give a greater amount of chill in the wedge, that is, the wedge, did not graphitize where the cooling was rapid and graphitized very markedly where the cooling rate was slower in the heavy end. The zirconium-titanium combination happens to be one mixture that was tried that was very effective in that respect. Added to a normal cast iron, we would have more white iron and still have a drop of at least 20 points in Brinell in the heavy end of the wedge. Now, that is an indication that we can so proportion elements, when we know more about them, as to make a nearly straight line function. In other words, we can control this chill at the light end and the graphitization at the heavy end to a point more nearly to what we desired by combining the things that tend in the direction of slow action with those that tend in a direction of rapid action.

MEMBER: As I understood the speaker, he regards the addition of titanium oxide as a method of introducing oxygen, because he said "ferrous oxide" did the same thing. Since titanium is such a decided reducing agent of metallic oxides, I would like to have the speaker expound the idea that titanium oxide will do other than add an inert suspended oxide to the iron.

MR. McELWEE: I would not attempt to do that, except to say that we had the same result in the iron when we added ferrous oxide as we did with titanium oxide. As I said in the beginning, I am not attempting to theorize on this; I am simply throwing into this discussion of graphitization and deoxidation of cast iron some experiments for somebody else to find the answers.

CHAIRMAN BOLTON: Mr. Comstock, can you give us the heat of formation of titanium oxide versus ferrous oxide?

MR. COMSTOCK: That of titanium oxide is a great deal higher. The ratio is two or three to one. According to handbooks, heat of formation of FeO is 65,700 cal., Fe_2O_3 is 197,700 cal. and TiO_2 is 218,400 cal.

MR. SCHNEIDEWIND: I would like to throw a little light on this last experiment that Mr. McElwee was explaining. When the titanium oxide was added to the molten iron and the cast poured, an analysis was made for contained titanium and I think it ran round 0.04 to 0.05 per cent of dissolved titanium. There must be an equilibrium between the oxide and the molten metal.

MR. COMSTOCK: Dr. Schneidewind have you ever found any iron that did not show a small amount of titanium. We never have found that. The point I am trying to make is that all cast iron contains some titanium. One may not have added any from the oxide and still have found the titanium content, because the iron contained titanium in the first place.

MR. SCHNEIDEWIND: We have a certain method of analysis which may not be absolutely accurate, but at least we found about 0.04 or 0.05 per cent more in the iron so treated than in the companion heat to which no oxide was added.

MEMBER: There seems to be some question as to whether the increase in the physical properties is the result of certain alloys. I was wondering if all the improvement could be attributed to the late additions of ferro-silicon. In the absence of the chromium in connection with silicon, would starting with a lower silicon and a late addition of ferro-silicon not do the same things as the alloy in combination with the graphitizing agent?

MR. McELWEE: It would be a different experiment. Commercially, the balance of silicon and chromium in an iron is to permit the foundry-man to alloy the heat of the iron that he has. Now, the other thing is entirely feasible but it means changing a man's cupola practice, which we, from an alloy standpoint, cannot always do; so we approached the problem from this angle for that reason.

Comparative Effects of Late Additions of Titanium and Silicon to Gray Cast Iron

By G. F. Comstock[1] and E. R. Starkweather,[1]
Niagara Falls, N. Y.

Summary

This paper gives the results of tests on three grades of gray cast iron, each containing various amounts of chromium, and treated shortly before pouring with either foundry ferrotitanium or ferrosilicon. A comparison of the effects of these late additions showed that, with few exceptions, increasing ferrotitanium additions gave increased strength and hardness, while increasing ferrosilicon additions gave either a smaller increase in those properties or a decrease. The difference in the effects of these two alloy additions was more marked as the amount of the additions increased. The improvement resulting from the titanium additions was due largely to finer graphite and was more definite in higher-carbon or higher-sulphur irons. Sulphur was found to be effective in suppressing dendritic microstructure in the low-carbon chrometitanium irons. The irons treated with titanium showed fewer flaws in the fractures of the test-bars than either the untreated irons or those treated with ferrosilicon.

1. The alloy most commonly used for adding titanium to cupola-melted gray iron is an alloy containing about 20 per cent titanium, 20 per cent silicon, 0.3 per cent carbon, 0.4 per cent aluminum and not over 0.1 per cent impurities other than iron. It is designated as foundry ferrotitanium to distinguish it from other kinds of iron-titanium alloys. The silicon serves to reduce the melting point of the alloy and improve its solubility in molten iron, so that the iron does not have to be up to electric furnace temperatures to obtain satisfactory recovery of the titanium.

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2. It is obvious of course that when this alloy is used in the foundry, both silicon and titanium are added to the iron and the theory has been suggested that the effects of this addition are due chiefly to the silicon and could be obtained by the similar use of ferrosilicon alone. Considerable data on this point have been collected in the research laboratory of the company with which the authors are associated and are briefly reported in this paper.

3. The iron used for this work was all melted in Tercod crucibles in an electric resistance furnace, and the 50-lb. heats were made of pig iron and low-carbon steel scrap. In one series of heats, iron pyrites also was used in the charge to raise the sulphur content. Three basic mixtures were used, their compositions and analyses being given in Table 1.

4. The composition variables investigated in each of the mixtures described in Table 1 were not only silicon and titanium but also chromium. Small amounts of chromium have been found to confer interesting physical properties on gray cast iron, especially when titanium is used with it. Each of the above mixtures, therefore, was divided into four series, with additions of high-carbon 65 per cent ferrochromium respectively as follows:—

None, giving practically no chromium content.
3 oz. per 50-lb. heat, giving about 0.23 to 0.33 per cent chromium.
6 oz. per 50-lb. heat, giving about 0.42 to 0.53 per cent chromium.
9 oz. per 50-lb. heat, giving about 0.65 to 0.77 per cent chromium.

AMOUNT AND TIME OF ADDITIONS

5. The ferrochromium was stirred into the melt in the crucible about an hour after charging the last of the pig iron, and about half an hour before the melt was poured.

Titanium Additions

6. One heat of each of the series described above was poured without further additions; another was treated with 1.00 per cent foundry ferro-titanium (8 oz. per 50 lb. heat), and a third with twice as much. These additions were stirred into the melt 15 to 25 min. after the ferrochromium, or when the temperature, as determined by optical pyrometer with the usual correction for emmisivity, was 2670°F. Two minutes after the ferrotitanium addition, the crucible was removed from the furnace, and when the temperature had dropped to about 2550°F., the castings were poured.

Table 1
Compositions and Average Analyses of Base Irons

Designation of Mixture	Low Carbon, Low Sulphur	High Carbon, Low Sulphur	Low Carbon, High Sulphur
Low-Carbon Steel Scrap, Lb.	7	5	9
No. 1 Foundry Pig Iron (2.84 per cent Si.), Lb.	30	25	32
Bessemer Pig Iron (1.68 per cent Si.), Lb.	13	10	9
Low-Silicon Pig Iron (0.77 per cent Si.), Lb.	0.	10	0
Iron Pyrites (Fe S_2), ounces	0.	0	3.5
Average Carbon Content, per cent	3.2	3.5	3.1
Average Silicon Content, (without additions) per cent	2.0	1.8	2.2
Average Manganese Content, per cent	0.63	0.77	0.68
Average Phosphorous Content, per cent	0.25	0.23	0.28
Average Sulphur Content, per cent	0.04	0.04	0.16
Average Titanium Content, (without additions) per cent	0.05	0.055	0.06

Recoveries

7. The 1 per cent addition of foundry ferrotitanium raised the titanium content of the iron to about 0.10 to 0.14 per cent, an increase of 0.05 to 0.08 per cent, and also increased the silicon content somewhat more, or about 0.05 to 0.20 per cent. The 2 per cent addition raised the titanium content by an increment of about 0.13 to 0.21 per cent, and the silicon content about 0.20 to 0.35 per cent, giving roughly two or three times as much increase as the 1 per cent addition. The recovery of titanium was about 30 to 50 per cent of the amount added, while that of silicon from the foundry ferrotitanium addition appears to have been about 30 to 90 per cent. This recovery appears variable because of the rather irregular silicon contents of the heats, possibly due to absorption of silicon from the Tercod crucibles.

Silicon Additions

8. Two or three more heats were included in each of the series described, these additional heats having been treated with 50 per cent ferrosilicon in amounts adjusted to give approximately equivalent increases in silicon content to the increases derived from the foundry ferrotitanium additions. Such amounts were of the order of 0.22 to 0.67 per cent, the recovery of silicon from ferrosilicon being higher than from ferrotitanium, or practically 100 per cent.

9. The actual recovery of silicon from such small additions, either of ferrosilicon or ferrotitanium, cannot be computed accurately because the increase was so small compared to the silicon content of the untreated iron and its normal and unavoidable variation between individual heats made by identical practice. This was not true of the titanium additions, since the titanium content of the untreated iron was much lower and more constant. The amount of titanium added was 4 to 8 times the amount in the original iron, instead of 1/20 to 1/5 as much, which was true of silicon.

10. The extra additions of ferrosilicon referred to above were made to the heats when their temperature was 2670°F., and two minutes before the crucibles were removed from the furnace for pouring, thus copying the practice used with ferrotitanium.

11. All these heats were poured into green-sand molds to form three vertical test bars 15-in. long and about 1.25-in. in diameter.

All the bars were allowed to cool in the sand. They were tested transversely on supports 12-in. apart, the deflections being measured from autographic load-deformation diagrams. The transverse strength was computed from the breaking load and the dimensions of the bar at the fracture, by using the formula that the strength of a beam is proportional to the width times the square of the height. Results from bars weakened by surface flaws at the tension side were rejected, and if more than one bar from a heat was defective in this way the entire heat was rejected and made over.

Tensile Tests

12.   The lower (as cast) halves of two broken transverse test bars from each accepted heat were machined to standard tensile test-bar form, about 0.8-in. in diameter between shoulders, and with ends threaded to fit the spherical-seated grips of the testing machine. The average results from these two specimens was taken as the tensile strength of the heat. Surface flaws were of course removed by the machining.

13.   One bar from each heat was sampled for chemical analysis near the transverse fracture, and Brinell hardness tests were made on cross sections of the same bars cut at about the same location. The latter determinations were made under standard conditions both at the center of the section and about midway between the center and the edge. The results of all these tests are presented in Tables 2, 3, and 4.

14.   The transverse strength, hardness, and tensile test data in Table 2 are plotted on Fig. 1, and Figs. 2 and 3 similarly represent data from Tables 3 and 4, respectively. The full lines on each chart show variations of the properties with ferrotitanium additions in the titanium iron, and the dotted lines the variations with ferrosilicon additions in the non-titanium iron. The horizontal co-ordinates represent alloy additions in every instance, and not chemical compositions. In Fig. 1, the horizontal co-ordinates for the dotted lines are, as shown in the chart, not regular, but represent the ferrosilicon additions found to give as closely as possible the same silicon pick-up as was obtained from the ferrotitanium additions to the corresponding titanium irons. Thus on all the charts each vertical line represents somewhere near the same silicon content on both kinds of curves. A reasonably good comparison is thus afforded between results obtained with the ferrotitanium con-

Table 2

RESULTS OF TESTS ON LOW CARBON, LOW SULPHUR IRONS

	Heat Additions, percent			Composition, percent			Transverse Strength, Lb. per 1.25 In. Dia.	Deflection, In. with 12-In. Span	Brinell Hardness		Tensile Strength, Lb. per Sq. In.
No.	Cr.	FeTi	FeSi	Carbon	Silicon	Titanium			Center	Near Edge	
1	0	0	0	3.15	2.08	0.049	4245	0.113	187	192	38,150
2	0	1.0	0	3.27	2.19	0.128	4275	0.100	204	204	36,075
3	0	2.0	0	3.18	2.45	0.218	4380	0.095	207	207	37,500
4	0	0	0.22	3.08	2.28	--------	4400	0.108	207	210	38,780
5	0	0	0.63	3.16	2.42	--------	4368	0.123	197	197	37,350
6	0.25	0	0	3.20	2.00	--------	4210	0.115	192	192	35,325
7	0.25	1.0	0	3.28	2.33	0.120	4533	0.103	215	217	38,125
8	0.25	2.0	0	3.16	2.39	0.172	4273	0.088	229	229	40,125
9	0.25	0	0.22	3.18	2.24	--------	4460	0.110	217	217	39,100
10	0.25	0	0.67	3.16	2.41	--------	4400	0.117	204	207	38,470
11	0.50	0	0	3.28	2.10	--------	4460	0.098	217	226	38,450
12	0.50	1.0	0	3.30	2.31	0.122	4515	0.092	229	229	40,500
13	0.50	2.0	0	3.12	2.40	0.240	4855	0.093	248	245	40,750
14	0.50	0	0.22	3.08	2.24	--------	4717	0.100	232	235	39,420
15	0.50	0	0.45	3.08	2.33	--------	4273	0.095	223	229	39,630
16	0.75	0	0	3.34	2.02	--------	4360	0.110	197	212	37,525
17	0.75	1.0	0	3.26	2.24	0.120	4503	0.093	245	245	37,625
18	0.75	2.0	0	3.25	2.42	0.210	4728	0.088	262	262	41,650
19	0.75	0	0.22	3.19	2.30	--------	4552	0.098	238	241	43,050
20	0.75	0	0.65	3.19	2.41	--------	4568	0.125	213	219	39,930

Table 3

RESULTS OF TESTS ON HIGH CARBON, LOW SULPHUR IRONS

	Heat Additions, percent			Composition, percent			Transverse Strength, Lb. per 1.25 In. Dia.	Deflection, In. with 12-In. Span	Brinell Hardness		Tensile Strength, Lb. per Sq. In.
No.	Cr.	FeTi	FeSi	Carbon	Silicon	Titanium			Center	Near Edge	
21	0	0	0	3.44	1.82	0.060	4142	0.137	179	179	30,950
22	0	1.0	0	3.40	1.86	0.093	4313	0.125	187	192	33,625
23	0	2.0	0	3.44	2.08	0.232	4427	0.102	207	207	40,150
24	0	0	0.22	3.46	1.97	--------	3995	0.117	187	192	34,350
25	0	0	0.45	3.42	1.96	--------	4080	0.125	197	197	34,400
26	0.25	0	0	3.56	1.82	0.064	4147	0.110	201	207	35,850
27	0.25	1.0	0	3.50	1.98	0.112	4485	0.118	207	204	35,600
28	0.25	2.0	0	3.55	2.03	0.188	4555	0.105	217	217	41,725
29	0.25	0	0.22	3.42	2.00	--------	4015	0.115	197	197	35,200
30	0.25	0	0.45	3.37	1.99	--------	4117	0.120	197	197	36,000
31	0.50	0	0	3.59	1.77	0.054	4083	0.113	179	192	31,575
32	0.50	1.0	0	3.44	1.90	0.108	4682	0.115	210	212	37,625
33	0.50	2.0	0	3.44	2.00	0.240	4672	0.092	248	248	41,235
34	0.50	0	0.22	3.42	1.90	--------	4168	0.102	212	212	36,070
35	0.50	0	0.45	3.45	2.04	--------	4175	0.107	197	207	35,070
36	0.75	0	0	3.50	1.79	0.049	4245	0.103	201	212	36,325
37	0.75	1.0	0	3.42	1.89	0.129	4307	0.088	241	248	36,650
38	0.75	2.0	0	3.43	2.01	0.236	4574	0.093	269	255	40,950
39	0.75	0	0.22	3.42	1.95	--------	4285	0.095	229	229	37,590
40	0.75	0	0.45	3.42	2.02	--------	4150	0.097	210	217	36,570

Table 4

RESULTS OF TESTS ON LOW-CARBON, HIGH SULPHUR IRONS

Heat No.	Heat Additions, percent			Composition, percent			Transverse Strength, Lb. per 1.25 In. Dia.	Deflection, In. with 12-In. Span	Brinell Hardness		Tensile Strength, Lb. per Sq. In.
	Cr.	FeTi	FeSi	Carbon	Silicon	Titanium			Center	Near Edge	
41	0	0	0	3.21	2.21	-------	4405	0.085	217	217	38,950
42	0	1.0	0	3.12	2.39	0.150	4705	0.09	212	217	40,750
43	0	2.0	0	3.09	2.46	0.262	4670	0.09	229	229	44,250
44	0	0	0.22	3.24	2.18	-------	4505	0.10	207	212	39,000
45	0	0	0.45	3.14	2.31	-------	4597	0.09	212	207	40,050
46	0	0	0.67	3.11	2.35	-------	4445	0.105	197	201	38,000
47	0.25	0	0	3.12	2.08	-------	4540	0.07	220	229	44,500
48	0.25	1.0	0	3.21	2.13	0.105	4997	0.095	223	217	42,350
49	0.25	2.0	0	3.05	2.51	0.202	5020	0.085	229	235	46,400
50	0.25	0	0.22	3.12	2.26	-------	4750	0.085	223	223	40,200
51	0.25	0	0.45	3.12	2.39	-------	4890	0.09	217	217	42,000
52	0.25	0	0.67	3.07	2.39	-------	4635	0.095	207	212	39,800
53	0.50	0	0	3.02	2.14	-------	4590	0.08	229	223	45,750
54	0.50	1.0	0	3.08	2.16	0.090	5142	0.09	223	229	45,300
55	0.50	2.0	0	3.08	2.28	0.188	5435	0.09	241	241	46,650
56	0.50	0	0.22	3.07	2.31	-------	4970	0.085	229	235	45,350
57	0.50	0	0.45	3.08	2.40	-------	4950	0.085	223	223	43,850
58	0.50	0	0.67	3.09	2.47	-------	5025	0.09	223	229	43,700
59	0.75	0	0	3.17	2.35	0.068	5070	0.065	241	241	46,550
60	0.75	1.0	0	3.10	2.38	0.142	5105	0.085	229	241	46,600
61	0.75	2.0	0	3.09	2.41	0.240	5240	0.085	241	248	46,800
62	0.75	0	0.22	3.10	2.28	-------	4958	0.095	241	241	45,450
63	0.75	0	0.45	3.16	2.37	-------	5017	0.08	248	248	48,200
64	0.75	0	0.67	3.08	2.42	-------	5258	0.09	223	235	44,200

taining silicon as compared with the same addition of silicon without any titanium.

15. The values plotted on the charts for hardness are averages of the results obtained from the two different locations on the bar chosen from each heat.

EXPLANATION OF DATA

16. A noticeable feature of all the curves on the three charts is the general upward tendency toward the right of nearly all the full-line curves, indicating an improvement in strength and hardness as the ferrotitanium addition was increased. Although there are some exceptions, it equally is apparent that the dotted-line curves, representing late silicon additions, are either located definitely lower than the full-line curves, or else show a downward direction at their right-hand ends where the titanium curves are rising. The conclusion seems unavoidable that in hardness as well as transverse and tensile strength, increasing additions of ferrotitanium containing silicon were more effective in promoting higher values than equivalent additions of ferrosilicon without the titanium. Therefore, it is evident from these results that titanium was a worth-while addition to these irons, and conferred properties which were not attained with silicon alone.
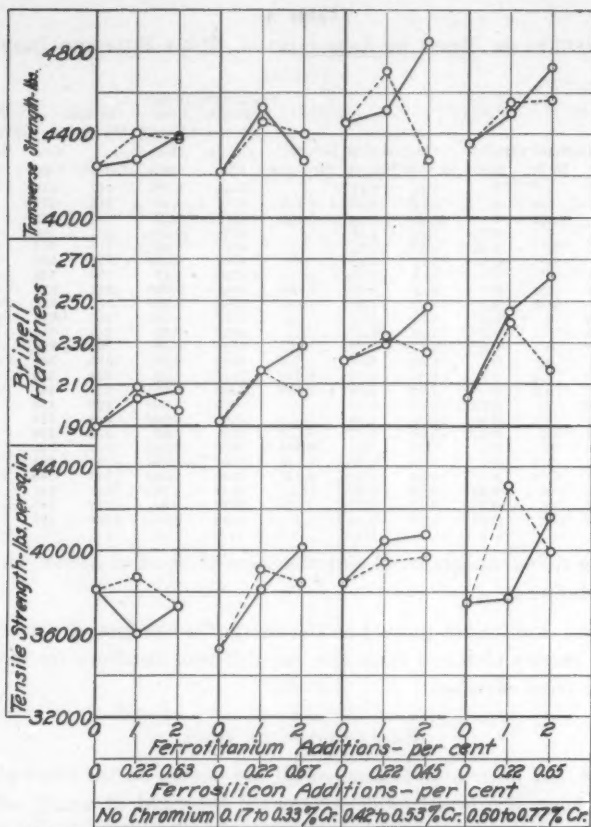
Full Lines Represent Foundry Ferrotitanium Additions
Dotted Lines Represent same Silicon Content from Ferrosilicon

FIG. 1—DATA ON IRON CONTAINING 3.08 TO 3.34 PER CENT CARBON AND
2.00 TO 2.45 PER CENT SILICON.

Effect on Microstructure

17. The reason for the improved properties following the addition of titanium undoubtedly is connected with its well-known effect on the size of the graphite flakes. This effect, illustrated by the photomicrographs Fig. 4 and 5, is believed to account for the increasing hardness produced in these irons by titanium, as well as for the improvements in strength. It has been found that even when enough titanium was added to an iron to increase its content

of ferrite and graphite, the Brinell hardness often likewise was increased, merely on account of the closer grain or finer graphite particles.

CHROMIUM PLUS TITANIUM

18. Chromium has quite a strong refining effect on graphite flakes in cast iron, but not so strong as titanium. When both are
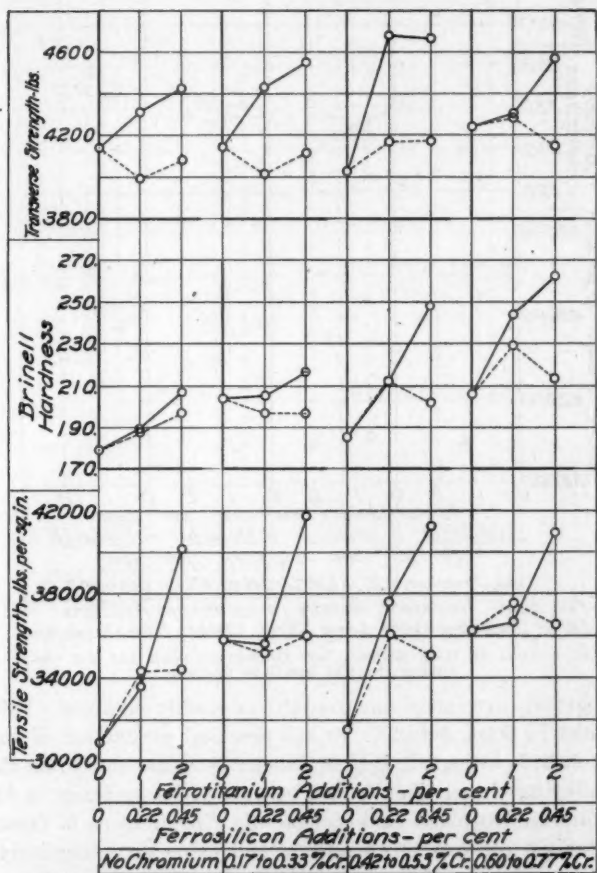
FIG. 2—DATA ON IRON CONTAINING 3.37 TO 3.59 PER CENT CARBON AND 1.77 TO 2.08 PER CENT SILICON.

Full Lines Represent Foundry Ferrotitanium Additions
Dotted Lines Represent same Silicon Content from Ferrosilicon
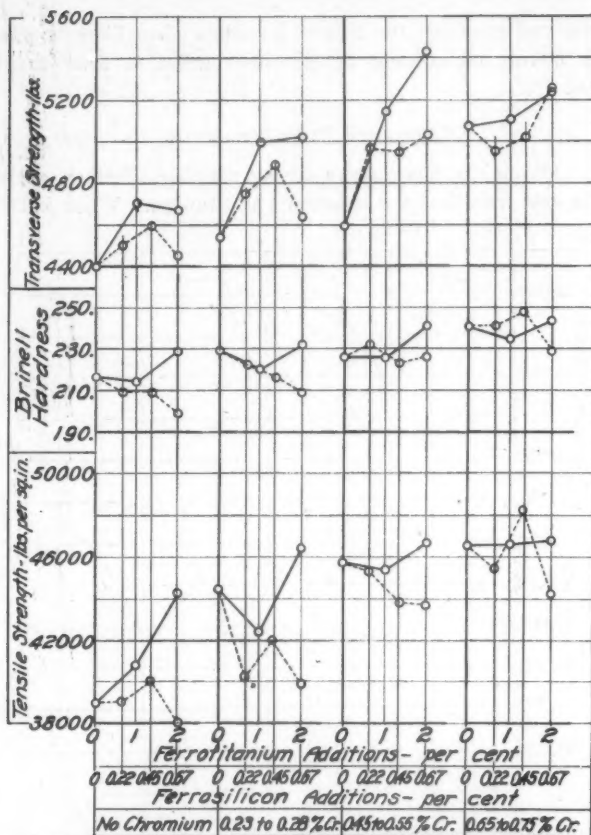
Fig. 2—Data on High Sulphur Iron Containing about 3.10 Per Cent
Carbon and 2.20 Per Cent Silicon.

used together, extremely fine graphite is readily obtained. This is
illustrated by Figs. 6 and 7. In the practical production of small
or thin-walled castings, it may be dangerous to use chromium alone
as an alloying element in the iron because of its tendency to form
hard white iron at places that cool rapidly. The addition of foundry
ferrotitanium corrects this tendency and helps very effectively to
keep the castings uniformly gray.

19. A comparison of Figs. 1 and 2 shows that titanium had
a more definitely beneficial effect on the strength of the higher-
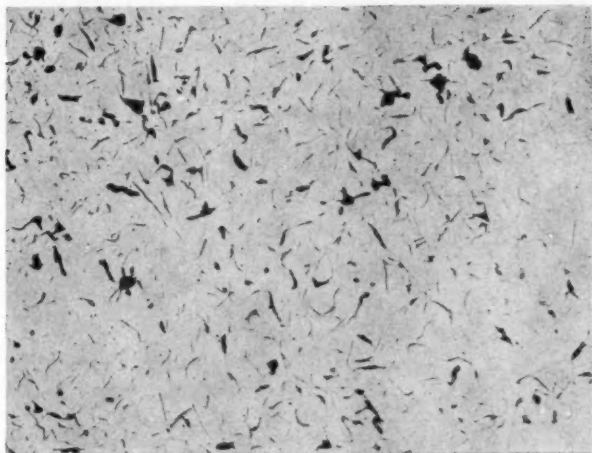
FIG. 4—TYPICAL GRAPHITE IN GRAY IRON TREATED WITH LATE SILICON ADDITIONS;
3.03 PER CENT CARBON, 1.70 PER CENT SILICON, 0.015 PER CENT TITANIUM.
CAST IN 1.25-IN. DIAMETER BARS, UNETCHED, 50X.
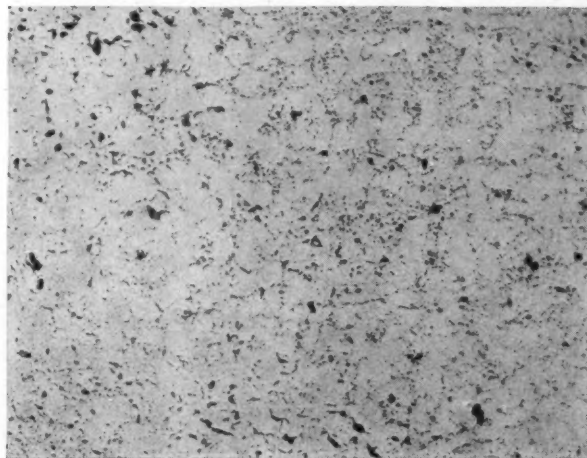
FIG. 5—TYPICAL GRAPHITE IN GRAY IRON TREATED WITH LATE FERRO-TITANIUM ADDITIONS;
3.27 PER CENT CARBON, 1.40 PER CENT SILICON, 0.222 PER CENT TITANIUM.
CAST IN 1.25-IN. DIAMETER BARS, UNETCHED, 50X.
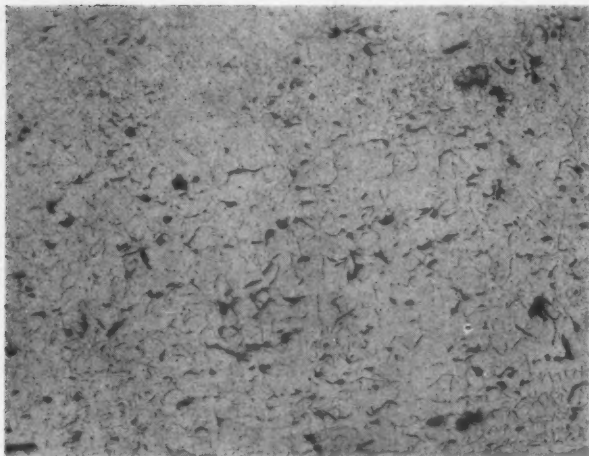
FIG. 6—TYPICAL GRAPHITE IN GRAY IRON TREATED WITH CHROMIUM ALONE; 3.29 PER CENT CARBON, 2.02 PER CENT SILICON, 0.79 PER CENT CHROMIUM, 0.051 PER CENT TITANIUM. CAST IN 1.25-IN. DIAMETER BARS, UNETCHED, 50X.
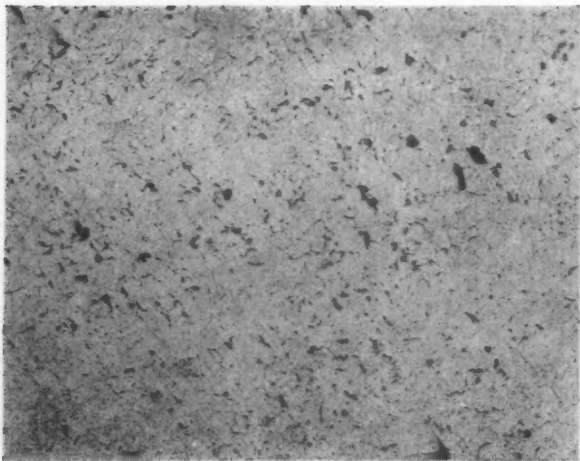
FIG. 7—TYPICAL GRAPHITE IN GRAY IRON TREATED WITH CHROMIUM AND TITANIUM; 3.27 PER CENT CARBON, 2.27 PER CENT SILICON, 0.80 PER CENT CHROMIUM, 0.084 PER CENT TITANIUM. CAST IN 1.25-IN. DIAMETER BARS, UNETCHED, 50X.
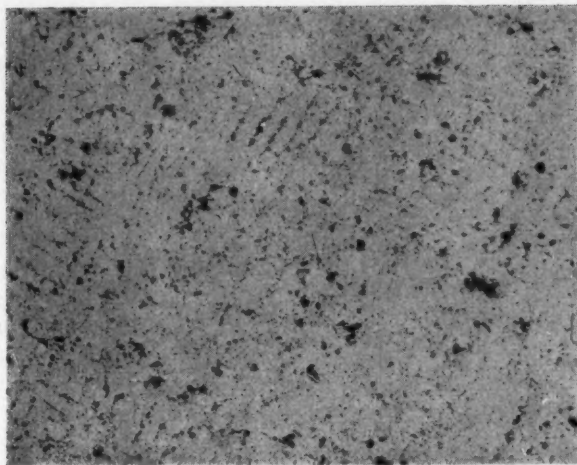
FIG. 8—TYPICAL GRAPHITE IN LOW-SULPHUR CHROME-TITANIUM IRON; 3.04 PER CENT
CARBON, 2.22 PER CENT SILICON, 0.58 PER CENT CHROMIUM, 0.089 PER CENT
SULPHUR, 0.294 PER CENT TITANIUM. UNETCHED, 50X.
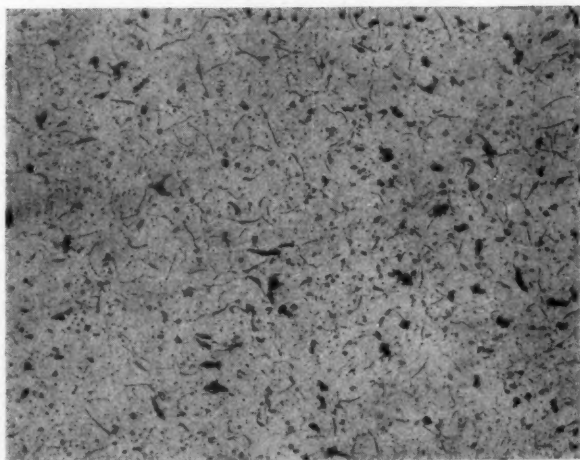
FIG. 9—TYPICAL GRAPHITE IN HIGH-SULPHUR CHROME-TITANIUM IRON; 3.07 PER CENT
CARBON, 2.25 PER CENT SILICON, 0.42 PER CENT CHROMIUM, 0.193 PER CENT SULPHUR,
0.255 PER CENT TITANIUM. UNETCHED, 50X.

FIG. 10—TYPICAL MICROSTRUCTURE OF LOW-SULPHUR CHROME-TITANIUM IRON,
SAME ANALYSIS AS FIG. 8. ETCHED WITH PICRIC ACID, 200X.
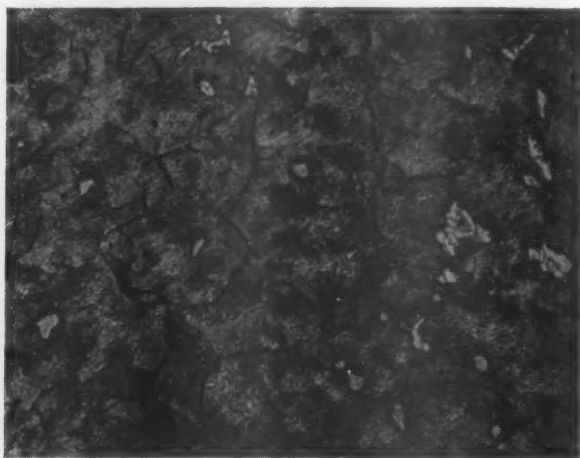
FIG. 11—TYPICAL MICROSTRUCTURE OF HIGH-SULPHUR CHROME-TITANIUM IRON,
SAME ANALYSIS AS FIG. 9. ETCHED WITH PICRIC ACID, 200X.

carbon iron of Fig. 2 than on that of the lower-carbon iron of Fig. 1. This seems to be in line with general experience with titanium in gray iron foundries.

20. In a search for an explanation of this difference, the microstructures were examined, and the lower-carbon titanium irons were found to show a more definitely dendritic structure, as illustrated in Figs. 8 and 10.

Titanium Plus Sulphur

21. After this was determined the paper by Boyles[2] on "The Freezing of Cast Iron" appeared, where the effect of sulphur in eliminating dendritic structure was disclosed. The sulphur in our melts, made in crucibles without scrap iron, was of course very low, as shown in Table 1, and the higher sulphur series of Fig. 3 was then made and tested to check Boyles' results.

22. It was found that the dendritic structure was indeed suppressed in the higher-sulphur iron, with improved results from the titanium treatment as indicated in Fig. 3. Comparative microstructures of the low and high sulphur irons are illustrated by Figs. 8 to 11 inclusive.

23. Since most cupola-melted cast iron has higher sulphur and higher carbon than the low-carbon low-sulphur iron of Fig. 1, the results obtained with titanium in cupola practice should be more closely in accordance with those shown in Figs. 2 and 3 where greater improvements due to titanium are indicated. The chief objection to high-sulphur iron in general practice is its hardening effect, like that of chromium, but in the chromium-titanium irons described, the hardness of the high-sulphur iron was no greater than that of the low-sulphur, while the strength was materially improved. The advantage of titanium in counteracting the chilling tendency at corners or thin sections is as applicable to high-sulphur irons as it is to irons containing chromium.

COMPARISON OF ALL TEST BARS

24. One more point regarding these series of titanium and silicon-treated irons might be of interest. A great deal of trouble was experienced in this work with surface flaws in the cast test bars. These flaws often governed the behavior of the bar in the transverse test, and, as noted above, a bar showing such flaws in

2 Boyles, A., "The Freezing of Cast Iron." A. I. M. E. Technical Publication. No. 809, 1937.

its fracture was rejected for this investigation. In spite of careful attention to the molds used, many heats had to be made over because two of the three bars tested showed surface flaws in their fractures. Some heats also were made over because their analyses did not come within the limits desired.

25. When the work was finished a compilation of all the cast test-bars was made, counting those that showed flaws in their fractures and those that did not, in each of the three classes, untreated, titanium-treated, and silicon-treated. A number of titanium-treated bars were probably counted that were made for other research work. No attention was paid in this compilation to the presence or absence of chromium or sulphur additions which were made at a longer time before pouring than the titanium or silicon addition, but no bars containing other alloys were included. The results are given in Table 5.

Table 5

PREVALENCE OF SURFACE FLAWS IN FRACTURES OF CAST BARS

Late additions	None	Ferro-Titanium	Ferro-Silicon
Total number of bars cast	96.	160.	137.
Number of bars with flaws	57.	49.	73.
Number without flaws	39.	111.	64.
Percentage of bars with flaws	59.5	30.6	53.2

26. There is no possible reason why the molds used for the titanium-treated iron should have been any better, since they were made by men who had no connection with this work, and used indiscriminately, no selection whatever being made. Also the pouring of all the heats was the same. The only explanation for the difference in prevalence of flaws seems to be a scavenging effect of the ferrotitanium addition, involving a change in the character of the included slag in the iron and its better elimination from the castings.

CONCLUSIONS

27. The addition of foundry ferrotitanium to the various kinds of gray cast iron described above has been shown in general to improve its transverse and tensile strength, and to increase slightly the Brinell hardness, with a marked decrease in the size of the graphite particles and a definite improvement in soundness of the castings. It also has been shown that these effects were not

similarly obtained by late additions of ferrosilicon to the iron.
Thus, it is evident that titanium is the effective ingredient in the
foundry ferrotitanium which is used generally in gray iron
foundries.

ACKNOWLEDGMENTS

28.   The thanks of the authors are due to numerous associates
for helping with this work, especially V. V. Efimoff, R. E. Bannon,
and A. S. Yocco who made and tested most of the castings. They
are also indebted to the Titanium Alloy Mfg. Co. for providing the
materials and equipment needed for the investigation, and for
permission to publish the results.

DISCUSSION

Presiding: PRESIDENT H. BORNSTEIN.

R. A. FLYNN[1] *(Written discussion)*: The reason for the structural
improvement of gray cast iron by means of late additions to the melt
is one of the interesting problems confronting the physical metallurgist
today. For it is now a commonly accepted fact that the addition of cer-
tain alloys to the melt, just before pouring, results in castings with
physical properties markedly superior to those of an iron of the same
final analysis, but with the alloy added early in the melt.

An important step in the solution of the general problem is to
evaluate fairly the effect of each alloy that is advocated as a ladle addi-
tion. In the case of titanium, the present paper provides interesting
data. However, the writer does not wholly agree with some of the con-
clusions.

The authors claim that the higher physical properties accompany-
ing titanium and chromium titanium additions are due to the production
of the very fine graphite as shown in Figs. 5, 7, 8 and 10. In this con-
nection, the writer's findings, based on the study of the structures of
more than a hundred heats in which twenty-one different late (ladle)
additions were studied, may be of interest.

The physical properties of irons treated with the various additions
fell into two very well defined groups. The first group of irons exhibited
light gray, equiaxed fractures, with random (normal) graphite distribu-
tion, while the second group showed no improvement over uninoculated
iron of similar analysis-yielding black dendritic fractures with fine
dendritic graphite in the microstructure (similar to Figs. 5, 7, 8 and 10)
and tensile strengths over 10,000 lb. per sq. in. lower than group I.
The reason for the black, dendritic fracture of poorly or uninoculated
iron is not that this iron contains more graphite, but rather that the
fracture surface contains more graphite due to the fracture following
the line of least resistance. (Planes of graphite segregation-dendrites.)

Thus, in Fig. 10, the fracture would follow the chains of fine graphite. The type of graphite shown in Figs. 5, 7, 8 and 10, is, therefore, considered by the writer to be a weak structure due to impotent ladle additions or too great a time interval between addition and pouring. The comparatively small increase in tensile strength due to titanium is believed due to the increased matrix hardness, generally encountered when a carbide forming element is added. When two irons of similar analysis, differing only in time of addition, are compared, and if a potent addition is used, tensile values may increase 10,000 to 15,000 lb. per sq. in. in favor of the late addition, and with no increase in Brinell hardness. Hardness is more a function of matrix hardness while tensile strength is very sensitive to directional weakness (graphite segregation).

The above discussion is not meant to convey the impression that titanium content is undesirable in a ladle addition. Several of the additions studied were high in titanium content and one was quite successful. Many other considerations in addition to actual alloy content contribute to the potency of ladle additions (e. g., time before pouring, temperature at which addition is made, particle size of inoculant, minor element content, gas content and gas analysis of melt, analysis of melt, and others).

Several details of technique might be profitably considered here. The writer believes that the effect of the late additions was greatly decreased by the practice of adding the ferro-titanium or ferro-silicon, continuing the heating for two minutes and then cooling down for an (unstated) interval of time. At this laboratory, actual foundry practice was simulated by pouring into a hand ladle and adding the ladle addition at the base of the stream of metal. In the case of both silicon and titanium alloys, recovery was always above 90 per cent. Where strength and freedom from flaws were required, a higher pouring temperature (2600°F. for 3.2 per cent T.C. iron) was found pre-requisite. A.S.T.M Standard Type "B" bars, 1.2 in. x 21 in., were poured in this study. The characteristically adherent slag of titanium melts tends to force more careful skimming technique upon the melter, and the writer wonders if this might not have resulted in fewer imperfections in the titanium treated heats. The effect of sulphur was studied by the writer and it was his conclusion that a good ladle inoculant is far superior to sulphur addition to convert a dendritic to an equiaxed structure.

V. A. CROSBY[2] (Written discussion): The authors are to be congratulated on the large amount of work involved in preparing this paper and on the scientific contribution it offers on the subject of graphitizing agents.

It does not appear that any conclusions can be drawn from Table 2, involving 20 heats; and if one must split hairs to establish a claim, it would appear that heats treated with late additions of silicon (Heats Nos. 17 and 18 versus 19 and 20) showed somewhat better tensile values for the late silicon.

Table 3, showing results in high carbon cast irons indicate that

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[2] Metallurgical Engineer, Climax Molybdenum Co., Detroit, Mich.

titanium additions are more potent than equivalent late additions of silicon. However, in this series it is a matter of great speculation as to what became of the late additions of silicon. For instance, Heat No. 24 contained 1.97 per cent silicon after 0.22 per cent FeSi addition, and contained 1.96 per cent silicon after 0.45 per cent FeSi addition. The same thing occurred in heats No. 29 and 30. It appears nevertheless that the data in Table 3 is in line with previous investigations, indicating that titanium is more effective in higher than in lower (3.00 per cent) carbon irons.

The data of Table 4 rather proves that the authors were working throughout the whole series with irons too high in original silicon, to derive much improvement in strength by a late addition of silicon or titanium. It is to be noted that chromium additions produced quite consistent improvements in all three series of test, and by way of explanation, it is the writer's opinion that the base iron was too soft and probably contained some free ferrite.

Hence, the addition of sulphur plus chromium (two carbide stabilizers) produced an iron of such a character that a late addition of silicon then resulted in the highest strength recorded in the series (Heat No. 63).

The writer's experience indicates that the chief advantage derived from a late addition of silicon is obtained by making this addition to a white iron or one containing a generous amount of free cementite. In the case of gray irons of eutectoid carbon content or less, the advantage of late additions of silicon results in changing the graphite characteristics from the dendritic fine variety to the normal flake formation with an attendant increase in transverse, deflection, impact, and tensile, along with a slight decrease in hardness.

DR. R. SCHNEIDEWIND[3]: While this paper was received too late for me to prepare a written discussion, I would like to say that I do not believe we realize how much work was involved in the research described here. There are over 393 bars reported upon:

As to the technique followed, I am wondering whether the method used will be comparable to commercial practice. A melting temperature of 2670°F. seems a little low in our experience at the University.

Concerning the data of Tables 4, 5 and 6, I would like to know whether the chromium additions were added at a time to give the maximum improvement. In Table 2, without chromium, the tensile strength is 38,000 lbs. per sq. in. With 0.25 of chromium, it is 35,000; with 0.50 chromium, it is 38,000; and with 0.75 chromium, it is 37,000 lbs. per sq. in.

Our experience has been that chromium does have some effect on strengthening irons in this section size, and our work with titanium has indicated that titanium is beneficial in the low carbon ranges or, rather, in the low carbon plus one-third silicon ranges. The carbon plus one-

third silicon range here is from 3.83 to 4.23 per cent total, in which I believe that titanium is not as effective as in the lower carbon irons.

Although the tensile strength increases in many cases, for example, irons 36 and 38, the tensile strength has increased 10 per cent, whereas the brinell hardness has increased from 201 to 269, which is not desirable. On the other hand, the ferro-silicon additions did not markedly change the tensile strength.

But most important, in studying the transverse strengths, although the transverse strength of irons containing additions of titanium, goes up slightly, it will be noticed that the transverse deflection goes down markedly. In Table 2, with the titanium additions, the loss in transverse deflection varies from 5 per cent to 23.5 per cent, whereas the late silicon additions have caused a decrease in deflection from 3 per cent to an increase of 27.5 per cent, and that marked effect in the decrease of transverse deflection through use of titanium carries through all of the tests in the paper and, in general, the ferro-silicon has increased it.

Now, we do not know whether it is strength or the deflection we need. I wonder whether the conditions of tests have brought out as favorably as possible the beneficial effects that titanium can bring out.

G. F. COMSTOCK and E. R. STARKWEATHER *(Author's Written Closure)*: We are interested in the views set forth by Mr. Flinn, and appreciate his expressing them in connection with our paper. It appears, however, that Mr. Flinn does not differentiate between fine graphite and dendritic structure, while we feel that these features are distinct and need not occur together. We agree with his feeling that the dendritic structure is comparatively weak, but believe that finer graphite is a source of strength and that titanium will strengthen iron in that way if the dendritic structure is not too prominent. Many practical tests in various foundries have furnished evidence in support of that contention.

We regret that Mr. Flinn did not see fit to disclose at least some of the data he apparently has acquired regarding the effects of titanium additions to cast iron, as it would seem that such data would have been of considerable general interest in connection with our paper. We hope he will publish them later.

We do not agree with Mr. Flynn that better results would be expected when titanium is added in a small hand-ladle immediately before pouring, or that such practice is more similar to regular foundry practice than adding the titanium two minutes before pouring. In nearly all practical experience with titanium cast iron it has been found better, on the basis of both recovery and physical properties, to add the titanium at the cupola spout so that it has about two minutes time for thorough distribution and scavenging in a large ladle, rather than to use the alloy in small hand ladles. We should like to emphasize again that in our experimental work all the crucibles were skimmed with equal care, the constant aim being to produce test-bars as free from flaws as pos-

sible from every melt to avoid the trouble and expense of making the heats over again.

We are grateful to Dr. Schneidewind for his interest in the paper, and believe that most of his comments have already been answered. His discussion of the deflections which we reported may be theoretically correct according to the values shown, but we believe that it is difficult to draw reliable conclusions from deflections measured with only a 12 in. span, since the results obtained are so small. For instance, in Table 2, to which he refers, the extreme range of deflection values reported is 0.088 to 0.125 in., and this limited range is the reason why the deflections were not plotted on our charts. It is true, however, that the silicon additions which gave lower strength than titanium gave in general somewhat higher deflection.

Mr. Crosby's discussion is also much appreciated. Our reply to his criticism of the silicon values shown in Table 3 is already embodied in paragraph 9 of the paper, so that a further reply seems unnecessary. It is admitted that Mr. Crosby may be right in objecting that the original silicon contents of our irons were too high for the highest possible strengths, but it is not evident that this affected seriously the comparison between titanium and silicon which was the object of the research. We wonder if Mr. Crosby meant to imply in his last sentence that low-carbon irons without late additions of either silicon or titanium always have fine graphite in a dendritic arrangement; this is not in accordance with our experience.

It becomes more certain as evidence accumulates that the exact control of the structure of gray cast iron is a difficult but important matter, and that a large number of significant factors must be considered. It is gratifying to note that all those who were kind enough to discuss our paper, agree in considering titanium as one of those useful factors.

Modification of the Saeger Fluidity Test
As Applied to Red Brass

By GEORGE P. HALLIWELL*, CHICAGO, ILL.

Abstract

A modification of the fluidity test, as developed by Saeger and Krynitsky, has been used for the determination of the flowing power of red brass. The method consists in permitting metal under a definite head to flow from the bottom of a preheated crucible into the fluidity spiral.

INTRODUCTION

1. The ability of a molten metal or alloy to completely fill a mold has long been of prime importance to the foundryman. Lacking a ready means to measure this property, he has used various expedients, such as change in mold design, gatings, sand conditions, pouring temperature, metal composition or the use of a "fluidizer", all of which have served their immediate purpose, either collectively or individually.

2. It is becoming increasingly more important, however, to know before pouring whether a metal or alloy has the required fluidity to make suitable castings and to know to what extent the presence of minor quantities of other elements, changes in melting procedure, or molding conditions will affect this property.

PREVIOUS WORK

3. It is not the purpose of this paper to discuss the factors that affect fluidity as that has been done by others** nor to outline the development of the fluidity test in its present state, as represented by the spiral of Saeger and Krynitsky[2]. A brief resume of this has been given by these authors and a more complete bibliography given by the British Iron and Steel Institute[1] and the Fluidity Committee of the American Foundrymen's Association[3].

PRESENT WORK

4. About two years ago, the company with which the author is connected became interested in the effect of changes in composition and of the presence of small amounts of beneficial and detrimental impurities upon the fluidity of red brass, nominally 85 per

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cent copper with 5 per cent each of tin, lead and zinc. In studying this problem, we used the method of Saeger and Krynitsky[2] shown in Fig. 1. Molten metal poured into the receiving basin (4) flows through the down gate (3) and horn gate (2) into the spiral (1). An overflow is attached to the receiving basin to maintain a constant pouring head.

DIFFICULTIES ENCOUNTERED IN USE OF SAEGER METHOD

5. In the beginning of the work, great difficulty was experienced in obtaining check results in the same heat or from successive heats of metal of approximately the same composition. It was recognized early that moisture content and other properties of the sand, the amount of ramming, and the temperature of the metal were of such vital importance that any deviation from the normal or average conditions of one test might easily account for

SECTION A---A

1—SPIRAL
2—HORN GATE
3—DOWN GATE
4—POURING BASIN
AND OVERFLOW

DIMENSIONS IN INCHES
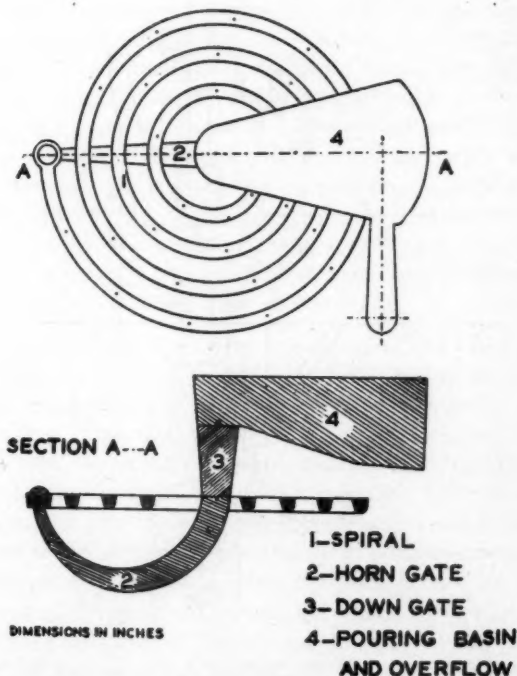
FIG. 1—ASSEMBLY OF FINISHED PATTERN USED FOR DETERMINING THE RUNNING QUALITIES OF METAL (AS PUBLISHED BY SAEGER AND KRYNITSKY, TRANS. A.F.A. (1931), VOL. 39, P. 522.)

irregularities in results. Particular effort was made to keep these factors as nearly constant as possible, but still inconsistent results were obtained.

6. It was finally decided that the most difficult factors to control were: (1) An accurate determination of the temperature of the metal as it entered the spiral; and (2) maintenance of a uniform rate of pour. A chromel-alumel thermocouple (No. 16 gauge) welded into the closed end of a Marshall protection tube (heat resisting alloy) was used with a Brown millivoltmeter and compensating leads to measure the temperature of the metal in the crucible. Cold junction temperature corrections were made manually on the instrument. Subsequent attempts to determine the temperature of the metal as it entered the spiral, by embedding a bare thermocouple junction in the bottom of pouring gate (*F* of Fig. 2) were unsuccessful. The inertia of the instrument needle

FIG. 2—FLUIDITY SPECIAL ASSEMBLY USING BOTTOM POUR CRUCIBLE.

prevented its easy adjustment to the rapid change in temperature from that of sand to that of the molten liquid before the latter solidified.

DOUBLE CRUCIBLE

7. Realizing that it would be impractical to determine the actual temperature of the fluid metal as it entered the spiral, recourse was made to a double crucible to minimize any temperature drop during pouring. A No. 12 crucible was inserted in a No. 20, and the annular space filled with crushed fire brick. A suitable heat resisting cement held the latter in place. The temperature drop with this method was never greater than 20°F. (while pouring six fluidity molds). While this procedure minimized the temperature error, it accentuated another factor, namely, the ease of pouring.

RATE OF POUR

8. Any improvement in the regulation of the temperature drop may be easily vitiated by a non-uniform rate of pour. A sudden surge or premature flow of the liquid metal from the pouring basin (4 of Fig. 1) changes the effective height of the head of metal and gives erroneous results. The use of a double crucible involving a total weight of 125 to 150 lb. with two men pouring increases the difficulty of maintaining a uniform rate of pour. If a rapid rate is used, there is greater danger of splashing metal into the down gate before the proper head is maintained. A slow rate of pour gave a shorter spiral, hence less sensitivity, without any increase in uniformity of results.

9. In the Saeger-Krynitsky spiral, a small overflow finger is used to regulate the proper head of metal. It has been our experience that it does not accurately serve the intended purpose. In attempting to use their method, we poured at various rates, but in every case the results seem to indicate that the spiral had solidified before the metal had filled the pouring basin to the overflow finger. The pouring well also was deepened and the angle of the incline to the downgate increased to prevent splashing or premature surging of the metal down the horn gate. This modification gave slightly improved results but still irregular. A further modification substituted a "teapot spout" type of gate similar to that used by Pilling and Kihlgren[4]. Results here also were inconsistent, due to our inability to control the rate of pour.

10.   The type of results obtained with the Saeger and Krynitsky spiral, using the deepened pouring well and the "teapot spout" type of gate, are shown in Table 1. These are typical of the variations found. Each group represents five or more spirals poured in succession from a single pot of metal.

Table 1

RESULTS OF FLUIDITY TESTS WITH SAEGER AND KRYNITSKY SPIRAL
AND WITH MODIFICATIONS OF POURING WELL AND GATE.

No.	Type of Mold	Temp. of Metal, °F.	Full Cross Section	Thin Cross Section	Total
Group A					
1	Skin Dried	2170-80	20½	2⅞	23⅜
2	Skin Dried	2170-80	18	1⅞	19⅞
3	Green	2170-80	18	¾	18¾
4	Skin Dried	2170-80	23¾	⅝	24⅜
5	Green	2170-80	13⅛	1⅜	14½
Group B	Type of Pouring Well*				
1	S	2025-35	7	⅞	7⅞
2	E	2025-35	8	⅛	8⅛
3	S	2025-35	10	4¾	14¾
4	E	2025-35	8½	3½	12
5	S	2025-35	14	2	16
6	E	2025-35	9½	7½	17
Group C					
1	E	2200-20	17⅛	1⅛	18¼
2	E	2200-20	20⅞	5¼	26⅛
3	E	2200-20	24	2	26
4	E	2200-20	25¼	1⅝	26⅞
5	E	2200-20	23½	6⅞	30⅜
6	E	2200-20	23¾	3	26¾
Group D					
1	T	2190-2200	29½	1¾	31¼
2	E	2190-2200	30⅜	2⅜	32¾
3	T	2190-2200	31	1½	32½
4	E	2190-2200	23½	4	27½
5	T	2190-2200	27	1½	28½
6	E	2190-2200	25½	2½	28
Group E					
1	E	2185-2200	20¼	½	20¾
2	T	2185-2200	24½	2½	27
3	E	2185-2200	28½	3	31½
4	T	2185-2200	23¾	¼	24
5	E	2185-2200	36	3½	39½
6	T	2185-2200	23¾	1⅛	24⅞
Group F					
1	E	2120-35	23	3	26
2	T	2120-35	28½	¾	29¼
3	E	2120-35	28	3½	31½
4	T	2120-35	23½	⅜	23⅞
5	E	2120-35	21½	¾	22¼
6	T	2120-35	29¼	3½	32¾

* S=Standard Pouring Well.
  E=Enlarged Pouring Well.
  T=Teapot Spout Type of Gate.

MEASUREMENT OF SPIRAL

11. In all fluidity spirals, there is a length at the end which does not fill the entire cross-section of the mold. The surface of this metal is irregular and oxidized, and the line of demarcation between this section and that which completely fills the mold is well marked. This latter area projects into the unfilled section as an elongated arrow head and represents that portion which just failed to fill the mold. The length of the spiral terminating in this arrowhead has been taken as a measure of the flowing power of the alloy, and indicated in the tables as "full cross section." From the above data, it will be seen that there is no regularity in results even from metal in the same pot. All preliminary work seemed to indicate that the cause of these variations lay in the technique of pouring. To one skilled in the practice of pouring fluidity castings, concordant results might have been obtained, but there is considerable difference in the technique of pouring fluidity spirals under laboratory conditions and commercial foundry conditions.

PRESENT METHOD

12. To minimize the personal equation as much as possible and to maintain a uniform flow of metal, a method of bottom pouring was devised similar to that used by Saito and Hayashi[5]. It is simple in principle, but requires the coordination of two operators. Metal from a ladle or crucible is poured to a definite height into a bottom pour receiving crucible, (A of Fig. 2), the tapered hole of which is closed with a graphite plug (B). The latter cemented to a stainless steel rod (C) is then withdrawn and the metal flows out from an initially constant head at a uniformly decreasing rate. Both receiving crucible and plug are heated to approximately the same temperature as the metal. Early experiments with a cold crucible and plug showed that not only was the temperature drop too rapid, but that frequently the plug froze at the bottom and spoiled the test.

13. The details of shape and dimensions are shown in Fig. 2. It will be noted that the horn gate is not used. In our early experiments with this method, but with the horn gate, we found that the metal quite frequently broke through from the first to the second round. While this might have been prevented by different conditions of founding, we eliminated this trouble by using the "teapot spout" type (D) previously mentioned. The metal enter-

ing the spiral (E) flows in at an angle but with its horizontal component in the same direction as the start of the spiral, while with the horn gate the horizontal component is zero. The method described in this paper causes less retardation of the metal in its flow along the fluidity channel by taking advantage of this horizontal component.

14. While no attempt has been made to determine any relationship between the length of the spiral and the size of the orifice in the crucible, a variation of 3/32 in. appeared to have no effect on the results. Care, however, must be taken to see that the plug seats closely in the hole so that no metal flows through before the proper time. It is best to remove all metal that adheres to the wall of the crucible or the plug before preheating, otherwise it will melt and prevent proper seating of the plug.

15. The procedure in this method was as follows:

(1) An ingot of metal (20 to 25 lb.) was melted under lump charcoal in a No. 12 crucible.

(2) When the metal was approximately at the temperature desired, the receiving crucible and plug were inserted in the same furnace for preheating.

(3) When the receiving crucible and plug were hot, the crucible containing the molten metal was removed and skimmed, keeping the receiving crucible and plug in the furnace under a low heat.

(4) As soon as the thermocouple was inserted, the receiving crucible was removed and placed over the gate of the mold, and the plug inserted.

(5) While one operator has completed the above operation, the other has finished taking the temperature and is ready to pour. To be sure that the metal in the receiving crucible was not chilled by passing over the cool lip of the melting crucible, about one fourth of it was first poured into an ingot mold.

(6) When the correct head of metal was attained, the plug was pulled at a given signal from the melter. These latter operations are easily synchronized, and keep the time interval between taking the temperature and pulling the plug to about 10 to 15 sec.

16. In a critical analysis of this method, it must be remem-

bered that we were seeking either to maintain a constant head of metal or to reduce to a minimum the personal equation involved. The method is not infallible, and several modifications have been suggested which, due to lack of time, could not be tried. In the present method, care must be taken to seat the hot crucible properly on top of the mold, so that dirt will not be dislodged and fall down into the pouring well. The crucible also must be centered so that the metal, as it flows out, falls directly into the down gate (F) and does not hit the sides of the receiving well (G). Both these operations have been facilitated by molding a seat, (H of Fig. 3) on top of the pouring basin.

OVERFLOW BASIN

17. Since the amount of metal used is more than required for the spiral, provision had to be made for the overflow of this excess. Without this, part of the metal momentarily remains in the receiving crucible and creates an artificial pressure sufficient to produce various results. Furthermore, the excess metal ultimately splashes on the floor. In using an overflow gate (I) and basin (J) such as is shown in Fig. 2, it is necessary to have the throat sufficiently large to permit ready exit of excess metal without the production of any back pressure. Since the time interval between
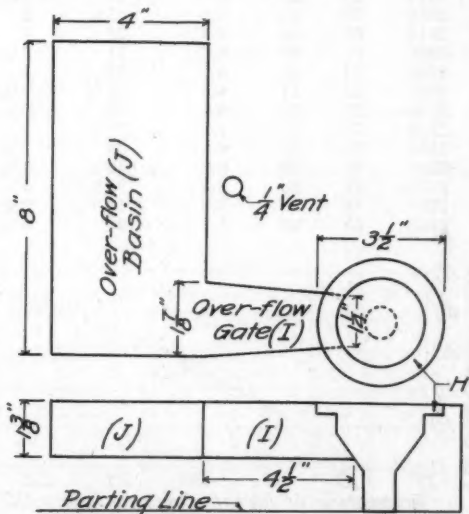
FIG. 3—POURING WELL, OVER FLOW GATE, AND BASIN ASSEMBLY.

the withdrawal of the plug and emptying the crucible and the subsequent formation of the fluidity spiral is only a few seconds, it will be seen that the prevention of such back pressure is necessary.

NUMBER OF POURS

18. Limited equipment prohibited a more ideal method of performing these fluidity tests. It is recognized that it would have been better to have poured as many tests as possible from a double

Table 2
RESULTS OF FLUIDITY TEST WITH MODIFIED METHOD USING BOTTOM POUR CRUCIBLE.

| No. | Charge | Temperature °F. | | | Moisture in Sand Per Cent | Length in Inches | | |
		Metal	Air	Sand		Full Cross Section	Thin Cross Section	Total
71	9	1820	25	19	5.9	7¾	0	7¾
64	8	1890	20	19	7.2	10½	¾	11¼
59	8	1910	31	27	6.6	14¾	7	21¾
60	8	1910	33	28	6.6	13¾	5	18¾
45	5	1920	24	20	6.6	19	4½	23½
55	6	1940	27	22	6.4	21½	1	22½
58	8	1950	31	27	6.6	20½	3½	24
50	7	1950	27	23	5.6	21½	3½	25
43	5	1970	23	20	6.9	21	2⅜	23⅜
46	5	1980	24	20	6.6	24	4	28
53	6	1990	25	20	6.4	21	1	22
47	7	2000	25	17	6.6	28	3½	31½
44	5	2010	20	22	6.9	24	3½	27½
52	7	2010	28	25	5.6	25½	2	27½
54	6	2010	25	20	6.4	28	1	29
49	7	2060	27	23	5.6	31	3¼	31¼
68	9	2060	27	20	6.3	33	4	37
65	9	2080	22	17	6.3	31½	3½	35
56	6	2120	28	22	6.4	33	2¼	35¼
62	8	2135	18	17	7.2	36½	3	39½
57	6	2140	28	25	6.4	38	1½	39½
66	9	2160	21	22	6.3	42½	3½	46
67	9	2170	27	23	6.3	38¾	4½	43¼
63	8	2200	18	18	7.2	40	2½	42½
70	9	2225	25	20	5.9	43	8½	51½
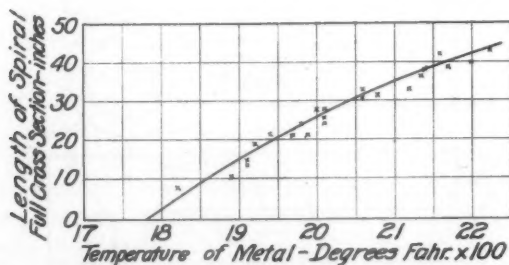
FIG. 4—FLUIDITY OF 85-5-5-5 ALLOY USING BOTTOM POUR CRUCIBLE WITH SAEGER KRYNITSKY SPIRAL.

walled crucible, where both composition and temperature could have been controlled more closely. It is believed, however, that with the technique used, the temperature drop from melting to receiving crucible is not much greater than would be obtained with a double walled crucible.

DATA

19. The results obtained are shown in Table 2, and the accompanying chemical composition in Table 3. Fig. 4 represents these data in graphical form. It is interesting to note that extrapolation to zero fluidity cuts the temperature axis at about 1775°F. which is well within the liquidus-solidus range. An average of all the points on the curve shows that there is a variation of about $+$ 2 in. This is considerably better than that obtained by us with the previous method. With better sand control than is available at our laboratory foundry, this average could undoubtedly be improved.

20. At the highest tmperature used, 2225°F., the full spiral was 43 in. long with an 8 in. tail. With more fluid metals and higher temperatures, it is probable that the complete spiral would be filled. In such case, it would be necessary only to use a lesser head of metal, or to use a spiral of larger diameter.

COMPOSITION

21. Minor changes in basic composition appear to have no noticeable effect upon the fluidity. The presence of 0.5 per cent nickel in all samples will not affect the fluidity within the series, although Kihlgren[6] claims that the fluidity is increased over that of alloy containing no nickel. It is expected that with this im-

Table 3

COMPOSITION OF REPRESENTATIVE SPIRALS IN EACH CHARGE
USED IN TESTS.

No.	Charge	Temp. of Metal, °F.	Cu. %	Sn. %	Pb. %	Zn. (Dif.) %	Ni. %	Fe. %
44	5	2010	85.33	4.90	5.39	3.66	0.55	0.17
46	5	1980	85.44	4.95	5.39	3.50	0.55	0.17
57	6	2140	85.10	4.94	5.29	3.97	0.55	0.15
55	6	1940	85.10	5.00	5.30	3.90	0.55	0.15
51	7	2120	85.21	4.94	5.40	3.73	0.55	0.17
50	7	1950	85.56	5.00	5.27	3.45	0.55	0.17
63	8	2200	86.13	4.62	5.21	3.37	0.55	0.15
64	8	1890	85.69	4.71	5.34	3.55	0.55	0.16
70	9	2225	84.94	4.96	5.24	4.13	0.55	0.18
65	9	2080	84.94	4.96	5.28	4.09	0.55	0.18

proved method of introducing the molten metal into the spiral, the effect of other elements or fluidizers may be better studied.

22. While we feel that the method is an improvement over that of Saeger and Krynitsky, although a little more involved, it is by no means impractical even in an average foundry. It is our sincere hope that others will be influenced to test its applicability and report their findings.

Conclusion

23. In conclusion, it may be said that a method of testing the fluidity of metals has been devised in which:

(1) The personal equation involved in pouring has been reduced to a minimum.

(2) The temperature drop from pouring crucible to spiral is probably less than in other methods.

(3) The sensitivity is increased.

Acknowledgment

Bibliography

1—*"Bibliography on the Fluidity and Viscosity of Metals,"* Bibliographical Series No. 1, British Iron and Steel Institute.

2—Saeger, C. M., Jr., and Krynitsky, A. I., *"A Practical Method for Studying the Running Quality of a Metal Cast in Foundry Molds,"* Trans. A.F.A., vol. 39, pp. 513-532 (1931).

3—*"Progress Report of Gray Iron Division Committee,"* Trans. A.F.A., vol. 45, pp. 33-38, (1937).

4—Pilling, N. B., and Kihlgren, T. E., *"Casting Properties of Nickel Bronzes,"* Trans. A.F.A., vol. 40, pp. 289-305, (1932).

5—Saito, D., and Hayashi, K., *"Investigation of the Fluidity of Metals and Their Alloys,"* Memoirs of the College of Engineering, Kyoto Imperial University, (1919) vol. 2; (1924) vol. 3.

6—Kihlgren, T. E., Discussion of Paper by Saeger and Krynitsky, Trans. A.F.A., vol. 39, p. 535, (1931).

DISCUSSION

Presiding: HAROLD J. ROAST, Canadian Bronze Co., Ltd., Montreal, P. Q.

H. W. MAACK[1] *(Written discussion):* Our experience (with a modified "Sipp Fluidity Spiral") parallels Mr. Halliwell's early tests, in that variations in the pouring operation gave such wide differences in results that we were unable to obtain reliable information about the effect of composition and temperature on the degree of fluidity of our various alloys. It is apparent from the tabulated results that Mr. Halliwell was able to overcome largely these difficulties by use of the pouring arrangement described. We hope that these tests may be continued in order to obtain data that will show the effect on fluidity, of variations in composition of the principal constituents and impurities of red brass, as well as the effect of differences in melting conditions and methods of treatment prior to pouring (deoxidation, etc.).

Mr. Halliwell's modification of design of the mold to direct the metal horizontally along the run of the spiral is a good one in our opinion. One point which is not clear is the exact location of the vent shown on the sketch. We would like to have more detailed information about this.

We question the use of a vent in molds for fluidity tests. In actual production it is seldom practical to place vents so they will be as effective as the ¼ in. vent might be from the author's sketch. Besides, it seems that the effect of the vent might be quite different for brasses of dissimilar compositions; 85-5-5-5 for example compared with brasses containing about 20 per cent of zinc. It is not likely that the high zinc alloy—throwing off more vapor—would be cast under more favorable conditions in a vented fluidity mold than one with lower zinc.

T. E. KIHLGREN[2] *(Written discussion):* Those interested in metal casting problems will agree that a method of measuring the fluidity of non-ferrous foundry alloys which requires no special skill and in which the personal equation is eliminated, would make it possible to correlate data from various sources and be generally useful. The author is to be commended on his effort in this direction, although he will, no doubt, concur that further improvement and simplification of test procedure and fluidity casting design is desirable. The paper should stimulate further work in this direction; in such work it should be recognized that a number of factors, including the condition and temperature of the melt, the nature of the mold material and similar variables, must be controlled in addition to the pouring of the spiral if reasonable accuracy is to be achieved.

If the data of Table 2 are plotted for each charge number on a scale such as to give a slope of about 45°, and a curve drawn through each set of points, it will be found that the scatter of points is con-

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[1] Chief Chemist and Metallurgist, Crane Co., Chicago, Ill.
[2] International Nickel Co., Bayonne, N. J.

siderable, particularly between 1900 and 2100° F. When all the points are considered together on one plot, the scatter is pronounced and spirals may vary as much as 8 in. in length at 2000° F. This scatter is also noticeable on the curve of Fig. 4, and suggests that a further refinement of method is needed before the effect of various impurities and alloying additions can be evaluated without a multiplicity of tests.

We have had occasion to run a large number of fluidity tests both on a wide variety of bronze mixtures and on 20 per cent nickel silvers containing varying amounts of tin, zinc and lead, using a method developed at our laboratory* and have had little difficulty in obtaining suitable temperature-spiral length (T-L) curves provided certain precautions are observed. No doubt, the author's difficulty in obtaining consistent results with this method was due, as he suggests, to inability to control the rate of pour. The method we have used requires that the spiral be poured from a large mass of metal (100-150 lb. melts) and the mold be filled very rapidly. Any attempt to pour the mold cautiously will result in irregular T-L curves that are useless. We have also found it desirable to pour a considerable number of spirals from each melt, over a range of 300-400° F. Also, inasmuch as temperature control is very important—a change of 50° F. altering the spiral length by 6-10 in. —we have found it useful to keep a record of the elapsed time during pouring. By plotting the time at which each casting is poured against the observed temperature, a smooth time-temperature cooling curve can be constructed which affords a check on the accuracy of the temperature measurements. Usually no difficulty is encountered in drawing a smooth curve through most of the points, and if only one or two out of 8 to 10 measurements are out of line, a valid correction can be applied, based on the curve. When the scatter of points is general and a smooth curve cannot be obtained, due to faulty contacts in the potentiometer or thermocouple contacts or some other reason, the data must be discarded. Incidentally, this procedure has been followed in all our experimental foundry work and has been found to give an excellent check on pouring temperature readings. It is not particularly feasible in general foundry work, although even there a check of temperature at the start and finish of pouring of commercial castings would be worth while.

Another precaution which we have followed in comparing the effects of alloying additions on the fluidity of the base mixture is to cast the alloys to be compared on the same day using the same heap of molding sand, the same melting conditions, thus cancelling several variables which might influence the results.

We have mentioned these precautions because we feel that until a method is developed which permits a closer duplication of results in single tests at a given temperature than has yet been devised, the most dependable results will be obtained by observing some of these factors, and especially by pouring a number of spirals from each heat over a

* *Physical and Casting Properties of Nickel Silvers*, Trans. AIME, vol. 117, p. 297, ·935, and *Casting Properties of Nickel Bronzes*, Trans. AFA, vol. 40, 1932.

range of temperatures, so that errors in measurement of temperatures, and variations in technique can be properly weighted.

Regardless of the method used for evaluating fluidity, the value of fluidity tests in general is debatable unless the test data can be found to bear a definite relation to the production of castings. Presumably, if it be found that an alloying addition improves the fluidity, it should then be possible to extend the pouring temperature range within which pressure tight castings can be obtained to a lower temperature. This suggests that by pouring small pressure castings in commercial shapes over a wide range of temperatures and ascertaining whether any differences in the bottom limit of pouring temperatures, as shown by hydraulic and fracture tests, are obtained, the validity of the fluidizing effects observed in fluidity tests would be checked.

We were interested in noting that all the melts contained 0.55 per cent of nickel. It is reasonable to expect, as the author suggests, that since the nickel content remained constant in all the melts, it will not affect the results within the series. We have found, however, that about this amount of nickel is quite useful in improving the fluidity of red brass. This effect was first reported by a manufacturer of bronze castings, on the basis of practical foundry tests on a modified red brass, in which it was found that good castings could be obtained at temperatures 100°F. below that which had been previously used in the nickel-free mixtures.

C. W. BRIGGS[3] *(Written discussion):* Attention is called to the author's conclusions:

1. The personal equation involved in pouring has been reduced to a minimum.
2. The temperature drop from pouring a crucible to spiral is probably less than in other methods.
3. The sensitivity is increased.

In regards to the first point, it is conceded that bottom pouring is the more accurate method providing the stopper is raised sufficiently to allow for a full stream of metal to exit from the nozzle. Comparable results can be obtained, however, with lip pouring if the distance between the mold and the crucible remains constant and the crucible is tilted by some controlled means such as an electric hoist. Manual control by one operator is satisfactory, providing the operator has a knowledge of the type of pour that is desired.

In regards to the second point, it is believed that the author's enthusiasm has allowed him to overstate the point. Transfer of metal from one crucible to another, even though the second crucible is heated, does not give the simplicity or the assurance that can be obtained by pouring directly from the melting-crucible into the mold. Comparable conditions can be obtained, however, by maintaining a thermocouple in the bottom pour crucible. Of course, sufficient time must be allowed for in this case to establish equilibrium conditions of the pyrometer read-

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[3] Naval Research Laboratory, Anacostia Station, Washington, D. C.

ings. Also, direct pouring from crucible to mold allows for faster manipulation of molds, especially when several fluidities are poured from the one crucible of metal.

To what does the author refer by his third statement? Is it meant that more accurate results can be obtained? If so, what is used as the means of comparison? Saeger's results or the author's results with Saeger's method? Presumably it is the latter. My interpretation of the author's data of Table 2 leads me to believe that only one fluidity spiral was poured for each melt. The only manner in which the degree of sensitivity (reproducibility of experimental results) can be obtained is to pour several molds from the same heat of metal at a definite temperature. Such a study would lead to the knowledge of the per cent of deviation from the mean results, which, of course, is what is desired.

Sensitivity has sometimes been spoken of as the length of run in inches per 10 degrees Centigrade. Under such standards the author has increased the sensitivity since Saeger shows (Trans. A.F.A. 1931, p. 513) a run of 0.9 in. per 10°C. and the author obtains 1.4 in. per 10°C. But this is not the per cent deviation of values; it is the result of the method used. Saeger's method calls for a ¼-in. orifice, whereas the orifice used by the author is the cross-section of his spiral or the ½-in. tea pot spout. I am unable to tell which from the drawing.

The curve presented in Fig. 4 is indeed interesting, primarily from the fact that curves of this type are constructed only from one melt of metal and not 26 as has been done by the author. Variations in melting which at times are beyond the melter's control are sufficient to cause considerable variations in the gas content of the melt. Gas content, especially oxygen, seems to be a very important factor in the fluidity of metals.

It is thought that the paper should be studied carefully by the Fluidity Committee of the A.F.A.

O. W. ELLIS[4] *(Written discussion):* Among the most important factors in the development of methods of testing are those having to do with the control of tests. Mr. Halliwell has brought to our attention means which can be adopted for bringing the Saeger-Krynitsky test under more perfect control. That this is desirable is shown by the variability of the results obtained when using the Saeger-Krynitsky test in its original form. The curve which Mr. Halliwell shows in Fig. 4 is some evidence of the value of the improvements which he has introduced.

One cannot help wondering whether core sand might not be used to make the receiving crucible (*A*) shown in Fig. 2. The author does not definitely state what material was used in making this crucible but the inference is that it was made of graphite. In experiments which I have conducted, I have found it possible to use cores for much the same purpose as that which Mr. Halliwell has in mind. Of course, such cores can be used only for single experiments, but they are cheaply made and

would be available in any foundry. I am interested in knowing whether the receiving well (G) is absolutely essential. Has Mr. Halliwell found it necessary to use this well, or could it be dispensed with and be replaced by a down gate of uniform section?

MR. HALLIWELL: In reference to the discussion by Mr. Ellis, although it is not stated in the paper, I believe it was mentioned in the presentation that the crucibles used were of the usual clay-graphite variety. With reference to the use of core sand for the mold, I see no real objection to it. In fact, it was part of our program to make the entire mold of baked core sand, but lack of time has prevented the initiation of this work. Such a mold would obviate the use of the cone shaped down pouring gate. It has been our experience that unless the crucible was well centered in the seat (H) (Fig. 3), part of the metal might be diverted along the overflow gate (I), giving a less effective head of metal and erroneous results. During the centering of the crucible, sand particles are likely to be dislodged into the pouring well, with disastrous effect upon the length of the fluidity spiral. By the use of a baked core sand mold, the crucible could be readily and accurately placed in the seat without loosening any sand.

MEMBER: Mr. Halliwell have you tried pouring any of these with a cold receiving crucible?

MR. HALLIWELL: Yes. It is mentioned in the paper that early experiments were made with a cold crucible and plug. The drop in temperature, however, was so great that the plug frequently froze to the crucible. This forced us to use the hot crucible, heating it to approximately the temperature of the metal. Before preheating, the crucible must be thoroughly cleaned of all adhering slag or metal, as otherwise they will melt, flow down into orifice and reduce its effective size or cause the plug to stick.

MEMBER: Does not the temperature of the receiving crucible make a difference in the length of the spiral?

MR. HALLIWELL: If there is any great difference, perhaps 300-400° F., I should think it would affect the length of the spiral. We generally heated the receiving crucible to about 100° F. in excess of the melting crucible, allowing for cooling in transit before spiral was poured.

CHAIRMAN ROAST: Is this mold, the spiral, in green or dry sand?

MR. HALLIWELL: Green.

MEMBER: Would there be any advantage gained by taking the temperature in the actual pouring cup rather than in the melting crucible? Wouldn't you get closer and more accurate control of your temperatures by that method?

MR. HALLIWELL: Yes, I should say that it would be a distinct ad-

vantage if the temperature of the metal could be measured in the pouring well or even in the receiving crucible. Such a method requires a very rapid adjustment of the thermocouple to the temperature and its register by the instrument. This is not possible with the Marshall tip pyrometer. Possibly a small diameter (No. 18) chromel-alumel bare thermocouple might be used. We tried to insert such a thermocouple in the base of the down gate, hoping to get a reading of the metal there before it entered the spiral. The inertia of the milli-voltmeter needle was so great, that the spiral solidified before the maximum temperature was recorded.

MEMBER: What are the approximate dimensions across the spiral?

MR. HALLIWELL: We used a 14-in. flask and the spiral itself is 8 in. across. The dimensions are all shown in Mr. Saeger's original paper.

MEMBER: What type thermocouple do you use?

MR. HALLIWELL: We used a chromel-alumel thermocouple with a Marshall protecting tip. Our best results were obtained by preheating the tip, thereby reducing the temperature lag.

K. R. VAN HORN[5]: How wide and thick was the spiral?

MR. HALLIWELL: Its cross-section is that of a rounded V, about one quarter of an inch thick and the same across its top face.

MR. VAN HORN: That is a point that may complicate these fluidity tests, that is the ratio of surface area to volume of metal poured. Guillet perhaps has made as many of these fluidity tests as anyone in the world. He has published a large number of papers on the subject, and the point he has stressed in these tests is that there is a minimum surface of metal exposed to volume poured. In other words, the metal in a ¼ in. spiral represents a small amount of surface of the metal exposed to the sand compared to the volume. Perhaps this is not the fluidity relationship that the foundryman is interested in. It would seem that a large amount of surface to volume may be desired to determine the fluidity of the metal under general casting conditions. It will be very interesting for Mr. Halliwell to apply this test to other brass or bronze alloys and perhaps he may find some inconsistent variations between the alloys. Possibly the results might not be exactly the same as a foundryman might expect from his experience in casting the alloys. The discrepancies, if obtained, might be attributable to the amount of surface exposed compared to volume of metal poured in the test. If, instead of the spiral round surface a larger flat surface was selected, there might be different fluidity results than those of the Saeger method of the speaker's.

MR. HALLIWELL: I am not enough of a foundryman to comment on that, but the point that comes to my mind is that the foundryman has to contend with a variety of small cross sections, containing small lugs, fins, or re-entrant angles, that must be accurately reproduced or the

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casting is rejected. Such a condition seems comparable to the point you raise.

MR. HALLIWELL *(Written closure):* It is of interest to learn that Mr. Maack's experience with the fluidity test has been similar to our own, which emphasizes still further the influence of the personal equation.

The vent used by us was placed in the cope at the end of the spiral. While we realize that such a liberal vent is not generally used in practice, it was intentionally made as large as the cross section of the spiral, in order to prevent the formation of a cushion of air or back pressure when the metal was rapidly poured.

I can not see wherein a high zinc alloy, because of its high zinc vapor, would cast better than a lower zinc content in a vented mold. With the size vent that we used the pressure should always be atmospheric. In an unvented mold, the low zinc alloy should have the advantage.

It is a pleasure to receive the discussion and criticisms from Messrs. Kihlgren and Briggs, both of whom have had considerable experience in fluidity testing. I accept Mr. Kihlgren's criticism that if the data is plotted to a slope of 45°, a wider scattering of results is seen, indicating the need of further refinement in the method. The results, however, are a marked improvement over those we obtained with the original method.

Both Mr. Kihlgren and Mr. Briggs advocate the use of a larger volume of metal and the casting of more spirals. In our early experiments, we poured six spirals from a double crucible containing 100-150 lb. of metal. Some of our results are shown in Table I of the paper. The variations in length were so great that we realized the method as we were using it was useless for our purpose. If the rate of pour can be accurately controlled, there is no question that multiplicity of tests from a single crucible with a measured falling temperature is the best method for research work, but such a procedure is wholly impractical to a foundryman interested in production. He needs a method whereby a single or at best two tests will give him the correct information. A foundry melting the same base composition alloy day after day in the same routine should be in a position to make single tests to determine the fitness of his metal for any particular job. He should be able to plot the results of his daily heats to tell immediately whether he had the required fluidity. Many foundries pour from 300 lb. heats or less and to suggest that they pour a number of tests would defeat the acceptance of any test.

Mr. Briggs admits the advantages of bottom pouring, but claims that comparable results can be obtained "with lip pouring if the distance between the mold and crucible remains the same and the crucible is tilted by some controlled means such as an electric hoist." He further states "manual control by one operator is satisfactory providing the operator has a knowledge of the type of pour that is desired." This seems to be the crux of the whole problem and the one that drove us to seek some method that would minimize or eliminate the personal equa-

tion. It is admitted that accurate results may be obtained by an intelligent man properly trained and with sufficient experience, such as Mr. Briggs has available at the Bureau of Standards, but what are foundrymen to do who have to contend with the vagaries of labor and who have no suitable electric hoist.

Mr. Briggs further contrasts the two methods in their "simplicity" and "assurance" of operation. Admitted that the original method is more simple to operate in theory at least, of what use it is to the foundrymen if he can not reproduce his results?

No doubt some arrangement could be used whereby a thermocouple could be placed in the bottom pouring crucible and so measure the temperature of the metal as it entered the spiral. Our experience, as stated in the paper, has been the inability of the instrument needle to adjust itself to the maximum temperature before the mold has been cast.

In questioning the sensitivity of our method, Mr. Briggs is correct in assuming that our standard of comparison is with our results using Saeger's method rather than with Saeger's results. If our results obtained with Saeger's method show wide variation, either the method is at fault or we have been unable to use it to the same success as Saeger. Others who have used the method have the same criticism, namely, the personal equation.

If sensitivity is measured by the length of the spiral per 10°C. then any modification of design or change in technique that results in an increase in the spiral is advantageous. If in our case it is the result of a ½ in. opening instead of ¼ in. should not the size of the opening leading to the spiral be changed? Perhaps the size or cross-sectional area and shape of the spiral should be changed to conform more to some conditions experienced by the foundrymen, as suggested by Van Horn in his oral discussion of this paper. If such changes in physical shape give us a better picture of the value of fluidity, then they should be made.

If this paper has done nothing else but initiate a reconsideration of the original Saeger method, which will result in its improvement, the author will feel well repaid for his efforts, and it is his sincere hope that the Fluidity Committee of the A.F.A. will study carefully its merits and demerits.

Silica Refractories for Cupola Service*

By W. J. Rees**, M.Sc.Tech., F.I.C., Sheffield, England

1. Cupolas are, in general, lined in one of the three following ways:

(a) With firebrick

(b) With bricks or blocks cut to shape from a natural siliceous rock or "firestone"

(c) With a plastic or semi-plastic material rammed round a former.

2. The firebricks used are in the *high heat duty* class with P.C.E. values between 31 and 33. Under special conditions, it may be desirable to use firebrick of the *super-duty* class, with a P.C.E. value above 33, in the melting zone. Special brick, made from materials other than fireclay, such as chrome, calcined or fused alumina etc., have been used experimentally but their service has not been commensurate with their much higher cost.

Silica Brick

3. Silica brick generally have been ruled out because of their high spalling tendency, but it is quite possible that silica brick with a clay bond instead of the normal lime bond would give good service, as with the change in bond the susceptibility to spalling is decidedly reduced. Silica brick made from quartzites containing from 1 to 3 per cent of titanium oxide finely dispersed, when properly graded, have a much higher spalling resistance than brick made from normal quartzites, and would be likely to behave satisfactorily in the cupola lining. These brick should be burned to a powder density of 2.36 to 2.40, so that in service there will be a little permanent expansion which will close the joints and prevent slag ingress.

Silica Firestone Block

4. Silica firestone blocks or bricks are not much used in British foundry practice, but their use in American foundries is being followed with interest. The "firestones," which are avail-

able in Great Britain, are fine-grained sandstones of the following chemical composition:

	PER CENT
Silica	92 to 94
Alumina	2 to 5
Iron Oxide	1 to 2
Lime	0.2 to 1.0
Magnesia	0.1 to 0.4
Alkalies	0.5 to 2.0
Ignition Loss	0.5 to 1.0

5. The photomicrograph (A of Fig. 3) shows the structure of a typical firestone. The chemical composition quoted above is not very different from the published analysis of the Ohio bedded silica firestone which is being used successfully in some American foundries.

PLASTIC OR SEMI-PLASTIC MATERIALS

6. The plastic or semi-plastic materials usually are highly siliceous, though a few ironfounders use mixtures containing a high proportion of fireclay grog with a bond of plastic fireclay. The siliceous mixtures used for the production of rammed linings and for patching are of two types:

(a) Natural silica "clays" which consist of sand grains of small size uniformly coated with a highly plastic siliceous clay.

(b) Silica rocks of "ganister" type which are crushed and ground with a plastic fireclay.

7. The plastic or semi-plastic siliceous materials fall in silica-alumina ratios between points A and B on the portion of the alumina-silica fusion point curve shown in Fig. 1. It is to be noted that in mixtures containing between 7 and 10 per cent alumina (that is 20 to 30 per cent of an average 30 per cent alumina fireclay), there is a rapid fall in the fusion point, and that the presence of fluxing impurities in the silica rock and fireclay move the lowest fusion point in the direction of a lower silica-alumina ratio.

8. Occasionally, synthetic mixtures of similar composition to the naturally bonded mixtures of type (a) are made by thoroughly mixing a high silica sand with a fine-ground plastic refractory

clay. It is exceedingly difficult, however, to obtain by mechanical mixing such uniform dispersion of the clay over the sand grains as is found with the naturally bonded siliceous materials. With wet mixing, the dispersion of the clay is assisted by deflocculating it by the addition of about 0.5 per cent of a mixture of sodium silicate and sodium carbonate, or by adding the clay in the form of a defloculated slip.

ROTARY FURNACE LININGS

9. In the lining of rotary melting furnaces, mixtures of both the natural and synthetic types are used, but where the melting temperature is higher than normal, it is advantageous to replace the clay bond with lime, producing a mixture which has a chemical
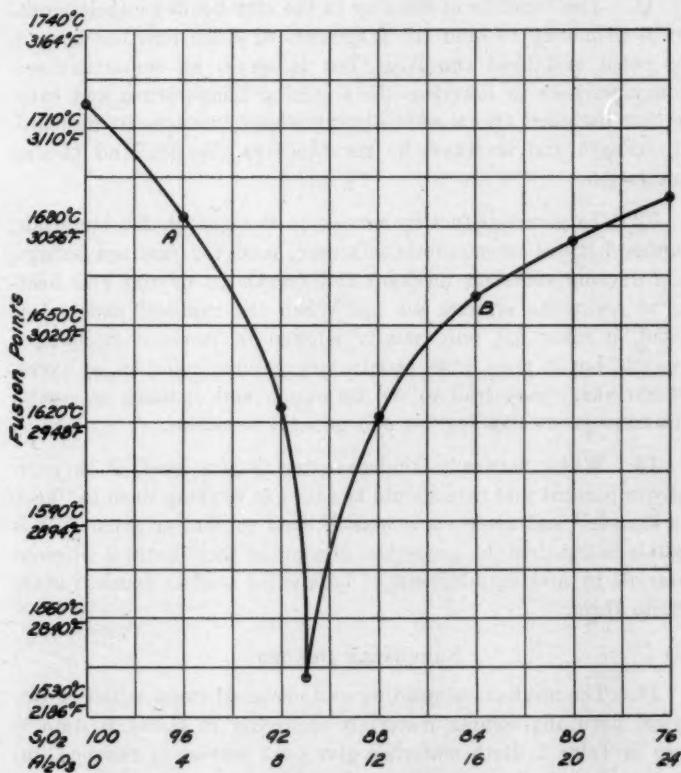
FIG. 1—FUSION POINT CURVE OF ALUMINA-SILICA.

composition and grading comparable with that of a good silica brick. The plasticity and "workability" of the lime-bonded mixtures may be increased by the addition of an organic bond such as dextrin or one of the cheaper by-products bonds of the "sulphite-lye" type.

10. It would seem to be feasible to line a rotary furnace with unburned silica bricks compacted under hydraulic-pressure, which would avoid the difficulties of uniform ramming and lamination with granular lining mixtures, and would give a much denser lining. These unburned bricks could, if desired, be molded with a tongue and groove to facilitate building together.

CUPOLA LININGS

11. The function of the clay in the clay-bonded cupola ganisters is primarily to bind the fragments of silica together both in the green and fired condition, but it serves an important secondary purpose in lowering the sintering temperature and causing that incipient fusion of the lining which promotes its mechanical strength and increases its resistance to abrasion and to slag penetration.

12. To give satisfactory service in the cupola, the materials, whether natural or synthetic mixtures, used for rammed linings, must develop sufficient mechanical strength on drying and heating to resist the stresses set up. When the rammed material is heated, it must not only vitrify enough to produce mechanical strength, but it must have satisfactory volume stability, as excessive shrinkage may lead to the formation and opening of cracks, with consequent slag ingress and general weakness.

13. When naturally-bonded materials are used, it is especially important that care should be taken in working them to "keep to a sample" and avoid variations in bond content or texture. It is possible to improve the properties of some of these natural siliceous materials by making additions of high-silica sand or crushed silica rock to them.

NATURALLY BONDED

14. The mechanical grading and chemical composition of two typical naturally-bonded materials occurring in Great Britain is given in Table 1. Both materials give good service in rammed linings, but it is important to keep the moisture content as low as is

Table 1

MECHANICAL GRADING AND CHEMICAL COMPOSITION OF ENGLISH NATURALLY-BONDED MATERIALS

B.S.I. Sieves	Mechanical Grading	
	No. 1	No. 2
	%	%
+ 5 mesh	0.5	3.5
+ 16 mesh	1.5	2.8
+ 30 mesh	1.5	5.5
+ 60 mesh	2.5	15.0
+ 85 mesh	8.0	14.0
+100	27.0	10.5
+150	21.5	10.0
+200	8.0	7.5
—200	4.0	2.5
Clay Grade	25.5	28.7
	100.0	100.0

CHEMICAL ANALYSIS OF AIR-DRIED MATERIALS

	No. 1	No. 2	Clay grade from No. 1.
	%	%	%
Silica	76.59	84.02	45.61
Alumina	14.70	9.85	35.55
Iron Oxide	1.90	1.15	2.15
Titanium Oxide	0.35	0.32	0.85
Lime	0.74	0.35	0.71
Magnesia	0.18	0.22	0.10
Potassium Oxide	0.62	0.58	1.24
Sodium Oxide	0.22	0.24	0.44
Loss on ignition	4.55	3.15	13.25
	99.85	99.88	99.90

Table 2

EFFECT OF WATER ON DRYING SHRINKAGE

Water added to Air dry material.	Wet to dry Shrinkage (linear)	
	No. 1	No. 2
%	%	%
5.0	0.4	0.5
6.0	0.7	1.0
7.5	1.0	1.2
10.0	1.9	2.0
12.5	2.5	2.5
15.0	3.5	3.0

compatible with adequate workability in ramming to avoid undue drying shrinkage, especially with materials of fine grain size such as No. 1 of Table 1. The effect of water content on drying shrinkage for the above two materials is shown by the figures of Table 2, obtained with representative samples.

15. Both these materials show a permanent expansion of approximately 1.0 per cent when a dry test piece is fired at 1450°C. (2642°F.); this compensates for the drying shrinkage if the added water does not exceed 7.5 per cent. In the forming of rammed linings in the cupola, great care must be taken to avoid laminations as these are likely to lead to spalling of the lining in service.

16. The deleterious effect of laminations in a rammed lining is shown in Fig. 2 which is from the lining of a rotary melting

FIG. 2—DELETERIOUS EFFECTS OF LAMINATIONS IN A RAMMED LINING—EFFECTS LIKELY LEADING TO SPALLING OF THE LINING IN SERVICE.

furnace in which trouble was experienced through spalling. The
method of ramming used must be such as will ensure complete
keying of the layers of material.

17. In Great Britain, while some foundries use materials
such as those detailed above for patching and general maintenance,
the majority use materials which are much coarser in fragment
size and which are described as cupola ganister. These materials
are made by mixing a crushed silica rock containing not less than
92 per cent silica with from 10 to 30 per cent of a refractory plastic
fireclay; the proportion of clay used depending on the character
of the silica rock, the chemical composition and plasticity of the
clay, and the required plasticity of the mixture.

18. The structure of Sheffield ganister rock, which is the
silica rock from the use of which the term "cupola ganister"
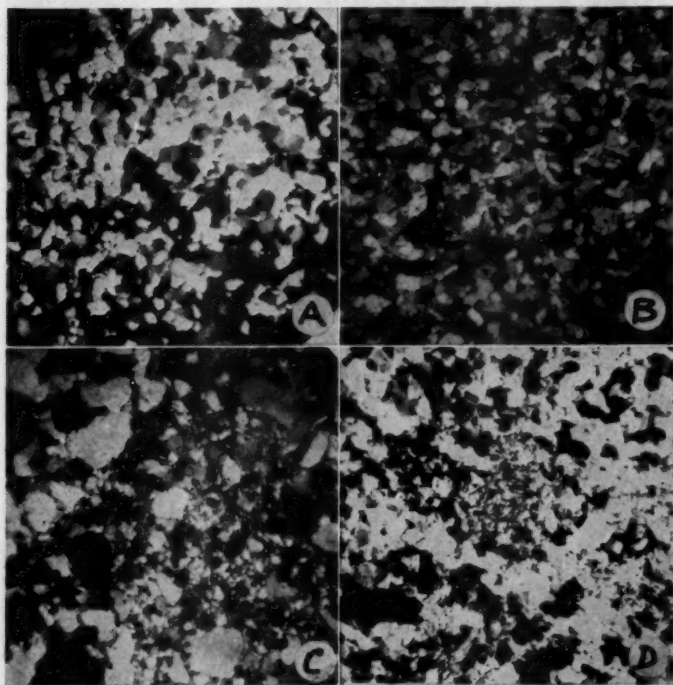
FIG. 3—PHOTOMICROGRAPHS OF SILICA REFRACTORIES—CROSSED NICOLS X20 (A) TYPICAL
FIRESTONE, (B) GANISTER, SHEFFIELD, (C) QUARTZITE, DURHAM, (D) QUARTZITE, ANGLESEY.

originated, is shown in *B* of Fig. 3. This rock contains 97-98.5 per
cent silica. Photomicrographs of other silica rocks used in the
production of cupola ganister are shown in Figs. 3 and 4. It is
desirable that the physical character of the silica rock should be
such that when it is crushed it will give angular fragments.
Rounded fragments tend to slip over each other, and so lead to
weakness, while angular fragments tend to interlock, and so pro-
mote strength.

19.   The plastic fireclays which are used in admixture with
the crushed rock have an alumina content between 25 and 35 per
cent and a P.C.E. between cones 26 and 33. In the best practice,
the silica rock used is of the Sheffield ganister type with above
97 per cent silica and the plastic fireclay of the aluminous type
(above 30 per cent alumina) with a P.C.E. above cone 30.
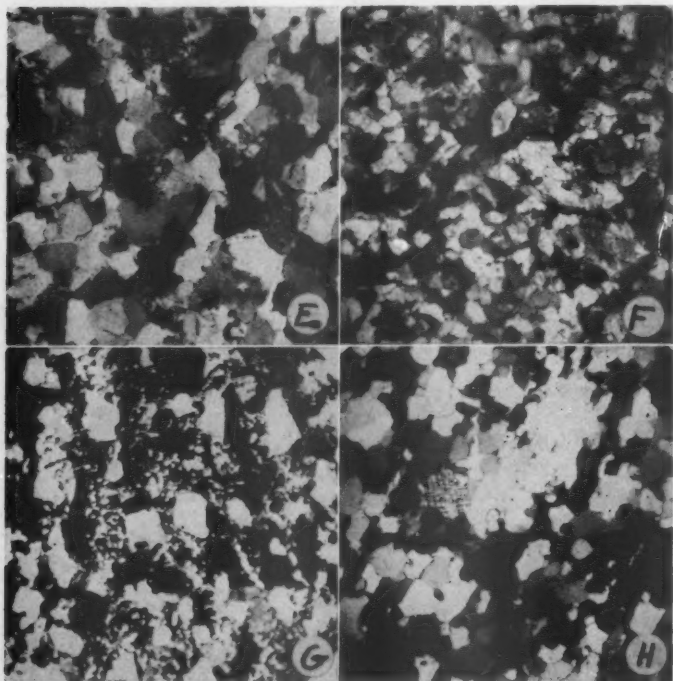
FIG. 4—PHOTOMICROGRAPHS OF SILICA REFRACTORIES—CROSSED NICOLS X20 (E) QUARTZITE,
SHROPSHIRE, (F) SILICA ROCK, ROYDS EDGE, (G) QUARTZITE, SOUTH WALES, (H) QUARTZITE,
NORTH WALES.

20. It is important that the matrix of the cupola ganister (*i.e.*, the finer fragments of silica plus the bond clay) should have a chemical composition which is not too near the eutectic (Fig. 1) as this will lead to rapid fusion and possible dripping or running out of the matrix and a "honeycomb" type of corrosion or wear. The use of a refractory clay of a "sticky" nature is advantageous as it will give better bonding in the unfired material, better workability with the minimum proportion of clay, and will allow the use of low water content.

21. All patching materials must have such characteristics as will enable them to "stay put," or develop adhesion, when heated. After the moist patching material is forced into position, the worn place made good, and the cupola heated, the following cycle takes place: (1) The water in the material is driven off; (2) the combined water in the bond clay is driven off; and (3) sufficient vitrification is induced to make the patch adhere to the rest of the lining. It will be seen that a severe condition is set up, and the material used must be one which will withstand the inevitable strains.

22. Patching materials must be used always with only just enough water to give the necessary workability, as over-wetting will not only necessitate slower drying-out, but will also tend to develop cracks and give a porous patch. Before patching, it is of course essential that the slag should be removed as completely as possible from the worn place in the cupola.

23. The actual standardization of the properties of cupola ganister is made difficult by the great variation in cupola conditions. It is by no means unusual for a ganister mixture which gives good service in one cupola to be totally unsatisfactory in another working under conditions which do not appear to be very dissimilar. It seems evident, therefore, that the method of using the ganister has a definite influence on its durability.

MECHANICAL GRADING AND COMPOSITION

24. The mechanical grading of a cupola ganister should be such that when it is dried off, and burned, it gives a closely packed lining with a low porosity and permeability. This close packing will also tend to quick and uniform sintering. To get this close packing, the proportion of grains or fragments of middle size should be low, and the proportions of coarse and fine fragments and grains roughly equal.

25. In most cupola ganister mills, the final grading produced is a fortuitous one, but it is possible by control of the grinding time to produce a ganister mixture which has a reasonably satisfactory grading. It would be preferable; and quite possibly advantageous, to first crush and screen the silica rock and then to mix the desirable proportions of the various screen grades with the required proportion of fine-ground refractory clay, and finally wet in a mixer which is so designed as to give thorough mixing without any further grinding. In this way, very complete control of grading, clay proportion, and water content would be possible.

26. The mechanical grading and chemical composition of cupola ganisters as delivered to iron foundries vary within the limits given in Table 3.

Table 3

MECHANICAL GRADING AND CHEMICAL COMPOSITION
OF CUPOLA GANISTERS

Mechanical Grading — B.S.I. Sieves

+	5 mesh	3 to 20 per cent
+	30 mesh	30 to 40 per cent
+	60 mesh	4 to 12 per cent
+	100 mesh	5 to 15 per cent
—	100 mesh	30 to 37 per cent

Chemical Composition of Air-Dried Materials

Silica	78 to 90 per cent
Alumina	5 to 15 per cent
Iron Oxide	1 to 3 per cent
Titanium Oxide	0.1 to 0.5 per cent
Lime	0.1 to 0.8 per cent
Magnesia	0.1 to 0.5 per cent
Alkalies	0.4 to 2.0 per cent
Loss on ignition	3.0 to 8.0 per cent

27. *Moisture.* The moisture (*i.e.*, added water) in the materials as used varies between 10 and 20 per cent, and the drying shrinkage is from nil to 2.0 per cent.

28. *Shrinkage and Expansion.* The volume change of dry test pieces fired at 1500°C.-1550°C. (2732-2822°F.) varies from a small shrinkage to an expansion of 7 or 8 per cent. A small expansion is preferable to a large expansion or a shrinkage, but where the melting conditions are severe it is essential to have a high silica content in the ganister mixture (bearing in mind, of

course, the necessity of keeping clear of the eutectic composition) despite the expansion it will show in service.

In addition to drying tests, grading tests, and chemical analyses, there are several useful laboratory tests for cupola ganister.

ADDITIONAL TESTS FOR CUPOLA GANISTER

29. *Refractoriness.* With fine-grained materials, cones may be made and their P.C.E. determined in the usual manner against standard cones. It is impossible to make cones from the coarse-grained materials, but a useful alternative is to make test briquettes 4 x $1\frac{1}{2}$ x 1 in. and heat them at 1550°C. (2822°F.) for 4 hours. With good cupola ganisters, there will not be much distortion of the briquette after this heat treatment, but it will have become vitrified and will have developed considerable mechanical strength.

30. *Slag Resistance.* A test briquette 3 x 3 x 2 in. is made with a cylindrical depression 1-in. diam. and 1-in. deep in the middle of one 3 x 3 in. face. This is dried, the depression filled with ground cupola slag, and heated for 2 hours at 1500°C. (2732°F.); after cooling, the extent of slag penetration and corrosion can be examined by breaking the briquette through across the middle of the depression.

An alternative method for evaluating slag resistance is to make a solid briquette, say, 2 x 2 x 2 in. and after drying, place on top of it a cylindrical pellet made from about 20 to 30 grams of ground slag (moistened with a dextrin solution), and then heat for 2 hours at 1500°C. (2732°F.). The slag attack is observed after cooling the test briquette. This second method is particularly useful for observing the effect of the porosity and permeability of the ganister on slag penetration, as well as its resistance to chemical corrosion. Both methods of test may be amplified by mixing fine-ground iron oxide (iron scale for example) with the cupola slag.

31. *Refractoriness-under-load.* As the actual pressure due to the weight of the charge in the cupola may have some influence on the effective refractoriness of the cupola lining or patching, it is useful to make tests on dried test-pieces under a load of 25 lb. per sq. inch. The method of conducting this test should be as detailed on p.p. 48-52 of the Manual of A.S.T.M. Standards on Refractory Materials.

32. *Simulative Tests.* With the co-operation of E. S. Renshaw, Ford Motor Co. Ltd., Dagenham, Essex, the Refractories Committee of the Institute of British Foundrymen have been able to make simulative tests in a cupola fitted with a continuous tapping device. A cupola of this type gives suitable conditions, as after the first tap is made, the cupola well is soon filled with slag only, the metal running immediately into a receiver on reaching the cupola bottom.

33. Test panels 6-in. square by 1-in. thick, backed with thin firebrick tile are fitted into the cupola well exactly opposite to each other, and are in contact with molten slag for the duration of the run, which is about 7 hours. The distance between the two

Table 4

DISTANCE BETWEEN PANELS

	Before Run Inch	After Run Inch	Erosion per Panel Inch
Standard Ganister	26.$\frac{7}{8}$	27.$\frac{3}{8}$	0.25
Test Material (a)	25.$\frac{11}{16}$	26.$\frac{7}{8}$	0.78
Standard Ganister	30.$\frac{5}{8}$	31.$\frac{9}{16}$	0.47
Test Material (b)	31.$\frac{1}{8}$	31.$\frac{5}{8}$	0.25

Ratio of Erosion of Test Material (a) to that of the Standard Ganister, 3 : 1

Ratio of Erosion of Test Material (b) to that of the Standard Ganister, 0.5 : 1

panels before and after the run, divided by two, may be taken as the measurement of the erosion. A material of known satisfactory behavior in the cupola may be used for standard panels alongside the other materials under test.

34. Table 4 gives the particulars of two materials tested in this way against a standard ganister. Where a cupola of this type is available, this method of test is to be recommended, as it more nearly simulates actual working conditions than any laboratory test for erosion or corrosion.

What's New in Crucible Furnaces

By RICHARD H. STONE*, SWISSVALE, PA.

1. At the 1934 Convention of this Association, a paper[1] was presented before the Refractories Section on this same subject of crucible furnaces and the present paper attempts to bring the subject up-to-date.

SEALED FURNACES

2. In an article[2] in METAL INDUSTRY, M. G. Corson advocates the development of the crucible furnace to give "sealed chamber melting." That is to say, the metal in the crucible should, at all times, be protected from contamination through contact with flame or flue gas. Such a furnace design must sacrifice some thermal efficiency as it is impossible to utilize the radiated heat ordinarily striking the metal from the furnace cover. The increased fuel consumption is offset, however, by the reduction in metal loss.

3. Such furnaces (Fig. 1) are now being used in a number of installations. Oil consumption averages 2½ gallons per 100 pounds of metal melted as compared with 1½ to 2 gallons per 100 pounds of similar metal. So far, the principal application has been to melt yellow brass borings to reduce zinc losses. Actual operating figures are less than one per cent zinc loss compared with 3 to 3½ per cent in the open crucible type. It is hoped that the application of this furnace will be extended to melting any non-ferrous alloys to which contact with the products of combustion is injurious.

INSULATION

4. Insulation of crucible furnaces is receiving more and more attention. It is sufficiently accurate to say that all the commercial furnaces, particularly tilting types, are now supplied with an insulated lining. The more expensive the fuel, of course, the greater the desire to conserve the heat. With a view to making the admittedly desirable features of "city" gas economically applicable,

[1] Stone, Richard H., *Modern Equipment Used in Crucible Melting*, Trans. A.F.A. vol. 42 (1934), pp. 265-277.

[2] Corson, M. G., *Second Thoughts About Foundry Furnaces*, Metal Industry, April, 1936.

* Vesuvius Crucible Company.

special attention has been given to efficient combustion and insulation of furnaces using this fuel.

5. Principal features of such furnaces are: (1) Use of the proportional mixer insuring uniformity of gas and air ratio; (2) A relatively thin lining of low thermal capacity and high emissivity resulting in maximum use of radiant heat; (3) Insulation by means of a layer of insulating brick or bulk material between the lining and the furnace shell (See Figs. 2 and 3).

6. The lining materials which have been successfully used for these furnaces included silicon carbide, mullite and sillimanite. The insulating material is of a refractory type capable of withstanding temperature as high as 2800°F. On tilting furnaces where there is no tong abrasion, the insulating refractory can be used alone with only the protection of an air-set wash which is gradually built up.

7. An actual test under the above conditions using city gas of 540 B.T.U. per cu. ft., melting 80 per cent copper, 20 per cent zinc, showed an average gas consumption of 2.75 cu. ft. per lb. or
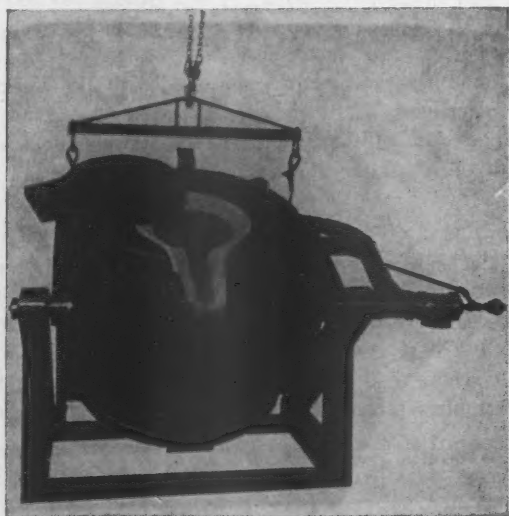
Fig. 1. SEALED CRUCIBLE TYPE FURNACE. FURNACE ATMOSPHERE CANNOT COME IN CONTACT WITH METAL IN CRUCIBLE. ECONOMICAL FOR MELTING YELLOW BRASS BORINGS AND TURNINGS. TILTING AXIS THROUGH END OF LIP FOR POURING INTO INGOT MOLDS ON MOVING TRAIN. (ILLUSTRATION, COURTESY FISHER FURNACE CO.)

5500 cu. ft. per ton of metal melted. The total B.T.U. imput per ton, therefore, was 2,970,000, showing an efficiency of 15 per cent when compared with the theoretical 450,000 B.T.U.'s required to melt 2000 lb. of this metal[3]. Without insulation gas, consumption averaged 5 cu. ft. per lb. of metal melted.

PROPORTIONAL MIXER

8. The proportional mixer mentioned above should always be installed wherever gas is used. It is virtually impossible to main-

[3] Pritchard, Oliver, *Making Brass Melting Efficient*, Industrial Gas, May, 1935.
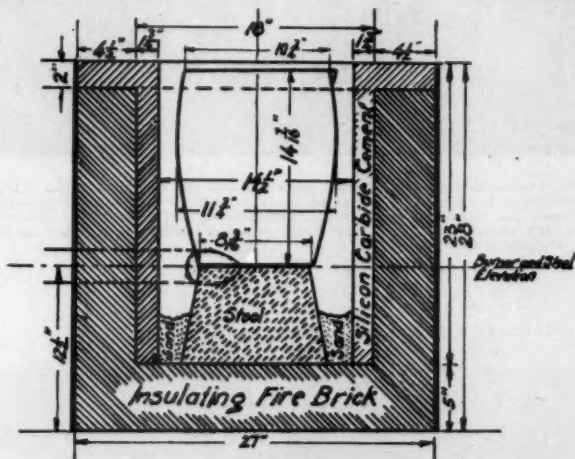
FIG. 3. CROSS SECTION OF CRUCIBLE FURNACE, WITH RELATIVELY THIN LINING OF SUPER REFRACTORY TYPE, BACKED WITH REFRACTORY INSULATING BRICK. (ILLUSTRATION AND DATA FROM "MAKING BRASS MELTING EFFICIENT" BY OLIVER PRITCHARD, INDUSTRIAL GAS, MAY, 1935).

	Feb. 4, 1935	Feb. 5, 1935
Dates of tests		
Number of heats	7	6
Weight of metal melted (Pounds)	1,355.5	1,161
Average zinc losses, as shown by difference in weight of loaded crucible, in and out of furnace (Per cent)	0.85	not taken
Average time required to heat metal to pouring temperature (2200°F.) per heat (Minutes)	44	42
Average time required to heat metal to pouring temperature (2200°F.) per heat—without initial heat (Minutes)	41	36
Average unit gas consumption, per pound of metal melted (Cubic feet)	2.9	2.75
Average unit gas consumption, per pound of metal, without initial heat (Cubic feet)	2.67	2.37
Metal melted	20 per cent Cu, 20 per cent Zn.	
Pouring temperature	2200°F.	
Size of crucible	No. 60, holding 180 to 200 lb. brass	
Calorific value of gas	540 B.T.U. per cu. ft.	
Temperature of furnace face, end of day	2500 to 2600°F.	
Shell temperature, end of day	375°F. maximum	
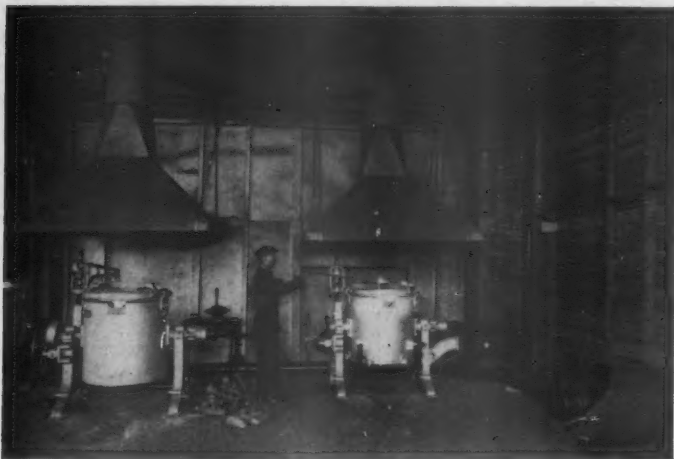
FIG. 3. INSULATED TILTING FURNACES, SIZES 150 AND 225 CRUCIBLES USING "CITY" GAS. LARGER FURNACE HAS SINCE BEEN CONVERTED TO FOUR BURNERS AS SHOWN ON SMALLER ONE. BEARING METAL IS MELTED AT OPERATING TEMPERATURE 2300°F. AND GAS CONSUMPTION 3.4 CU. FT. PER LB. OF METAL. (ILLUSTRATION, COURTESY BROOKLYN UNION GAS CO. AND CAMPBELL HAUSFIELD CO.)
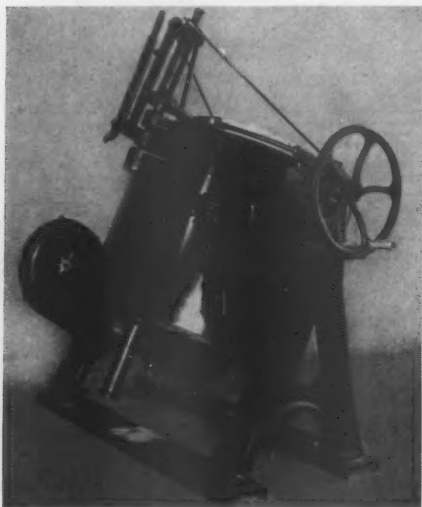
FIG. 4. TILTING TYPE CRUCIBLE FURNACE USING COKE FUEL WITH FORCED DRAFT. (ILLUSTRATION, COURTESY CRUCIBLE FURNACE CO.)
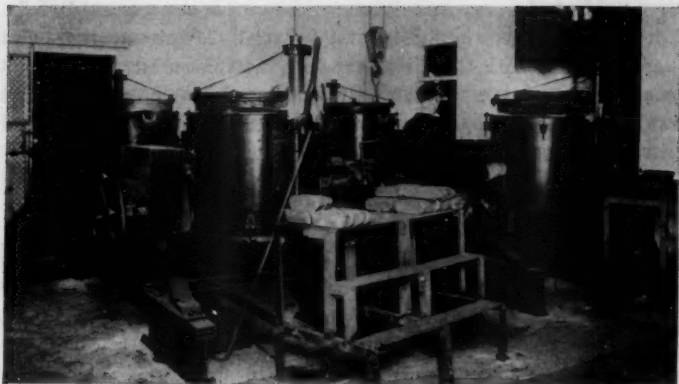
FIG. 5. BATTERY OF FOUR TILTING FURNACES, OIL FUEL, SIZE NO. 225 LONG LIP CRUCIBLE, MELTING HIGH GRADE BRONZE FOR SLEEVES AND OTHER PARTS FOR HEAVY MACHINERY. CASTINGS WEIGHT 800 TO 2250 POUNDS. (ILLUSTRATION, COURTESY NORDBERG MANUFACTURING CO.)

FIG. 6. SIX STATIONARY FURNACES, SIZE NO. 80 CRUCIBLES. SILICON CARBIDE LININGS WITH INSULATION. ELECTRIC FUEL OIL HEATERS ON PANEL IN REAR, THERMOSTATICALLY CONTROLLED, PERMIT USE OF NO. 6 OIL INSTEAD OF NO. 4 AT A SAVING OF 2 CENTS PER GALLON. (ILLUSTRATION, COURTESY FISHER FURNACE CO.)

tain a uniform furnace atmosphere without this equipment. Briefly, it reduces the gas supply to atmospheric pressure at the point where it mixes with the air in the line. At this point, the air line is narrowed and the resulting air velocity draws in the gas. The quantity of gas used then depends on the velocity of the air and the ratio of gas to air is constant.

DIFFUSION COMBUSTION

9.   The use of the diffusion type of combustion in crucible furnaces has seemed desirable because of the efficiency of the luminous flame and the longer crucible life that might be expected with the non-oxidizing furnace atmosphere. However, the luminous type flame requires a long path and this is very inconvenient to provide in the crucible furnace. It has been tried by placing several crucibles in one furnace but with this arrangement it is difficult to handle each crucible individually.

10.   A recently developed pot furnace[4] employs this principal ingeniously with excellent efficiency. The lining is composed of special curved tiles layed end to end forming a continuously ascending spiral chamber surrounding the pot. This arrangement gives the requisite long path to utilize the luminous type of flame. Although the flame temperature is several hundred degrees lower than the ordinary type, more rapid heating results from the increased radiation.

COKE FIRED FURNACES

11.   Interest has been revived recently in the coke fired crucible furnace using forced draft through the development of a new design (Fig. 4). The blower and motor are mounted on the furnace and move with it in the tilting type. The air is discharged from under the crucible rest block, thus insuring more even distribution. Practically silent operation may possibly be considered as one of the desirable features of this, in common with all coke fired crucible furnaces.

12.   Fuel cost, in one installation melting 85-5-5-5 red brass and using coke, is $2.03 per ton. This is approximately only two-thirds of the fuel consumption in pit furnace practice in the same shop.

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[4] Industrial Gas, Nov. 1937, p. 24.

13.   Another feature of this forced draft, coke furnace is that the various sized crucibles all have a relatively small diameter in proportion to the height. While not new, as crucibles with these proportions have long been made, it is noteworthy and illustrates the tendency in modern crucible furnace design to draw on all the known principles and combine them to attempt to produce the ultimate efficiency.

TALL NARROW CRUCIBLES

14.   The advantage of the small diameter crucible is, obviously, that the heat supplied has a shorter path to travel to the metal at the vertical axis of the crucible. Since much of the heat, particularly during the melt-down period, is supplied by radiation, the efficiency increases not simply as the reduction of the distance, but as the square of the distance. The application of this principal is particularly important in the case of aluminum melting crucibles. Contrary to expectations, aluminum, although it melts at 1420°F., requires as much fuel to melt as the rich copper alloys. This is because of its high specific heat which is 0.22 as contrasted with 0.09 for brass. The extra fuel, therefore, is consumed during the melt-down period, and the use of as narrow a crucible as practicable, would, therefore, be desirable to increase the radiant heat delivered to the metal.

15.   To get the same capacity with a narrow crucible as with a wide one means an increase in the height. This is really an added advantage as it, in turn, requires a deeper furnace, giving opportunity for longer flame travel; the problem in crucible furnaces ordinarily is one. of providing sufficient flame length to develop and utilize the available heat.

SUMMARY

16.   In conclusion, the trend of development of crucible furnaces may be summarized as follows: The tilting furnaces have been greatly improved in ease of operation, efficiency and applicability. For example: for rolling mill practice, they are now made to tilt about the end of the lip as an axis so metal may be poured directly into ingot molds in the train as they successively come up to the pouring station. (See Fig. 1).

17.   The tilting furnaces are commonly used in larger sizes commencing with 375 lb. and up to 1200 to 1500 lb. capacity. (See

Fig. 5). Naturally, they are required where larger castings are poured. The lift-out type of furnaces have enjoyed a considerable increase in number of installations where many small, similar castings are made, such as in the plumbing goods shops. The common typical installation consists of a battery of a half dozen to twenty or more No. 60, 70 or 80 size furnaces using oil or gas fuel. (See Fig. 6). The working conditions are quite tolerable, the operation of burners is largely under automatic control and the use of the same crucible for both melting and pouring results in rapid, efficient operation at minimum cost. For greater flexibility, a combination of tilting and lift-out furnaces is installed. (See Fig. 7).

FIG. 7. COMBINATION OF TILTING AND STATIONARY FURNACES USING OIL FUEL. SUCH AN INSTALLATION FULLY DEVELOPS THE FLEXIBILITY OBTAINABLE WITH CRUCIBLE MELTING EQUIPMENT. (ILLUSTRATION, COURTESY MONARCH ENGINEERING & MANUFACTURING CO.) THE FOLLOWING FIGURES MAY BE TAKEN AS TYPICAL OF OIL FIRED, TILTING FURNACES:

Crucible size ....No. 175 long lip.
Number of heats per crucible...116.
Average time per melt...1.17 hrs.
Oil consumption ....1.969 gal. per 100 lb.

The Role of Silicon in Non-Ferrous Castings

By H. W. Gillett,* Columbus, Ohio

Abstract

Silicon plays an important role in non-ferrous castings but, because its effects were not clearly understood, it was looked upon as a poisonous element by non-ferrous foundrymen for some time. Much of the mystery surrounding its effects has been dispelled by research and and it now is used extensively in many non-ferrous alloys. In discussing his subject, the author begins with the uses of silicon in ferrous metals, its properties and sources. He then discusses the use of silicon in magnesium-base, aluminum-base, copper-base and nickel-base alloys. Under the aluminum-base alloys, he describes the action of silicon in general and specifically gives the properties of low silicon-aluminum alloys, aluminum die casting alloys, a special piston alloy, containing in addition to silicon, small percentages of magnesium, copper and nickel. Following a discussion of the more complex aluminum sand casting alloys, the author discusses precipitation hardening at some length. Copper alloys are next on the list for consideration. Under this section, in addition to the plain copper-silicon alloys, the effects of tin, zinc, manganese and iron on the copper silicon alloys are explained. Following a resumé on the foundry behavior of the copper-silicon alloy series, the author tells of the function of nickel in copper and copper-silicon alloys. He then discusses the role of silicon in cupronickel, monel and other nickel-base alloys. The paper is replete with information as to the physical and mechanical properties of many non-ferrous alloys involving silicon. The author, in many instances, gives trade names, contents, properties, foundry behavior, heat treatment of a large number of alloys. Not the least valuable portion of the paper, is the bibliography at the end, containing nearly 50 references to the various alloys discussed.

* Battelle Memorial Institute.

1.   The chemist is not quite sure whether silicon is a true metal or just a metalloid. Some authorities accord it a place with the metals, but with reservations. Hume-Rothery says it acts like a metal above 2000° F., like a non-metal below that temperature. Abnormalities in coefficient of resistance and thermo-electric power, show it to act differently from most metals but on the whole, it has a metallic character, so it is usually termed a "border-line" metal rather than a metalloid. One might expect such a material to hop across the border now and then — that is — to have a somewhat unpredictable behavior. It does just that, and its vagaries can be either fascinating or irritating, according to one's point of view.

FERROUS USES

2.   The ferrous metallurgist relies upon adjustment of silicon to bring about graphitization in cast iron, or, by limiting its quantity, to give white iron. Cast iron is considered to be highly sensitive to the time of addition of silicon so that an iron with all the silicon added with the charge and one to which part of the silicon is added in the ladle, seem quite different entities, though the analyses made by the usual methods may be identical.

3.   Retaining too large a proportion of the total silicon for the final addition may destroy the good effect of the dosing. Different lots of ferrosilicon may act somewhat differently, some of these differences being traceable to impurity content. A better understanding of the apparent vagaries of silicon in cast iron (which really are vagaries only because we do not fully understand the variables involved) seems to be the path to better and more uniform and controllable cast iron.

4.   The steel foundryman, too, is dependent on silicon. If he leaves it out, he gets a rimming steel, of no use as a casting. With proper amounts, properly added, he gets sound castings. But if he introduces part of the silicon by reduction in the furnace from an acid lining, he may meet the so-called "over reduced" condition where the steel is again porous and addition of more silicon makes a bad matter worse.

5.   The action of silicon in cast steel usually is referred to as "deoxidation," but matters of gas solubility and of holding gas in forced solution while the steel freezes may be of as much importance as the actual deoxidation.

6.   The deoxidation products into which SiO_2 enters, and

which remain suspended or entrapped in the steel, also may exert a profound influence. Steel men are still struggling with their vagaries.

7. In such uses as have been mentioned, silicon acts more like a chemical reagent than it does like an alloying metal, though some high silicon steel castings, especially those with other alloying elements, are very promising. Some of the new high yield strength, low alloy steels utilize up to 1 per cent or more silicon as a strengthening alloy.

8. As we go to higher silicon wrought steel, the electrical resistivity and the magnetic properties conferred by considerable amounts make "silicon sheet" of great value in the electrical industry, but in such steels one has to contend with a peculiar type of brittleness that does not follow ordinary rules.

9. On raising the silicon to very high amounts, as in Dur-iron, we get valuable corrosion resistance, but extreme brittleness.

PROPERTIES OF SILICON

10. Silicon itself is an unusual sort of material. It is very light, sp. gr. 2.4, i. e., between aluminum and magnesium, its melting point is nearly 2600° F., and its hardness is way up in the scale, readily scratching glass. Its crystal structure is not like that of ordinary metals but resembles that of the diamond. In the highest purity so far available, it shows no ductility; it is excessively brittle. The corrosion-resistance is so satisfactory that anyone who can find out how to make ductile wrought silicon has a rich reward waiting for him. A rich reward also awaits the chap who can make a tough, corrosion resistant, high silicon ferrous casting. However, as things stand, the metal itself is not commercially useful except as a reducing agent in smelting, and we must turn to its alloys for real applications.

11. In spite of the non-metallic type of lattice, it enters into solid solution with a few other metals and, up to a certain limit of silicon, the solution may be as ductile as though both metals were orthodox ones. Beyond that limit, it has, in ternary alloys, a strong tendency to combine into definite chemical compounds, silicides, with the third metal, if the third metal is of the proper class. Silicides of nickel, cobalt, manganese, iron and magnesium are among the best-known. Silicon declines to combine with, and in fact, to get along at all with, some other metals.

SOURCES OF SILICON

12. High silicon pig and ferrosilicon have been available to the steel maker for a long time, but sufficiently pure silicon for use in non-ferrous alloys without dragging a lot of iron along with it, has been available for only about 30 years, although cuprosilicon was made by Cowles at Lockport a few years earlier.

13. At present, cuprosilicon is made from metallic silicon, so nearly all the alloying additions of silicon stem back to the commercially pure metal, which is available in 98 per cent grade, the impurities being roughly approximately equal amounts of iron, aluminum and calcium. The price is not far from 15 cents per pound. On a volume basis, it is 10 per cent cheaper than 20-cent aluminum, and only one-third the cost of 12-cent copper. It is just about the same in volume cost as zinc.

USES OF SILICON

14. A CaSi alloy, the so-called calcium silicide, is used as a deoxidizer in cast iron, steel, especially stainless steels, and for some non-ferrous alloys. A CaMnSi alloy is similarly used. Since the CaO and MnO formed tend to slag the SiO_2 formed, there seems to be considerable virtue in the use of the complex "silicides" for deoxidation.

15. Silicon is not used, and probably is not useful, in alloys with tin, lead, or zinc as base. There are indications that it may be useful in silver, but silver does not fall within the scope of the ordinary non-ferrous foundry. For present purposes, we may confine our attention to silicon as an alloying element with magnesium, aluminum, copper and nickel.

SILICON IN MAGNESIUM ALLOYS

16. Because its specific gravity is very low, even lower than that of aluminum, and not much above that of magnesium, silicon is a logical choice for an alloying element in light alloys. Because of its brittleness, no success has attended what efforts have been made to produce usable alloys in which silicon predominates. With magnesium, silicon forms the chemical compound Mg_2Si. This compound is of peculiar interest in aluminum base alloys, but there is little as yet of commercial importance in the use of silicon in magnesium base alloys. Morioka[1] says magnesium-zinc-silicon alloys have improved corrosion resistance, and Bollenrath[1a] lists among Elektron casting alloys one with

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1. Superior numbers refer to bibliography at end of paper.

1.1 per cent silicon. The alloy contains Mg_2Si in a magnesium matrix. The properties given for sand castings are:

    14,000 — 18,000 lb. per sq. in. tensile strength
     7,000 — 9,000 lb. per sq. in. yield (0.2 per cent)
        2 — 4 per cent elongation
        4 — 7 per cent reduction of area
       41 — 46 Brinell

17. The silicon is considered to aid in producing sound, dense castings, but since the mechanical properties are inferior to those of the magnesium-aluminum alloys, the silicon alloy has only limited use, chiefly for pressure castings. The specific gravity is the lowest of any commercial magnesium casting alloy.

SILICON IN ALUMINUM ALLOYS

18. Though silicon is now well established as an alloying agent with aluminum, it was far from being in that state 25 years ago. All kinds of troubles were then ascribed to the silicon impurity in commercial aluminum, and some authorities were especially fearful of so-called "graphitic" silicon. In aluminum for electrical conductors, silicon lowers the conductivity but other disadvantageous effects are lacking. In castings in general, the old-time "enemy" of the aluminum foundryman has turned into a friend.

MODIFIED ALUMINUM-SILICON ALLOYS

19. The straight aluminum-silicon system illustrates the unpredictability of the action of silicon.

20. The metallurgist is accustomed to consider the eutectic of a binary system as a definitely fixed composition. If aluminum and silicon are melted in a normal manner, the aluminum-silicon eutectic is at 11.6 per cent silicon and the eutectic is of rather coarse structure. Melting beneath a sodium fluoride flux under suitable conditions, places the eutectic at 14 per cent silicon and at a lower temperature. The alloy behaves as though it were supercooled.

21. The same effect is produced by addition of about 0.015 per cent metallic sodium. The action of the flux has been traced to a slight reduction of sodium by aluminum — a mass action effect. The supercooled or "modified" alloy is of much finer structure and the properties, sand cast, rise from 20,000 lb. per sq. in. tensile strength and 3 per cent elongation for the regular,

to 27,000 lb. per sq. in. tensile strength and 12 per cent elongation for the modified alloy.

22. Incidentally, rate of freezing does not affect the modified alloy so much as it does most alloys, so that heavy sections do not differ greatly in properties from light ones.

23. These alloys were discussed at some length before the A. F. A. just as they were beginning to become commercial[2]. Many other discussions appeared about that time, [3, 4, 5, 6, 7, 8].

24. Hypereutectic alloys have not found much use, though recent German information[9] indicates that such alloys may have some place as bearing metals for service that is not too severe. Although the high-silicon alloys are interesting, there is much more actual use of those with lower silicon contents.

Low Silicon Alloys with Aluminum

25. Good corrosion resistance and remarkable fluidity in the mold are characteristic of the aluminum-silicon system. The aluminum-silicon alloys with 5 per cent silicon find use as architectural castings and even statuary, for these reasons. In resistance to atmospheric corrosion, the aluminum-silicon alloys have few peers among all the aluminum alloys.

26. From both the engineering and the foundry points of view, the ductility is the most important asset. The alloys not only show remarkable room temperature ductility, but it is also retained at high temperature. Hence, unlike many other aluminum alloys, the aluminum-silicon alloys are not hot-short and the tendency to crack in the mold around hard cores is reduced greatly. This pleasing behavior is held even with relatively low silicon content, down even to around 3 per cent.

27. A 5 per cent silicon alloy is used as welding rod in making the familiar aluminum furniture where the parts are assembled in jigs and the weld must give to the thermal contraction without cracking. The net thermal contraction is reduced and there is greater ability to resist what contraction stresses do occur. Such a welding rod is used for the great majority of aluminum and aluminum alloy welding and has been a veritable life-saver in this field.

28. The hot ductility conferred by silicon, plus the good running qualities and the corrosion resistance, make aluminum-silicon and aluminum-silicon with copper, or copper plus nickel,

much favored die casting alloys. The usual silicon content ranges from 1.5 to 5 per cent, while the straight 12 per cent silicon alloy also finds some use. One expert expresses it that silicon is the "staff of life" to the aluminum die casting industry. The die castings run around 30,000 lb. per sq. in. tensile strength with elongations of 1 to 3.5 per cent.

29. Permanent mold castings, as well as pressure die castings, are easier to make in the silicon-containing alloys.

PISTON ALLOY

30. One of the most interesting of the permanent mold castings is the "Lo-ex" piston alloy, with some 12 per cent silicon, 1 per cent magnesium, 1 per cent copper, and 2 per cent nickel, which, by precipitation hardening heat treatment, is brought to about 100 Brinell. The coefficient of expansion, due to the fact that silicon itself has a low coefficient, is 15 per cent lower than that of other suitable piston alloys, so that an advantageous closer initial fitting of the piston is possible. By proper design, alloys of higher expansion can be used with suitable slots or struts, but even in slotted designs, the lower expansivity is desirable. The alloy is also about 10 per cent lighter than other piston alloys.

31. Buick, Cadillac, Hudson, Pierce Arrow, and Studebaker, according to the listed specifications, carry these aluminum-silicon alloy pistons for 1938.

32. This is a good example of the fundamental properties of an alloying element, in this case the low expansivity and low specific gravity of silicon, carrying through to the alloy and finding engineering utilization.

MORE COMPLEX SAND CASTING ALLOYS

33. The modified 14 per cent silicon alloy is not notable for strength, as modern aluminum alloys go. The 5 per cent silicon alloy without other alloying elements, is still weaker. The proportional limits are very low, so that in the general run of sand castings, a compromise composition or range of compositions is favored. One scheme is to combine the precipitation hardening effect of the compound, Mg_2Si with the toughness of the aluminum-silicon matrix—so a 7 per cent silicon—0.3 per cent magnesium alloy, of favorable casting properties, can be employed and precip-

itation hardened to give a rather nice combination of properties—
say 32,000 lb. per sq. in. tensile strength, 22,000 lb. sq. in. yield
point and 3.5 per cent elongation. This alloy has a sp. gr. of
only 2.63, since both the added alloys are lighter than aluminum.
This alloy is good for pressure castings. The machinability of the
precipitation hardened alloy is better than that of the unhardened
straight silicon casting alloys.

34. Almost any proportion of copper-silicon in which silicon
is 3 per cent or more and the sum of copper + silicon about 10 per
cent, makes a good casting alloy with strength rather better than
No. 12, with slightly better ductility, in the cold, and vastly im-
proved castability, hot ductility and freedom from cracking. A
somewhat preferred composition is around 4.5 per cent copper,
5.5 per cent silicon, but the composition is not critical and any
practical foundryman can see the opportunity of sweetening up
his No. 12 scrap with aluminum and silicon when he has to make
castings that would crack with No. 12. Even 1 or 2 per cent silicon
is helpful in the aluminum-copper alloys [8a]. It is a far cry from
the old days when silicon was considered a poison.

HEAT-TREATED — (PRECIPITATION HARDENED) ALLOYS

35. The improvement in properties of alloys containing both
silicon and magnesium, in which the compound Mg_2Si is caused
to separate by a precipitation hardening treatment, has been re-
ferred to in connection with some of the alloys mentioned. Pre-
cipitation hardening first came to light by means of silicon.

36. Wilm, working in Germany with an alloy of aluminum
plus 4 per cent copper and 0.5 per cent each of manganese and
magnesium (now known as duralumin) noted that if this alloy
were cooled rapidly from the proper temperature and tested at
once, he got one set of values, but if it were allowed just to lie
around at room temperature, he got another set. The alloy becomes
harder and stronger and, a rather unusual feature, retains almost
as much ductility as it had in the soft, weak state. This was a
mystery till Merica[10], publishing his findings in 1919, postulated
and demonstrated the idea of precipitation hardening, due to a
sloping solubility line in the equilibrium diagram.

37. No silicon was intentionally added, but the amount car-
ried as impurity by commercial aluminum was sufficient to produce
an effective amount of Mg_2Si. The precipitation hardening effect
found in duralumin was at first ascribed solely to separa-

tion of $CuAl_2$, but it was later found that both $CuAl_2$ and $MgSi_2$ are involved. Further work unscrambled the alloy into one hardened by $CuAl_2$ and another hardened by Mg_2Si, both of which are commercially useful. It happens, however, that neither of the unscrambled alloys will strengthen at a very notable rate at room temperature. Heating is necessary. The simultaneous presence of the two compounds, however, causes the precipitation to occur readily at room temperature.

38. Hence, it is fair to credit silicon with producing the accidental results that called Wilm's attention to the phenomenon and that set a problem for Merica to untangle.

39. Although the heat-treated alloys of the Al-Mg_2Si and Al-$CuAl_2$ series are most extensively used in wrought forms, they are in wide use as castings. By increasing the silicon over that of the original duralumin, a so-called "super duralumin" is produced with enhanced mechanical properties.

Mechanism of Precipitation-Hardening

40. Not only in the duralumin family of silicon-containing alloys, but in many of the copper-base and nickel-base alloys with silicon, precipitation-hardening phenomena are to be reckoned with, so it is well to review the mechanism involved.

41. This has been so clearly described by Davenport and Bain[11] that the most effective way to recall it is to show their figures, Figs. 1 and 2. Fig. 1 shows the slanting solubility line necessary for precipitation hardening. In a system having such a line, a composition lying in the right hand corner of the solid solution field, heated above the temperature of complete solubility, is in a potential condition for precipitation hardening.

42. If the alloy can, by quenching or relatively rapid cooling, be supercooled below the solubility line without the separation of the second constituent, or compound, that would separate on sufficiently slow cooling, the solid solution is supersaturated. By regulated reheating of the supersaturated solution (or as in the unusual case of duralumin, by aging without reheating), the second constituent or compound, as the case may be, is caused to separate. As is shown in Fig. 1, at some low temperature level, the separated particles are extremely fine.

43. According to Edwards, Frary and Jeffries[12], in duralumin and the like, the particles are of the order of $1/10,000,000$ in.
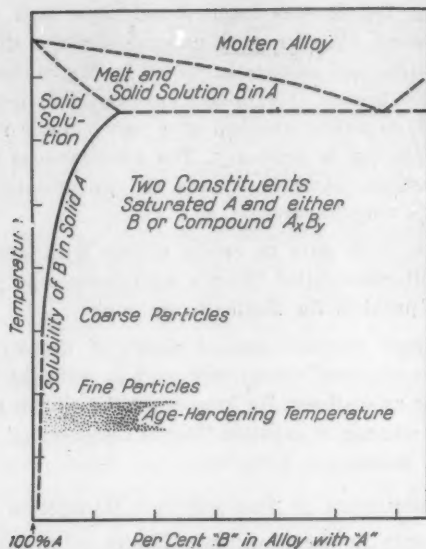
FIG. 1—TYPICAL CONSTITUTION DIAGRAM ILLUSTRATING THE SOLID SOLUBILITY CHARACTER-
ISTICS NECESSARY FOR AGE HARDENING (DAVENPORT AND BAIN).
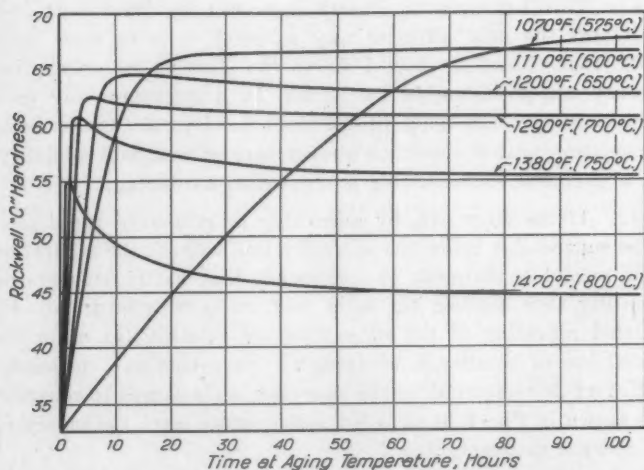
FIG. 2—TYPICAL AGE HARDENING OR QUENCH AGING BEHAVIOR. THE HARDNESS OF A
QUENCHED 75 PER CENT IRON— 25 PER CENT TUNGSTEN ALLOY AFTER AGE HARDENING
AT VARIOUS INDICATED TEMPERATURES (SYKES).

diameter, *i. e.*, of sub-microscopic fineness. At the optimum disper-
sion, the maximum of hardness and strength is brought about by
these particles. If the temperature is raised, the particles spheroid-
ize and agglomerate to something of the order of 1/10,000 in.
diameter, when they can be readily seen under the microscope,
the hardening power is gone, and they are now merely inclusions
without appreciable strengthening effect.

44. The separation of the fine particles from the supersatur-
ated solid solution is a time-temperature phenomenon. If the lowest
temperature is used at which atomic mobility is just sufficient to
allow precipitation and this temperature held for a sufficiently
long period, generally days or weeks, the alloy hardens to its
maximum, spheroidization does not go on, and the hardened alloy
is stable. The temperature levels at which such precipitation occurs
without agglomeration vary with the alloy. In duralumin, it is at
room temperature. In 75 per cent iron 25 per cent tungsten, it is
at 1070°F., and in the average run of alloys, it is well above room
temperature.

45. Ordinarily, one does not wish to heat as long a time as
is required to secure precipitation without agglomeration, and
prefers to heat a bit higher and for a shorter time, thus securing
precipitation but at the same time having some agglomeration, so
that something less than maximum hardness is achieved. If this
aging temperature is well above the temperature of service, one
has an added assurance that no further agglomeration will go on
at the lower temperature, so the alloy will be stable and permanent-
ly retain the properties secured by the heat-aging.

46. The effect of increasing the aging temperature is shown
in Davenport and Bain's Fig. 2. In that particular case, if one
is satisfied with 62 Rockwell "C", the aging treatment would be
5 hrs. at 1290°F. rather than 55 hrs. at 1070°F., for obvious com-
mercial reasons. But if we try to push the treatment to too high
a temperature, say 1470°F., the agglomeration phenomenon pulls
ahead and the hardness level is quite inferior. Going still higher
will produce full agglomeration, *i. e.*, the alloy will be fully an-
nealed and softened. Fig. 2 shows the action of a well-behaved,
controllable, precipitation hardening system.

47. All systems with a slanting solubility line are not well-
behaved and controllable. In some, the tendency toward separation
of particles is so great that sections large enough to be useful

cannot be quenched quickly enough to prevent precipitation and thus hold the supersaturated solid solution ready for controlled reheating. Others are so sluggish that the solid solution is retained on slow cooling, as in a casting, without needing any quench.

48. Some alloys, that can be quenched to supersaturated solid solution, do not allow a separation of the precipitation and agglomeration effects and hence are not sufficiently controllable on reheating to make a reheating treatment useful. The sloping solubility line is a prerequisite to, but not an assurance of, ability to be precipitation hardened. However, the sloping line is always a challenge to the metallurgist to find out whether, by addition of other alloying elements, he cannot bring the precipitation and agglomeration phenomena definitely under control.

49. In the copper-base and nickel-base alloys of silicon, sloping lines exist, and the behavior and properties of the alloys are related to the response or lack of response to precipitation hardening.

ALPHA ALLOYS OF COPPER

50. It is well known that there is quite a similarity in the properties of alpha solid solutions of other elements in copper, so much so that the term "bronze" is often applied to such alloys, even though they may contain no tin. The alpha alloys increase in strength as the alloying element is added up to its limit of solubility, and up to this limit the alloys retain a good deal of toughness.

51. When the solubility limit is exceeded and appreciable amounts appear of the next phase (beta, at least at high temperature, but in some systems beta itself transforms), strength rises considerably and toughness falls quite sharply. If a suitable small amount of beta or its decomposition product can be produced, a compromise between strength and toughness can be had, but such a duplex alloy is usually quite critical as to composition; with too high copper, it is not up to the optimum in strength, and with too low copper, it may be far too brittle.

52. Such alpha-forming alloys as tin, zinc, aluminum, and silicon can replace each other, their replacing power being in approximate inverse ratio to the maximum percentage that can be held as alpha at any temperature, i. e., about 32 per cent zinc, 12 per cent tin, 7 per cent aluminum or 4 per cent silicon, or combinations of these in which one part of tin is replaced by 2 parts of zinc, 2/3 part of aluminum or 1/5 part of silicon, give pure

alpha solid solutions. Of course, nickel and manganese also form alpha solutions with copper over a wide range of compositions, and have replacing power for other alpha-formers. The position and slope of the binary boundary curve will, of course, be modified by the addition of another element, and if two of the additions combine to form a compound, the state of affairs, outside of the alpha field, will be specific to the composition.

COPPER-SILICON ALLOYS

Binary Alloys

53. Very little use is made of binary alpha copper-silicon alloys. One with only 0.10 per cent silicon is used for strong high conductivity wire, because excess silicon left after deoxidation is not so harmful to conductivity as other deoxidizers, especially phosphorus. As Clamer[13] points out, silicon, as silicon-copper, is especially useful as a deoxidizer in alloys free from lead. A little lead in an alloy containing much silicon, or a little silicon in one containing much lead produces drossy, dirty castings. Even a few hundreths of a per cent of either in the presence of the other raises hob in ordinary practice. Some wrought alloys with silicon are intentionally leaded, but it is probably necessary to deoxidize these very carefully before letting the silicon and the lead get together. Tellurium has been suggested instead of lead, to improve machinability of copper-silicon-zinc alloys[13a].

54. Perhaps the most important use of straight copper-silicon is as welding rod. Such an alloy is used for joining copper rail bonds to steel rails and in general non-ferrous welding. Slight modifications of the binary alloy find a wide variety of uses.

COPPER-TIN-SILICON ALLOYS

55. The chief urge to use an alpha copper-silicon alloy offhand, would seem to be as a substitute for tin, on account of the wide difference in cost and the smaller amount of silicon required. Davis[14] mentions a wrought alloy of 2 per cent tin 0.75 per cent silicon, but data on cast copper-tin-silicon alloys are very meager.

56. In view of the calculation that 4 per cent silicon is equivalent to 10 per cent tin, it is surprising to find so little information on bronzes of fairly high tin content plus silicon. Hanson and Wheeler[15] briefly studied wrought alloys, the only data on the as-cast state being that 2 per cent silicon 5 per cent tin had a Brinell of 115 and one with 3 per cent silicon, 5 per cent tin, 150.

They comment on the very strong hardening effect of silicon and the brittleness when the solubility limit is overstepped.

57.   From the work of Guillet, Ballay and Le Thomas[16], it would appear that one might replace part of the tin in a true bronze by 1/3 as much silicon as the tin that was left out, as far as tensile and hardness go. Toughness, and especially the bearing properties, might not be retained. However, it would be interesting to try making bells from the ternary alloy.

COPPER-ZINC-SILICON ALLOYS

58.   Silicon is commercially used in brass. Gould and Ray[17] tried adding silicon to regular brasses and found that 2 to 3 per cent silicon in 85:15 and say 1 per cent in 60:40 had interesting properties. Koehler[18] in Germany, advocates about 15 per cent zinc and 3 per cent silicon. Lebeder[19] advocates 10 per cent zinc, 5 per cent silicon which gives him 62,000-74,000 lb. per sq. in. tensile strength, but which he says must not be overheated in melting and must be poured cold if cracking is to be avoided. Unpublished work in this system shows, at other compositions, a still more interesting combination of properties, but a degree of erraticness that has so far held them back from much actual use.

59.   An important effect of silicon in brasses is to make them more easily weldable. The presence of the SiO_2 skin produced, holds back the volatilization of zinc even from brasses quite high in zinc, thus making them more suitable for welding.

60.   An alloy for pressure die casting[20] contains about 4 per cent silicon, 15 per cent zinc with 0.15 per cent manganese. Under the chill casting conditions of pressure die casting, it gives 85,000 lb. per sq. in. tensile strength and 8 per cent elongation. If gravity cast in a sand or plaster mold, the strength is halved. Of course, the die casting of even this fluid alloy is still something of a task compared with die casting of the white metals. In building up an alloy for pressure casting, one will pile in as much silicon and zinc as he dares in order to hold the melting point down. As this industry advances, silicon probably will be found as essential in copper-base die casting alloys as it has proved in aluminum-base die casting alloys.

61.   Wrought alloys of about this composition but with the silicon even up to 5 per cent are known, under the names of Tombasil, Brastil and Silzin. Tanabe and Koiso[21] in alloys which

they designate "SSZ," use a bit lower silicon, 3.5 per cent and add 1 per cent aluminum along with 15 per cent zinc—or in those with 10 per cent zinc, the silicon is held at 4 per cent and 1 or 2 per cent aluminum used.

62. Cadmium could be used in place of zinc, and it is stated[22] that an alloy of 3 per cent silicon 0.5 per cent cadmium has superior casting properties in making billets for wrought products, to the straight silicon alloy. The casting properties of the zinc-containing copper-silicon alloys are so good and the difficulties met so much reduced, that a few per cent of zinc would be present in all the copper-silicon casting alloys were it not that the corrosion resistance, for most uses, is not so good in such alloys as in those that are zinc-free.

COPPER-SILICON-MANGANESE ALLOYS

63. Alpha solid solutions of copper with silicon and manganese, tin, aluminum or iron in moderate amounts are, broadly speaking, as corrosion-resistant as pure copper, perhaps more so in some environments. Wrought alloys of the copper-silicon series with small additions of one or more other elements not harmful from the corrosion point of view, have assumed an important industrial position, especially since such alloys are weldable with proper technique. The advantages of a weldable material with the corrosion-resistance of copper and very much higher strength obtained with cheap alloying elements, are obvious for such uses as tanks for hot-water heaters and for the chemical industry in general.

64. Most of these wrought alloys are probably straight alpha, such as the low (1.5 per cent silicon, 0.25 per cent manganese) types of Everdur, and Silcudur[23] (with 2.2 per cent silicon and 0.7 per cent manganese). Silcudur is said to be difficult to cast, even for ingots for wrought products but to give sound ingots if not overheated. Most of the wrought alloys run higher in silicon, for example the grade of wrought Everdur one hears the most about, carries 3 per cent silicon and 1 per cent manganese. As is seen in Figs. 3 and 4[24], the boundary is moved to lower silicon by the introduction of 1 per cent manganese (i.e., the silicon solubility is reduced), and a new field, for Mn_2Si, appears.

65. The Everdur composition recommended[25] for castings is 4 per cent silicon, 1 per cent manganese, just at the border line for production of some Mn_2Si with absence of the gamma phase.
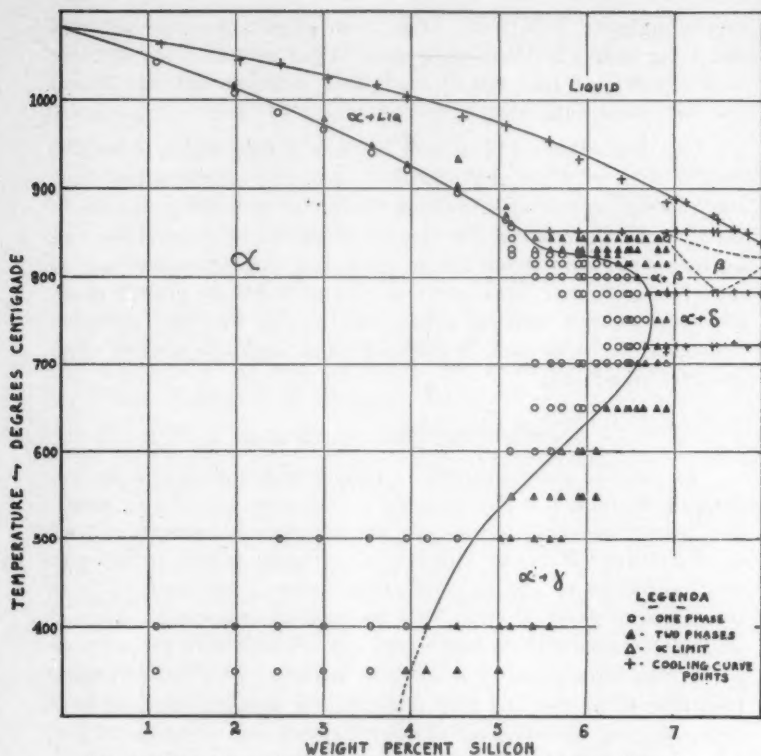
FIG. 3—ALPHA-PHASE BOUNDARY OF THE COPPER-SILICON SYSTEM (SMITH).

The gamma phase is sometimes considered to be essentially the compound Cu_5Si. Its presence in appreciable amounts leads to brittleness.

66.   The 4:1 cast Everdur is stated[26] to give, as normal properties 50,000 lb. per sq. in. tensile strength, 20,000 lb. per sq. in. yield point, 20 per cent elongation, 25 per cent reduction of area.

67.   Clause and Goederitz[27] mention "Isima" alloys of such compositions as 12 per cent manganese and 3 per cent silicon, or 5 per cent manganese and 1.2 per cent silicon whose properties are not stated but which are said to be capable of doubling their hardness by suitable precipitation treatment. Smith shows the solubility line for the 5 per cent manganese alloy to be of the same general type as in Fig. 1. He indicates that alloys of 5 per cent

manganese with 1 to 3 per cent silicon ought to respond to precipitation hardening treatment.

VARIANTS OF THE COPPER-SILICON ALLOYS

68. Analogous to the various grades of wrought copper-silicon-manganese alloys are those with 1.5 per cent and with 3 per cent silicon with additions of other elements than manganese. Herculoy[28], Olympic Bronze and Duronze have such silicon contents; Herculoy contains also about 1.5 per cent zinc and 0.5 per cent tin. Castings of this alloy are said[28] to give 55,000 lb. per sq. in. tensile strength, 25,000 lb. per sq. in. yield point, 40-50 per cent elongation, 25 per cent reduction of area.

69. Olympic bronze is an alloy of this class, which contains 1 per cent zinc. Duronze is another trade name given to "high copper-silicon alloys containing small quantities of other metals," probably chiefly tin. Duronze castings (grade 2) are said to show 40,000 lb. per sq. in. tensile strength, 30 per cent elongation,
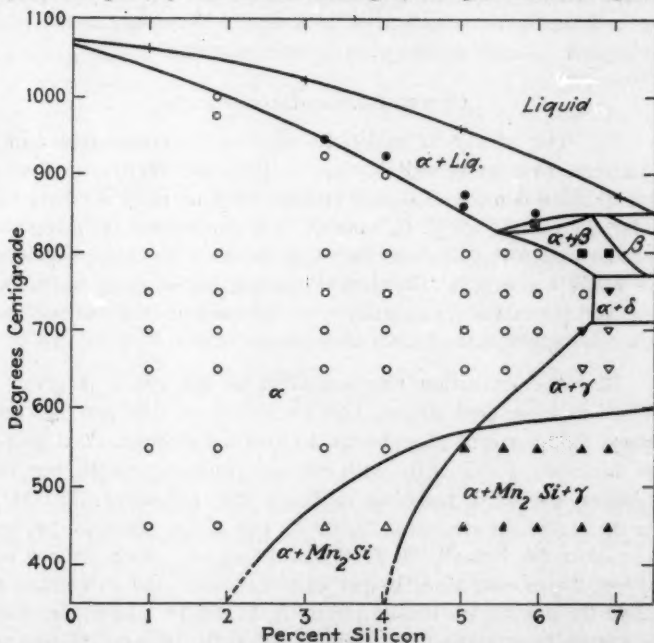
FIG. 4—SECTION OF THE TERNARY COPPER-MANGANESE-SILICON DIAGRAM AT 1 PER CENT MANGANESE (SMITH).

40 per cent reduction of area, all minimum. Melting under a glass
or sand cover, avoidance of heating over 2200°F., and allowing
the metal to cool before pouring to release gas, are advocated.
Dietz and Weiser[29] favor an alloy of 2.75 per cent silicon, 1 per
cent tin, 0.25 to 0.50 per cent manganese.

70. It appears that either manganese or zinc is useful as a
deoxidizing addition to the copper-silicon alloys, one or the other
appearing in nearly all the alloys.

71. The use of silicon, without other deoxidizer, in steel may
produce an embrittling type of oxide inclusion, while the silicon-
manganese combination may avoid this. Whether the analogy is close
or not in the copper alloys, the copper-silicon alloys seem to behave
better when manganese or some other deoxidizing element is
present. An alloy, somewhat used in England, according to
Arnott[30], contains not more than 4 per cent silicon without man-
ganese, but with 2 per cent zinc and shows 36,000-50,000 lb. per
sq. in. tensile strength, 16,000-22,000 lb. per sq. in yield, 10-25
per cent elongation, and 100-120 Brinell. These are rather wide
limits and indicate a somewhat erratic behavior.

COPPER-SILICON-IRON ALLOYS

72. The utility of additions of iron to manganese and to
aluminum bronzes is well known, so it is not surprising that an
iron addition is made to silicon bronze. Such an alloy is commercial
under the name "P. M. G." metal. Definite data correlating com-
position and properties are scarce, since the alloy is a proprietary
one made up with a "hardener" containing copper, silicon and
iron, and the mixes are calculated on the basis of so much hardener,
so much copper, and if desired, so much zinc.

73. A composition representative of one grade is given by
Roast[31] as 4 per cent silicon, 1.50 per cent iron, 0.05 per cent man-
ganese, 0.05 per cent phosphorus, and his list of mechanical proper-
ties includes, for P.M.G. with 90 per cent copper 10 per cent
hardener for acid resisting castings, the following: 50,000 lb.
per sq. in. tensile strength, 20,000 lb. per sq. in. yield, 30 per cent
elongation, 96 Brinell, 20 ft.-lb. Izod impact. With 88 per cent
copper, 2 per cent zinc, 10 per cent hardener, the properties are
52,000 lb. per sq. in. tensile strength, 24,000 lb. per sq. in. yield,
25 per cent elongation, 95 Brinell, 16 ft-lb. impact. If the per-
centage of hardener is increased to 15 per cent and no zinc used,

the properties given are 40,000 lb. per sq. in. tensile strength,
33,000 lb. per sq. in. yield, 3 per cent elongation, 120 Brinell,
5 ft.-lb. impact.

74. Properties stated by the Cramp Brass & Iron Foundries
Company for one grade of P. M. G. are 45,000 lb. per sq. in. tensile
strength, 20,000 lb. per sq. in. yield, 15 per cent elongation, all
minimum.

75. Incidental data are available in discussions of the cavi-
tation problem by Mousson[32] and by Kerr[33]. The former gives for
an alloy of 92.43 per cent copper, 1.81 per cent zinc, 0.08 per cent
manganese, 3.67 per cent silicon, 0.67 per cent iron (and 1.34
per cent missing from 100 per cent!) 50,000 lb. per sq. in. tensile
strength, 28,000 lb. per sq. in. yield, 18 per cent elongation, while
the latter gives for 94 per cent copper, 5 per cent silicon, 1 per cent
manganese, 45,000 lb. per sq. in. tensile strength, 22,000 lb. per
sq. in. yield, 20 per cent elongation; for 92-94 per cent copper,
3-4 per cent silicon, plus iron, 45,000 lb. per sq. in. tensile strength,
22,000 lb. per sq. in. yield, 18 per cent elongation, and for 92-94
per cent copper with 3-4 per cent silicon, plus zinc, iron and
aluminum, 72,000 lb. per sq. in tensile strength, 35,000 lb. per
sq. in. yield, and 15 per cent elongation.

76. The use of iron in the alloys opens the door to getting
some of the required silicon from high-silicon pig iron in the
making of the hardener and thus decreasing the cost of the silicon.
One could, so to speak, extract silicon and some iron from molten
pig iron by means of molten copper, though it seems doubtful if
such a process would make the final alloy as dependable in com-
position as would be the case were it made up from copper, cupro-
silicon, and iron in a fashion analagous to the making of manganese
bronze.

77. According to Brice[34], the solubility of iron in the alpha
copper-silicon alloy is very low. According to Kinzel[35], the proper-
ties of P.M.G. probably are due to separation of an iron silicide.

78. One user of P.M.G. has commented that it was desirable
to let the castings cool way down in the mold and that this is
probably connected with the formation of some such compound.
P.M.G. metal is also available in wrought form.

FOUNDRY BEHAVIOR OF THE COPPER-SILICON-ETC. CAST ALLOYS

79. The cast alloys are highly attractive in metal cost and in mechanical properties. They are extraordinarily fluid and on these scores are attractive to the foundryman. Nevertheless, they have been a thorn in the side of almost every foundryman who has struggled with them because of their very high shrinkage. To keep them sound and free from shrinkage cavities, they must have ample gates and very heavy risers, preferably blind ones, to delay their cooling as much as possible.

80. Those who approach the casting problem as they would one in cast steel, manganese bronze or aluminum bronze, can work out the feeding problem and that of avoiding anchoring in the mold that would prevent free contraction and cause cracking, but those who approach the alloys with the idea that tin bronze technique can be applied, will be due for a rude awakening.

81. The idiosyncrasies of the silicon bronzes are well brought out in the discussion by Roast[31]. Beside the feeding problem, it must be remembered that the alloy compositions usually are very close to the solubility limit and a slip in composition that takes them over the line and produces the gamma phase will mean a brittle casting. Moreover, the cast alloys are sensitive to even traces of lead, as Patch[36] points out, so that getting any lead-containing scrap mixed up with the charge or melting in a hearth-type furnace that has previously melted leaded alloys, is prone to cause trouble. If zinc is added, high grade, lead-free spelter must be used.

82. Further, the alloys are very gas-sensitive, reminding one of "over-reduced" steel. They seem to take up hydrogen, either as such from a smoky flame or that formed by reaction with the moisture in the products of combustion of gas or oil. Electric melting (avoiding a hearth that is contaminated with lead), or a coke fire would simplify avoidance of hydrogen pick-up, though crucible melting with a glass cover for the melt to keep hydrogen and water vapor away is used with success. Degasification by treatment with calcium boride is also reported as useful.

83. The difficulties seem to be somewhat less when appreciable amounts of manganese or zinc are used. How much the zinc lowers the corrosion resistance is a matter of argument, but foundrymen

in this country hesitate to use much zinc since the greatest field for the copper-silicon castings is in corrosion resistant service.

84. British comments indicate that the zinc addition is common over there, and high corrosion resistance is even claimed for an alloy called Tungum, containing 14 per cent zinc, 1 per cent aluminum, 1 per cent nickel, 0.25 per cent iron and 1 per cent silicon. If the claim is correct, the aluminum and silicon must be quite potent in protecting the brass from corrosion. Wrought products are particularly aimed at with this alloy, but castings are also made. Rowe[37] comments on the low zinc alloys of copper-silicon-zinc that, in spite of the much better mechanical properties and lower cost than gun metal, they have made no marked headway on account of the high liquid shrinkage.

COPPER-ALUMINUM-SILICON ALLOYS

85. The addition of silicon to aluminum bronze was discussed by Cowles[38] in the late '80's. He mentions an alloy of 9 per cent aluminum and 1 per cent silicon.

86. Published information on silicon additions to aluminum bronze seems to be chiefly from the metallographic point of view. Brice[34] and Herman and Sisco[39] comment on the excessive hardness and brittleness when silicon over the alpha solubility limit is added to aluminum bronze.

87. Information on an alloy of 90 per cent copper, 7 per cent aluminum, 3 per cent silicon, has kindly been supplied by F. L. Wolf of the Ohio Brass Company. He reports that this alloy has been in use for high strength castings for 5 years with no reports of failure. It gives 70,000 lb. per sq. in. tensile strength, 50,000 lb. per sq. in. yield (0.02 in. extension), 5 per cent elongation as cast and requires no age hardening treatment. The alloy is sensitive to lead so pure metals are required in its manufacture, but is not sensitive to gasiness if proper care is taken in melting and casting.

ALLOYS HARDENED WITH Ni_2Si

88. Following the work of Corson[40] on precipitation hardenable alloys containing various silicides, there was a great deal of activity with the so-called "Tempaloy" type, for example, 95 per cent copper, 4 per cent nickel, 1 per cent silicon. Naturally, the corrosion resistance of such an alloy would be very high, the cast-

ing properties good, and very high strengths were attainable by precipitation hardening treatment. One foundry spent years of active and intelligent effort in trying to control the casting and heat-treatment variables, but finally gave up the attempt, probably because of the shrinkage difficulties. No record of this work has been published.

89. These alloys have been studied[41] chiefly in respect to wrought materials. If the nickel in a copper-nickel-silicon alloy is tied up with silicon as the compound Ni_2Si, the matrix is left with much the properties of copper. So when a high electrical conductivity plus greater hardness than is possible with deoxidized copper is required, such an alloy finds use. Castings for welding electrodes, electric furnace electrode holders and the like, are being made commercially with about 2.5 per cent nickel, 0.5 per cent silicon, which can be precipitation hardened to a useful combination of hardness and conductivity, e.g., 173 Brinell and 41 per cent[41] though not so good as can be gotten with copper-chromium, or copper-chromium-silicon. Foundry difficulties necessarily will be present in any high conductivity alloy because of the high copper content. Any of these high-conductivity alloys are touchy.

90. Knowing that the presence of zinc avoided some of the troubles of Tempaloy, the Ohio Brass Company investigated a copper-nickel-silicon-zinc alloy. This is best described in Mr. Wolf's own words:

"We first started to experiment with silicon alloys about six or eight years ago. After considerable work, we finally decided that our best bet was an alloy of the following composition:

	Per Cent
Copper	79.5
Nickel	4.0
Silicon	1.5
Zinc	15.0

"After casting, the alloy was given a heat treatment of 1450 to 1500°F. for one hour, water quenched and age hardened at 800°F. in a Homo furnace for 6 hrs.

"If everything 'clicked' right (we always crossed our

fingers), this metal would give the following physical proper-
ties:

Tensile............50,000 to 60,000 lb. per sq. in.
Yield..............50,000 lb. per sq. in.
Elong..............1 to 2.5 per cent
Izod Impact.......about 5 ft.-lb.
Brinell............140 to 180

"I do not recall that we had much trouble with gasiness
but, like alloys of this type, it did have fairly high shrinkage
and castings required a lot of feeding.

"The alloy was extremely sensitive to lead, 0.2 per cent
or less produced a very coarse grain structure which was ac-
companied by an intercrystalline weakness which led to cracks
between the crystals during solidification.

"The effect of lead was minimized by pouring at as low a
temperature as possible. However, we had jobs with such thin
sections that we had to pour on the high side. We finally dis-
covered that we had to use electrolytic copper and the purest
kind of zinc, but even so we ran into contamination at times.

"About six months after castings reached the field, we
began to receive a rebound, reports of failures all over the
country, and we were in a terrible mess with some of our
largest customers. We found that the alloy in question was
just too sensitive and difficult to handle in a commercial way,
so we threw it overboard and in its place developed the copper-
aluminum-silicon alloy referred to above."

There is no apparent reason why the zinc in the copper-nickel-
silicon alloy referred to above, should cause trouble, and it is not
known what actually lay beneath the difficulty. Castings of
straight copper-nickel-silicon, precipitation hardened alloy have
been in service for long periods without failure. Most precipita-
tion hardening alloys, aged at elevated temperature, are thoroughly
stable at ordinary temperature, so if the phenomenon studied by
Wolf were studied further, an explanation might be found that
would lead to avoidance of the trouble he met.

SILICON IN CUPRONICKEL

91. For large practical application of the nickel-silicon com-

bination one goes to the higher nickel contents of the copper-nickel series.

92. In the cupronickels of around 70 per cent copper, 30 per cent nickel, as Kihlgren[42] has recently shown, around 1 per cent manganese plus 0.5 to 0.75 per cent silicon produce noteworthy effects. Silicon strengthens this alloy in an amazing fashion. The use of manganese plus silicon is to be noted; with silicon alone, the melt is too sluggish. Even traces of lead are fatal to this alloy.

SILICON IN MONEL

93. If the proportions of copper and nickel are reversed, *i.e.*, to the ratio 70 nickel—30 copper, as in monel metal, and to this is added silicon and precipitation hardening carried out, alloys with steel-like strength but with the corrosion resistance of monel are obtained.

94. Monel, as a natural alloy, already carries about 1.5 per cent silicon and 0.5 to 0.75 per cent manganese. If the silicon is increased to 3 or 4 per cent, considerable amounts of iron can be tolerated, so the silicon can be added as ferro. The strength, as-cast, rises from 70,000 lb. per sq. in. tensile strength, 35,000 lb. per sq. in. yield, 25 per cent elongation, 135 Brinell, progressively to 100,000 lb. per sq. in. and over tensile strength, 75,000 lb. per sq. in. yield, with only 1 per cent elongation but a hardness of 300 Brinell.

95. When the silicon rises over 3 per cent, by slightly increasing the copper, the solubility limit of Ni_2Si is reached, and precipitation-hardening becomes possible. By staying on the alpha side, one can get 100,000 lb. per sq. in. tensile strength, 65,000 lb. per sq. in yield and 10 per cent elongation.

96. If the silicon is brought up close to 4 per cent, the tensile strength, as-cast, stays at around 100,000 lb. per sq. in., the yield rises to 75,000 lb. per sq. in., elongation falls to about 4 per cent at around 245 Brinell, which is a bit hard to machine, that is, the alloy precipitation-hardens on cooling in the mold However, by quenching, the alloy can be softened to 160 Brinell. It then may be machined, reheated and precipitation-hardened back to the as-cast properties or even a trifle harder. The precipitation-hardened alloy with its Ni_2Si particles, has non-galling properties, useful in valve seats.

97. For still higher gall resistance, the silicon can be taken above 4 per cent and the as-cast Brinell of 375 can be softened to 275 and again raised to 375 by reheating, but such an alloy is devoid of ductility.

98. If this family of alloys is built up from the metals rather than from monel scrap, it is necessary to add manganese. A trace of magnesium also is used ordinarily for deoxidation and to combine with sulphur, which is the absence of magnesium is deleterious to nickel alloys.

SILICON IN OTHER NI-BASE ALLOYS

99. Still another alloy of the valve seat class, has been suggested by Bolton[43] in which equal percentages of copper and nickel are used together with 2 per cent silicon. This also softens on quenching and hardens on reheating. A nickel-base casting analog of Duriron, Hastelloy D with some 87 per cent nickel, 3 per cent copper, 10 per cent silicon has good resistance to H_2SO_4, fair to HCl, and is said to be tougher than Duriron. It shows 35,000 lb. per sq. in. tensile strength with zero elongation.

GENERAL SUMMARY

100. The alloys of aluminum with silicon have advantages that have been demonstrated and have come into very wide use. Many of them are covered by standard specifications and they are supplied by all up-to-date aluminum foundries. The use of silicon tends to make aluminum alloys more fool-proof.

101. The silicon-bearing monel and other nickel alloys will evidently reach a position of importance.

102. The wrought copper-silicon alloys, with small additions of other metals, have won a definite place among corrosion resistant alloys. The cast alloys are still in the proprietary and trade-name stage. The casting alloys that are supplied to licensees as ingot or for which hardeners are supplied so that the foundryman has to use what he is given instead of building his alloys as he sees fit from the ordinary metals and rich alloys of commerce, are not necessarily the last word. It would seem quite possible that other elements such as aluminum or iron might serve a useful purpose in the copper-silicon-manganese series, and it seems odd, in view of all the evidence as to the utility of manganese in the copper-silicon and nickel-silicon series, that manganese is not used in the copper-

silicon-iron series. The proponents of P.M.G. metal state, however, that manganese additions have been studied and are not helpful.

103. Cast material of the same type is required in installations using wrought copper-silicon alloys for corrosion resistant purposes, so that market seems assured. Some of the cast alloys have given interesting results in service involving both corrosion and cavitation-erosion.

104. Outside of rather specialized service where corrosion resistance outweighs everything else, the cast alloys, though potentially useful for high strength and doubly desirable because of low metal cost, have made rather slow progress because of the foundry difficulties in handling them. Valve makers, for example, are using them for special work but are not substituting them widely for tin bronzes where the latter will serve. This probably means that when the overall cost is computed, the ordinary bronze has enough advantage in easier handling in the foundry and in ability to utilize scrap, to overbalance the lower material cost of the copper-silicon alloys.

105. It seems obvious that these alloys are likely to prove troublesome in the hands of inexpert foundrymen, but are suitable for expert foundrymen who specialize in alloys of high shrinkage characteristics and understand their requirements. Very close control of gating and heading, composition, and melting practice are required. Not only are the alloys themselves touchy, but if silicon-containing scrap gets mixed up with charges of ordinary leaded bronzes, the silicon will injure those mixes.

106. Nevertheless, it seems quite in order to expect that as interested foundrymen experiment with the copper-silicon series of alloys and develop better knowledge of how to handle them and how to adjust the composition to secure the best compromise between foundry behavior and mechanical properties, the alloys may attain a much more important position among engineering alloys. Even the old standby, manganese bronze, has had its preferred composition altered as experience accumulated. Moreover, even manganese bronze is not yet successfully produced by every brass foundry that tries to make it.

107. These alloys have tantalizing peaks of excellence, as to the mechanical properties, they show wonderful fluidity, and their raw materials are among the cheapest of the non-ferrous metals. Now that the sensitivity to lead, the necessary precautions for

melting without gassing, and the necessity for feeding are becoming more clearly understood, the production of sound castings can be tackled with greater assurance than in earlier days. But it will be a long time before the various series of copper-base alloys with silicon become fool-proof.

108. However, you cannot keep a cheap alloy down when it has the potentialities possessed by the copper-silicon family. Foundry metallurgists are ingenious animals, who have solved equally tough problems. Some day, they will work out the kinks of this problem and develop alloys and methods that utilize the known advantages and avoid the disadvantages conferred by silicon.

Bibliography

1—Morioka: Nippon Kinzoku Gakukai Si, vol. 1, June, 1937, pp. 80-91.

1a—Bollenwrath, F., *"Physikalische und mechanische Eigenschaften der Magnesium legierungen in Werkstoff Magnesium,"* V D I Zeit., vol. 82, 1938, pp. 5-28. See also: Luyken, H. "Sand und Kokillenguss von Magnesium legierungen," IBID, 89-93.

2—Guillet, L., *"A New Aluminum Alloy—Alpax,"* Trans., AMERICAN FOUNDRYMEN'S ASSOC., vol. 32, Part 1, 1924, pp. 309-349.

Basch, D., and Sayre, M. F., *"Foundry Treatment and Physical Properties of Silicon-aluminum Sand Castings,"* IBID., vol. 32, Part 1, 1924, pp. 350-394.

3—Archer, R. S., Kempf, L. W., and Hobbs, D. B. *"Heat Treatment of Aluminum-silicon Alloys,"* TRANS., AM. INST. MIN. MET. ENGRS., vol. 78, 1928, pp. 198-227.

4—Archer, R. S. and Kempf, L. W., *"Modification and Properties of Sand-cast Aluminum-silicon Alloys,"* TRANS., AM. INST. MIN. MET. ENGRS., vol. 73, 1926, pp. 581-621.

5—Edwards, J. D. and Archer, R. S., *"The New Aluminum-silicon Alloys,"* CHEM. MET. ENGNG., vol. 31, 1924, pp. 504-508.

6—Edwards, J. D., *"Properties and Manufacture of Aluminum-silicon Alloys,"* CHEM. MET. ENGNG., vol. 27, 1922, pp. 654-655.

7—Edwards, J. D., *"Thermal Properties of Aluminum-silicon Alloys,"* CHEM. MET. ENGNG., vol. 28, 1923, pp. 165-169.

8—Jeffries, Z., *"Aluminum-silicon Alloys,"* CHEM. MET. ENGNG., vol. 26, 1922, pp. 750-754.

8a—Schofield, T. H., Phillips, C. E. and Archbutt, S. L., *"Note on the Effect of Silicon on the Mechanical Properties of 7% Copper-Aluminum Alloy,"* vol. 61, 1937, pp. 79-82.

9—Wiechell, H., *"On the Development, Construction and Use of Light Metal Bearings,"* Automobiltechnische Zeit., vol. 40, May 10, 1937, pp. 235-240.

10—Merica, P. D., Waltenberg, R. G. and Freeman, Jr., J. R., *"Constitution and Metallography of Aluminum and Its Light Alloys With Copper and With Magnesium,"* Bur. Stand., Bul. 337, 1919, 15. p.

11—Davenport, E. S. and Bain, C. E., *"The Aging of Steel,"* Trans., Am. Soc. Metals, vol. 23, 1935, pp. 1047-1106.

12—Edwards, J. D., Frary, F. C. and Jeffries, Z., *"The Aluminum Industry,"* New York, McGraw Hill Book Co., Inc., 1930. See vol. 2, p. 138.

13—Clamer, G. H., *"Hardeners for Copper and Copper Alloys,"* National Metals Handbook, Am. Soc. Metals, 1936, pp. 1118-1121.

—St. John, H. M., Eggleston, G. K. and Rynalski, T., *"Influence of Silicon In Foundry Red Brasses,"* Am. Inst. Min. Met. Engrs., Tech. Pub. 300, 1930, 18. p.

13a—Burghoff, H. L. and Lawson, D. E., *"Effect of Tellurium on Mechanical Properties of Certain Copper-base Alloys,"* Metals Technology, Sept. 1937, Tech. Pub. No. 844, 10 p.

14—Davis, C. H., *"Alloys of Copper,"* Am. Soc. Test. Materials, Symposium on High-strength Construction Metals, March, 1936, pp. 79-94.

15—Hansen, D. and Wheeler, M. A., *"Properties of Some Special Bronzes,"* J. Inst. Metals, vol. 57, 1935, pp. 93-108.

16—Guillet, L., Ballay, M. and Le Thomas, A., *"Influence of Silicon on Copper-nickel Alloys of Low Tin Content,"* Comt. rend., vol. 195, July 11, 1932, pp. 89-92.

17—Gould, H. W. and Ray, K. W., *"Effects of Silicon on the Properties of Brass,"* Metals and Alloys, vol. 1, 1930, pp. 455-457, 502-507.

18—Koehler, M., *"Ueber α und α-β Silico-Gussmessing,"* Dissertation, Tech. Hochschule zu Aachen, 1934, 52 p.

19—Lebeder, K., *"Silicon Brasses,"* Liteinoe Delo, vol. 8, No. 8, 1937, pp. 41-46.

20—Freeman, Jr., J. R., *"Die-casting of Brass,"* Trans., Am. Inst. Min. Met. Engrs., vol. 117, 1935, pp. 371-380.

21—Tanabe, T. and Koiso, G., *"Tetsu to Hagane,"* vol. 21, 1935, pp. 575-584.

22—American Brass Co., *"Improvements in or Relating to Copper-silicon Alloys,"* Brit. Patent 471,965, Sept. 10, 1937.

23—Anon., *"Copper-silicon-manganese Alloy for Water-heating Equipment,"* Metal Ind. (Lond.), vol. 51, 1937, p. 604.

24—Smith, C. S., *"Alpha-phase Boundary of the Ternary System Copper-silicon manganese,"* Am. Inst. Min. Met. Engrs., Tech. Pub. 292, 1930, 32 p.

25—Jennison, H. C. and Hibbard, W. R., *"Properties of Silicon*

Bronze (Everdur)," NATIONAL METALS HANDBOOK, Am. Soc. Metals, 1936, pp. 1108-1110.

26—Bedworth, H. A. and Weaver, V. P., *"Sand Casting Copper-Silicon Alloys,"* TRANS, A.F.A., vol. 44, 1936, pp. 193-210.

27—Claus, W. and Goederitz, A H. F., *"Gegossene Metalle und Legierungen,"* Berlin, M. Krayn Tech. Verlag, 1933, 345 p. See p. 1.

28—Wilkins, R. A., *"Herculoy, A Copper-silicon-tin-zinc Alloy,"* METALS AND ALLOYS, vol. 4, 1933, pp. 123-126.

29—Deitz, Jr., L. S. and Weiser, H. H., *"Metallic Alloy,"* U. S. Patent 2,062,448, December 1, 1936.

30—Arnott, J., *"Silicon as an Alloying Element,"* FOUNDRY TRADE JOUR., vol. 44, 1931, pp. 397-398.

31—Roast, H. J., *"Problems in Bronze,"* TRANS., A.F.A., vol. 45, 1937, pp. 399-422.

32—Mousson, J. M., *"Pitting Resistance of Metals Under Cavitation Conditions,"* TRANS., AM. SOC. MECH. ENGRS., vol. 59, 1937, pp. 399-408.

33—Kerr, S. L., *"Determination of the Relative Resistance to Cavitation Erosion by the Vibratory Method,"* TRANS., AM. SOC. MECH. ENGRS., vol. 59, 1937, pp. 373-397.

34—Brice, L. J., *"Some Properties of Silicon-aluminum-bronzes,"* J. INST. METALS, vol. 45, 1931, pp. 209-228.

35—Kinzel, A. B., *"Silicon: Its Applications in Modern Metallurgy,"* MIN. & MET., vol. 14, 1933, pp. 489-492.

36—Patch, N. K. B., *"Lead in Silicon Bronze May Cause Trouble,"* THE FOUNDRY, vol. 65, January, 1937, pp. 34, 78.

37—Rowe, F. W., *"Soundness in Castings,"* METAL IND. (LOND.), vol. 52, January 7, 1938, pp. 33-35.

38—Cowles, E. H., *"Aluminum Bronze and Brass as Suitable Materials for Propellers,"* TRANS., AM. INST. MIN. MET. ENGRS., vol. 18, 1890, pp. 484-496.

39—Herman, S. F. and Sisco, F. T., *"Effect of Certain Alloying Elements on Structure and Hardness of Aluminum Bronze,"* AM. INST. MIN. MET. ENGRS., Tech. Pub. 365, 1930, 23. p.

40—Corson, M. G., *"Copper Alloy Systems With Variable Alpha Range and Their Use in the Hardening of Copper,"* TRANS., AM. INST. MIN. MET. ENGRS., Proc., Inst. Metals Division, 1927, pp. 435-450.

41—Gonser, B. W. and van Wert, L. R., *"The Age-Hardening Characteristics of Some Copper-nickel-silicon Alloys,"* METALS AND ALLOYS, vol. 5, 1934, pp. 251-255, 281-283. Also: *"Certain Age-hardenable Copper Alloys,"* METALS AND ALLOYS, vol. 7, 1936, pp. 269-270.

42—Kihlgren, T. E., *"Production of Pressure Tight Castings in 30 Per Cent Cupro-nickel,"* TRANS. A.F.A., vol. 45, 1937, pp. 225-260.

43—Bolton, J. W. and Weigand, S. A., *"Method of Making Alloys,"* U. S. Patent 2,061,648, November 24, 1936.

DISCUSSION

Presiding: H. M. ST. JOHN, Detroit Lubricator Co., Detroit.

NOTE: *In the absence of Dr. Gillett, this paper was presented by Dr. C. H. Lorig of Battelle Memorial Institute.*

CHAIRMAN ST. JOHN: In opening this paper for discussion, I would like to call attention to one implication which possibly was not intended. In the abstract, the first sentence reads, "Silicon plays an important role in non-ferrous castings, but because its effects were not clearly understood, it was looked upon as a poisonous element by non-ferrous foundrymen for some time." The inference from that might be that, properly handled, it is not a poisonous element. The author, I do not believe, contended that. In paragraph 53, the author points out that in the presence of much lead a mere trace of silicon will cause trouble, and in the presence of much silicon a mere trace of lead will cause trouble. So, under certain conditions in the copper alloys, silicon is definitely very poisonous.

HAROLD J. ROAST[1]: I would rather confine my remarks to the things I know a little more about and perhaps that isn't a bad idea. And that is the foundry practice in connection with the copper-silicon-iron series, to which reference is made. I think perhaps that in the summary of the paper, as well as in the paper itself, a new stress was placed upon the idea that the shrinkage of these alloys was somewhat comparable to steel and manganese and aluminum bronze. That has not been our experience in Canada. On the other hand, I believe the statement that some better provision has to be made than one would use for 85-5-5-5. We have made quite a reasonable quantity of this material and it is gaining in popularity for hydraulic work, turbine runners, water pumps, and so forth.

Reference was made to one of these castings in yesterday morning's paper and a picture was actually shown of it, and that casting is in use and in satisfactory use.

We get three-month reports on it and we are waiting to find out when cavitation will set it. The particular benefit of the copper-silicon-iron alloy is claimed to be its resistance to cavitation, which is one of the rather elusive problems in connection with hydraulic work.

I do agree with the statement with regard to the undesirability of lead, but I understand that some work has been done on up to perhaps 0.2, 0.3 per cent, which rather indicates that the benefits obtained in machining quality outweigh the difficulty of the presence of that much lead. I know from experience that more lead certainly will give a great deal of trouble.

In regard to the question of lead in the nickel-silicon, et cetera, group, I find myself in the position of having to admit that one can make, for example, piston rings with something like 5 per cent nickel

[1] Vice President, Canadian Bronze Co., Montreal, P. Q.

and 5 per cent lead and 1 per cent silicon, a very sad alloy when you look upon the results in castings, full of worm holes and every indication of silicide of lead, but when you machine all that off you do get a casting that at all events some customers like. We must not therefore, assume it is impossible to make an alloy with lead and silicon with reasonably large proportions of both for some specific use. If you are running a foundry with leaded bronze and silicon-bronze side by side, it is rather a risky undertaking. It is very desirable to keep these two alloys apart.

W. B. Goudielock[2]: Like Mr. Roast, my experience has been with the copper, silicon and iron alloys. Dr. Gillett mentioned the physical properties that were quoted of 45,000 lb. per sq. in. tensile strength. That was taken from the Navy Spec. 46B28 covering the use of silicon bronzes. Those are the lowest values permissible and are really not an indication of what one obtains. As a matter of fact, I recently saw actual castings of copper silicon iron alloy PMG metal weighing as much as 10,000 lb. to which test pieces were attached, which tested to 57,000 lb. per sq. in. tensile strength and 40 per cent elongation in 2".

Mr. Roast mentioned the fact that lead has been alloyed with silicon bronzes. We have actually introduced lead in sufficient quantities to give free-cutting properties in rod form. We still have difficulty with appreciable lead contamination in castings, but it is possible to obtain castings with that amount which machine well.

The only other point I would like to mention is in connection with the use of high silicon pig-iron in the manufacture of these alloys. If you use high silicon pig-iron or ferrosilicon, very unsatisfactory castings result. The only way to introduce silicon and iron into copper to manufacture sound castings is by the use of a pre-alloy or hardner which is covered by patents.

Chairman St. John: The possibility was mentioned of using a small percentage of lead where there is a fairly high percentage of silicon. I would like to ask Mr. Roast if the same thing could be done with, say 8 or 10 per cent of lead and 0.2 per cent or 0.3 per cent of silicon?

Mr. Roast: Is it to be supposed that it is a lead-bronze?

Chairman St. John: The ordinary run of copper-tin-zinc-lead mixtures with as much as 10 per cent of lead. Do you think it is possible to make a casting with 0.2 per cent or 0.3 per cent of silicon?

Mr. Roast: The ordinary answer to that, in my opinion, is no, because you get so much trouble that develops in the castings under ordinary circumstances. The instance I gave when on my feet before indicates that for some special reasons it makes a very rotten casting, with no kind of surface at all, and even subject to inter-crystalline shrinkage in the center, but you sacrifice all that sort of thing for a wearing surface. But in general, no.

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[2] Metallurgist, Phelps-Dodge Copper Products Co., New York, N. Y.

L. W. KEMPF[3]: Dr. Gillett refers to the very interesting 13 per cent or 14 per cent silicon alloys, mentioning their desirable combination of tensile properties. To anyone engaged in making relatively high strength aluminum castings, a combination of 27,000 lb. per sq. in. tensile strength with an elongation of 7 per cent or 8 per cent is quite interesting. Those experienced in the field might very well ask why there isn't some commercial use of this alloy. As far as I know, last year, and even the year before that, and the year before that, there wasn't a ton of this material produced in the form of castings in this country. The reason is that the commercial production of castings of this fascinating alloy with such properties is a horse of an entirely different color. As Dr. Gillett has stated, the properties are the result of a fine structure obtained by the addition of sodium to the molten alloy just before casting. Sodium is a very volatile and combustible metal at the melting point of aluminum, and in order to obtain the maximum properties, a certain, specific concentration of sodium is necessary. If there is more or less than this optimum concentration, properties suffer quite decidedly. It thus becomes quite a neat job after adding an excess of sodium to pour the casting at exactly the moment the proper sodium concentration is reached. If you are clever enough to do this, the resulting alloy is good. It has an excellent combination of properties. But most foundrymen haven't been able to do the trick consistently, ourselves included.

GEO. P. HALLIWELL[4] (written discussion): Dr. Gillett has presented us with an excellent correlated abstract and bibliography on the role of silicon in non-ferrous alloys. In such a paper it is obviously impossible to elaborate on all the various effects produced by silicon on the properties of even a few of our more common alloys. There is one alloy, however, upon which I should like to comment, and that is Manganese bronze or more strictly speaking alpha-beta brass.

Frequent reference is made by foundrymen to the addition of small amounts of silicon to manganese bronze for the purpose of strengthening the alloy. The amounts added vary from a few hundredths to one per cent, although it is probable that only from one-half to two-thirds of this amount is actually retained.

In looking up the literature on this subject, it is surprising to find that there is a dearth of reliable and quantitative data. In the work referred to by Dr. Gillett, that of Gould and Ray, 1-3.5 per cent of silicon was added to an alpha-beta brass, which resulted in a change of structure from the alpha-beta complex to pure gamma. This is equivalent to a change in apparent zinc content from 42 per cent to at least 59 per cent. These authors show that the progressive change in structure is accompanied by an increase of 10,000 lb. per sq. in. tensile (1.5 per cent Si) and about 4 per cent in elongation (1 per cent Si).

Recent data obtained in our laboratory do not confirm these results.

Whereas these authors were using a ternary alloy, we were using one that was more complex as far as composition was concerned, but essentially the same with respect to alpha-beta structure. The composition and properties of our alloys are shown in Table 1 of this discussion.

Table 1

ALLOY	No. 1	No. 2	No. 3	No. 4	No. 5
Copper (Per Cent)	58.08	57.84	58.08	58.18	58.27
Iron (Per Cent)	1.37	1.40	1.18	1.00	0.76
Aluminum (Per Cent)	0.76	0.78	0.77	0.85	0.71
Manganese (Per Cent)	0.28	0.27	0.29	0.30	0.31
Silicon (Per Cent)	0.008	0.05	0.09	0.29	0.48
INGOT					
Tensile (lb. Per Sq. in.)	72875	76000	76000	78000	90000
Elongation (Per Cent)	35	37.8	34.5	22.5	11.5
SAND CAST					
Tensile (lb. Per Sq. in.)	71600	74600	75500	73000	82100
Elongation (Per Cent)	39	35	34	20	9
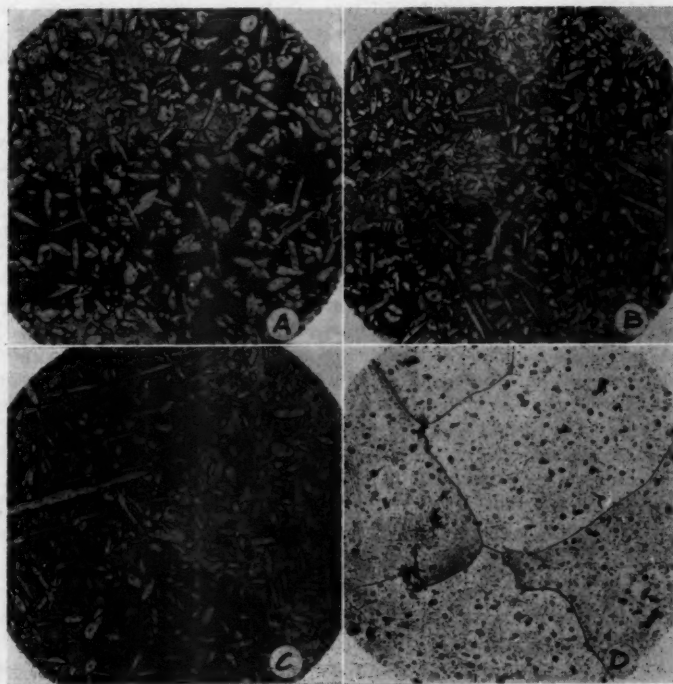
FIG. 5 (A) SI. 0.008 PER CENT. (B) SI. 0.09 PER CENT, (C) SI. 0.29 PER CENT, (D) SI. 0.48 PER CENT.

These results show a greater increase in tensile strength per unit
of silicon and a decrease rather than increase in the elongation.

The structural effects produced are shown in the photomicrographs
of Fig. 5.

M. G. CORSON *(written discussion):* There are a lot of things I
should like to say on the subject of Dr. Gillett's paper. I realize that
he did not intend to give more than a condensed review of the literature
of the subject and can not be blamed for taking all printed information
at its face value. I, personally, should prefer to see Dr. Gillett's own
and documented conception of the technological and economical value of
various silicon containing alloys.

I should like to hear, for instance, what Dr. Gillett thinks of the
modified aluminum silicon alloys. I sent out recently a number of in-
quiries concerning aluminum base alloys and only two foundries out of
twelve mentioned the modified alloys among their compositions. Of
course, the Aluminum Co. of America makes ingots for the modification
process, but is anyone actually using them? Yes, a test bar of the
modified alloy may show 30,000 lbs. per sq. in. ultimate strength, but
will an actual casting possess this strength too? From what I have
seen, I conclude that the modified alloy would have had a strength of
40,000 lbs. per sq. in. were it not usually hopelessly porous.

What interests me most are Dr. Gillett's statements on copper-
nickel-silicon alloys. In Paragraph 88 he says: "Naturally the corrosion
resistance of such an alloy would be very high, the casting properties
good, ... "

Permit me, who developed the Cu-Ni-Si alloys in a pioneering
research, to state, that I never claimed any higher corrosion resistance
for the precipitation hardened Cu-Ni-Si alloys. I, like anyone else who
enjoys a reasonably good metallurgical education, know that no preci-
pitation hardened alloy can approach the same corrosion resistance as is
proper to its matrix. No variant of heat treated steel is as corrosion
resistant as pure iron, no variant of duraluminum is as corrosion
resistant as pure aluminum (commercially pure) and no hardened copper
alloy is as corrosion resistant as pure—deoxidized copper. The best a
metallurgist can expect from a precipitation hardened alloy is that its
resistance to corrosion will not be pitifully low, and if his particular
alloy is 80 per cent as resistant to a given medium as its matrix, he
better offer a prayer of thanksgiving to the Lord, lest his resistance
drops to 50 per cent.

Dr. Gillett cites next the experience of the Ohio Brass company
with an alloy containing 15 per cent zinc, 4 per cent nickel, 1.5 per cent
silicon, balance copper. He tells us that the alloy failed to stand up to
expectations and he says that "it is not known what actually lay
beneath the difficulty."

³ New York, N. Y.

I can not help but to offer a severe criticism of the last statement and of the procedure adopted by the Ohio Brass company. In my paper given before the A.I.M.E. meeting in 1927, I have shown that copper will dissolve 5 per cent nickel silicide or 4 per cent nickel plus 1 per cent silicon at the temperature of 900°C. This—if the alloy were previously thoroughly refined by hot rolling.

At 800°C. even straight copper would not dissolve more than 3.5 per cent nickel silicide. However, Mr. Wolf, of the Ohio Brass company, tried to force a 15 per cent zinc brass to dissolve 5 per cent nickel silicide plus 0.5 per cent of excessive silicon. Now it is known that an aluminum bronze may take 9 per cent aluminum in solid solution and a tin bronze will take 14 per cent tin in solid solution. But what metallurgist would try and make a casting containing 15 per cent zinc and 9 per cent aluminum or 15 per cent zinc and 14 per cent tin?

The same applies to nickel silicide. A 15 per cent zinc brass might dissolve at 800°C. about 2.0 per cent nickel silicide, but not a bit more. The excess has to be rejected to the grain boundary forming a nice shell utterly devoid of ductility. That is why Mr. Wolf got such poor results.

Had he taken 1.5 per cent nickel and 0.35 per cent silicon, his alloy would show an entirely different behavior. He could guarantee 160 Brinell (perhaps 180) and a tensile strength of 75,000 lbs. plus 25 per cent elongation. And those would be quite conservatively stated values. I, personally, would not brag about them, but see to it that by judicial heat treatment my castings would show 85,000 lbs. ultimate strength and at least 50 ft. lb. Izod. Even these figures I would not consider as being the best possible.

So there is nothing mysterious about Mr. Wolf's disappointment. True enough, butter does not spoil porridge, but nickel silicide is not butter—even in relation to copper.

H. W. GILLETT *(written closure):* The discussion has thrown light on several points that were omitted or insufficiently emphasized in the paper. Kempf has made it clear why the modified Al-Si alloys are not used. Corson's explanation of the trouble Wolf met is probably correct, though I do not recall that Corson or anyone else had made clear just what reduction in solubility of nickel silicide would be brought about by zinc. Halliwell's extremely interesting discussion of the effects of small variations of Si content in manganese bronze, compared with the results of Gould and Ray, indicates that a specific examination of the case at hand may be necessary, to supplement the intuition of people with a "reasonably good metallurgical education." One might wish that Corson had explained why the straight Tempaloy, with which, in the laboratory, he can get 75,000 and 25 per cent, is, from the commercial point of view, almost as much of a curiosity as is a modified Al-Si alloy.

It is possible that when the alloying elements are reduced to the point where the nickel silicide behaves itself, the shrinkage troubles make handling the alloy too much of a nuisance. As stated in the text, these alloys were carefully studied for a long time by people who were, at

the outset, as enthusiastic as Corson still is, but the results were not published. An unsuccessful attempt was made to get some of this information released as discussion of this paper.

I quite agree with Mr. Roast that the Si alloys are not quite as mean in shrinkage as manganese bronze. Since he agrees that they are worse than 85-5-5-5 in this respect, we are stating about the same thing in different words. Since some might read the paper who have not had personal experience with the necessary risering, I still maintain that paragraph 80 is a proper statement of the situation.

Factors Influencing Annealing Malleable Iron

By A. L. Boegehold,* Detroit, Mich.

Abstract

Experiments are described in connection with the determination of the shortest practical annealing cycle for two production castings representing the largest and smallest casting in current production at the time experiments were made. The experiments include a determination of the best rate of heating to the primary graphitization temperature, the length of time required to complete primary graphitization as affected by rate of heating to that temperature, effect of rate of cooling from the primary graphitization temperature to 1400°F., and the effect of rate of cooling from 1400°F to the Ar_1 metastable. The information obtained is used in arriving at the location of the temperature range of heating for the simultaneous existence of alpha iron, gamma iron and iron carbide and also the location on cooling of the Ar_1 metastable.

1.   An investigation conducted by the author a number of years ago to determine the shortest time cycle for annealing iron made by the Saginaw Malleable Iron Division of General Motors Corporation, disclosed, among other things, that the teaching of the literature in some respects did not apply to the iron being investigated.

2.   Several investigators have contributed important findings regarding detail effects associated with annealing white iron. These have been considered and compared with results obtained with the iron under consideration in this paper.

Type of Iron Studied

3.   The iron we studied differs somewhat in composition and method of manufacture from much of the malleable iron produced

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today in this country. Its manufacture has been described by Mayne[1]. Briefly, steel scrap, silvery pig iron and foundry returns are melted in a cupola producing an iron containing 2.70-2.75 per cent carbon, 1.15-1.20 per cent silicon, 0.38-0.42 per cent manganese, about 0.05 per cent phosphorus, and 0.09-0.11 per cent sulphur. This iron is transferred to a direct arc electric furnace and heated to casting temperature and 3 to 4 per cent steel scrap added. Aside from the melting method which differs from that commonly used in the manufacture of malleable iron, the composition is somewhat higher in carbon and silicon, lower in phosphorus and higher in sulphur than iron made by other manufacturers of this article. It will be appreciated, therefore, that the findings of other investigators would not agree necessarily with results obtained with this iron.

FINDINGS OF SOME PREVIOUS INVESTIGATIONS

4. W. R. Bean and co-authors[2] showed that temper carbon size and ferrite grain size increased with size of section and Phillips and Davenport[3] showed that the austenite-cementite distribution in white iron increased in grain size with increasing section size, corresponding to the increase in temper carbon and ferrite grain size noted by Bean. It will be realized from the evidence in this paper that the relation between austenite-cementite grain size and temper carbon size, disclosed by Phillips and Davenport, was obtained purely because of the particular laboratory annealing cycle used and that it is not necessary to submit to the presence of large temper carbon spots even though the white iron freezes with coarse austenite-cementite grain size. The nature of experiments reported in the literature indicate a general absence of realization that section size and its consequent effect upon microstructure had any influence upon rate of annealing.

5. D. P. Forbes in discussing work of Schneidewind and White[4] stated that large castings take longer to anneal than small ones. A quantitative evaluation of this difference however does not seem to be reported. In all the literature examined by the author, annealing experiments have been conducted on small test pieces such as $\frac{5}{8}$-in. round tensile test bars, or similar specimens of small size as cast. Several years prior to the experiments described in this paper, we had found that iron having

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1 Superior Numbers refer to corresponding bibliography references as appended to paper.

large, widely-separated, temper-carbon spots required much longer
for first stage graphitization than iron in which the temper
carbon spots were close together[5, 6].

HEAVY PRODUCTION CASTING STUDIED

6.   Since slowly solidified castings may develop large, wide-
ly-separated, temper-carbon spots during annealing and since an-
nealing time increases because of greater distance between
temper-carbon spots, it was decided that we could not expect
experiments on $\frac{5}{8}$-in. round test bars, which solidify with compara-
tive rapidity, to indicate what to expect from castings weighing up
to 50 lb. and which have sections which solidify much more slowly
than a $\frac{5}{8}$-in. round.   On this account, our plan provided that an-
nealing tests would be made on samples cut from a heavy produc-
tion casting.

7.   The casting selected was a brake flange hub shown in
Fig. 1.   This casting was usually the first to show mottle if the iron
changed slightly toward the soft side, indicating that its rate of
solidification was slowest of the castings then in production.
Samples were cut from this casting at three locations marked A,
B and C in Fig. 1.   It was presumed that these three sections would
represent three different solidification rates, the fastest of which
would be considerably slower than a $\frac{5}{8}$-in. round test bar.
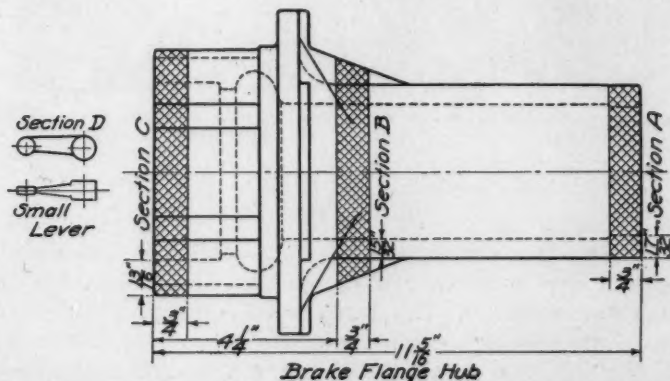
FIG. 1—SKETCHES OF CASTINGS FOR ANNEALING TESTS—THE CROSS-HATCHED AREAS OF
THE BRAKE FLANGE HUB REPRESENT RINGS MACHINED FROM CASTING FROM WHICH
SPECIMENS WERE CAST FOR ANNEALING TESTS.
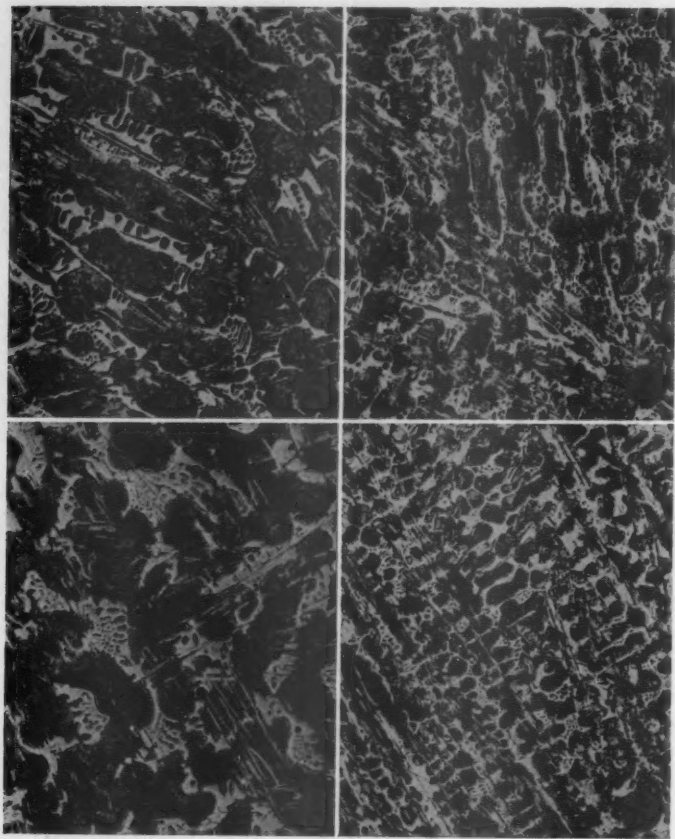
FIG. 2—MICROSTRUCTURE IN AS-CAST CONDITION OF SECTIONS *A*, *B* AND *C* FROM BRAKE
FLANGE HUB CASTING AND SECTION *D* FROM SMALL LEVER CASTING—100X.

SMALL CASTING

8.   To study also the effect of rapid solidification rate, anneal-
ing tests were made on a small casting also shown in Fig. 1,
designated section *D*. This small lever measured about ⅜-in.
through the heaviest section. The micrographs in Fig. 2 show the
"as-cast" structure found in sections *A*, *B*, *C* and *D*. As described
by Phillips and Davenport, these structures reflect the effect of
solidification rate.

Investigation of Primary Graphitization

9. The primary graphitization part of the annealing process, consisting of holding at 1700°F., was naturally the first phase to be investigated. Commercial annealing frequently is done at 1725 to 1750°F. but higher temperatures usually are avoided because of objectionably fast furnace deterioration and warpage of castings. It was decided that tests would be made at 1700°F. as a standard so that there would be some margin of temperature remaining to take advantage of in production to obtain complete annealing of iron on those days when melting conditions produced an iron somewhat more difficult to anneal than that in this investigation. In this way, we would be assured fairly well that any cycle worked out in the Laboratory would produce completely annealed iron in production.

10. The first question to be settled before starting the primary graphitization was: How fast should the hard iron be heated to 1700°F.? At the time this work was done in 1934, our own experience[5, 6] was contrary to the literature we examined on the subject. Hayes and co-authors[7] said that "The ideal method would have been to introduce the test specimens into the furnace while at temperature—but the specimens were charged in a cold furnace and brought to 1832°F. in 2 hr. after passing the critical temperature." If we assume the critical temperature at 1425°F., as given by the authors, then the heating rate was about 200°F. per hr. Hayes and co-authors[8] also stated: "When heating to temperatures of 1700 to 1750°F., there is no difference in the temper carbon form whether heated to that temperature in 3 hr. or in 7 min." White and Archer[9], using 3/8-in. sections for annealing experiments, stated: "It is evident that the rates of heating and cooling are without any metallurgical effect and depend only upon the practical considerations of the annealing room."

Findings Compared With Those of Other Investigators

11. It is not intended to contradict the conclusions of the authors cited above because we know that, with iron melted under certain conditions, their conclusions are correct, especially referred to light sections. We found, however, that there are irons to which these conclusions do not apply. For example, in our reports[5, 6] cited previously, we described how 5/8-in. diameter bars required 26 hr. to complete first stage graphitization when placed in a furnace already at 1700°F., whereas similar samples of the same iron

heated from 1550 to 1600°F. at 30°F. per hr. and from 1600 to 1700°F. at 72°F. per hr. showed no massive combined carbon after 18 hr. at 1700°F.

12. No experiments were made to determine the minimum time required to eliminate the massive carbides after slow heating with this cycle but, in the light of later experiments, it would have been considerably less than 18 hr. The temper-carbon spots obtained when the iron was placed in a furnace at 1700°F. were so large that at 100 diameters, they measured as much as 3/4-in. in diameter and only one or two appeared on a 2¾x3¾-in. micrograph. When heated slowly, the temper-carbon spots were much smaller and closer together. Fig. 3 shows the contrast between these structures obtained in sections 5/8-in. round as-cast. The influence of distance between temper-carbon spots upon rate of annealing also was discovered by Schneidewind and White[10].

13. After the completion of our work, we came across a reference by Kluytman[11] who, in 1926, stated that rapid heating to annealing temperature is detrimental to the properties of malleable iron as it produces coarse graphite. C. S. Smith and K. Palmer[12] also discussed this phenomenon, although their slowest heating rate was 11.3 hr. to reach 1697°F. at constant rate or approximately 150°F. per hr.

CAUSE OF FORMATION OF LARGE TEMPER-CARBON SPOTS

14. Although our past experience was that rapid heating caused formation of large temper-carbon spots, it was thought advisable to determine this point again with the iron in current production. Therefore, the first cycle consisted of heating to 1700°F. at 228°F. per hr. The specimens were placed inside a heat resisting annealing box with a thermocouple placed with the junction touching the specimens. Another thermocouple, outside the box, was used to control the furnace temperature. The fluctuations in temperature of the control couple were almost entirely absent from the thermocouple inside the box showing the temperature of the samples being annealed.

15. Samples from sections A, B, C and D were pulled after 8 hr. at 1700°F. The microstructure of these samples is shown in Fig. 4. The contrast in size of temper-carbon and progress of carbide elimination is strikingly proportional to section size or to solidification rate. The smallest section shows the smallest in size
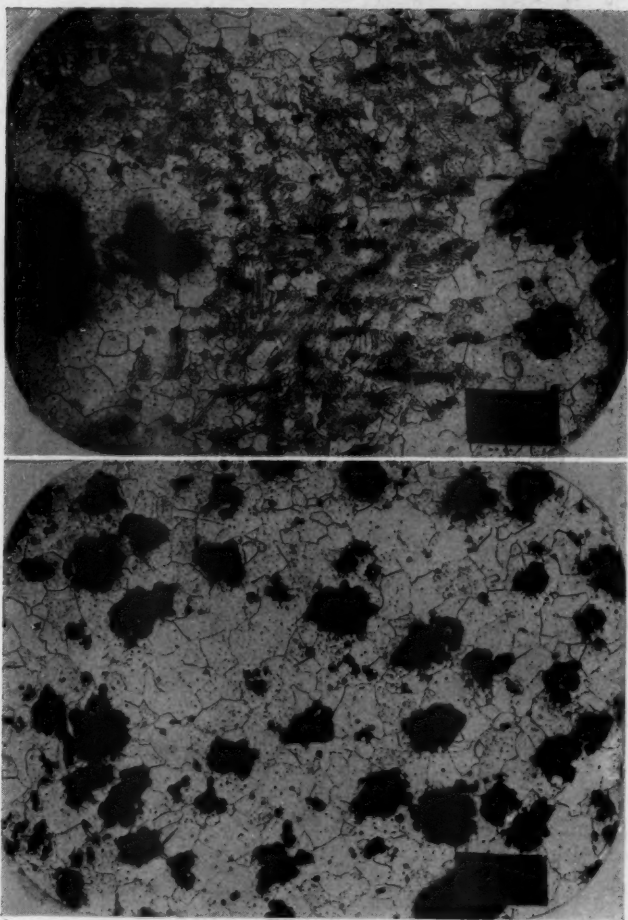
Fig. 3—Initial Heating Rate Affects Temper Carbon—Structures *A* and *B* Obtained in the Same Iron in ⅝-in. Round Test Bars Merely by a Difference in Rate of Heating to 1725°F. in Annealing Cycle—100x.

A—Placed in Furnace at 1725°F., Held 23 Hr., Cooled to 1400°F. at 56° per Hr., Cooled to 1325°F. at 5.35° per Hr. and from 1325 to 1250°F. at 9° per Hr.

B—Heated to 1725°F. at About 80° per Hr., Held 19 Hr., Cooled to 1475°F. at About 85° per Hr., 1475 to 1360°F. at 18.3° per Hr. and 1360°F. to 1335°F. to 2.5° per Hr.
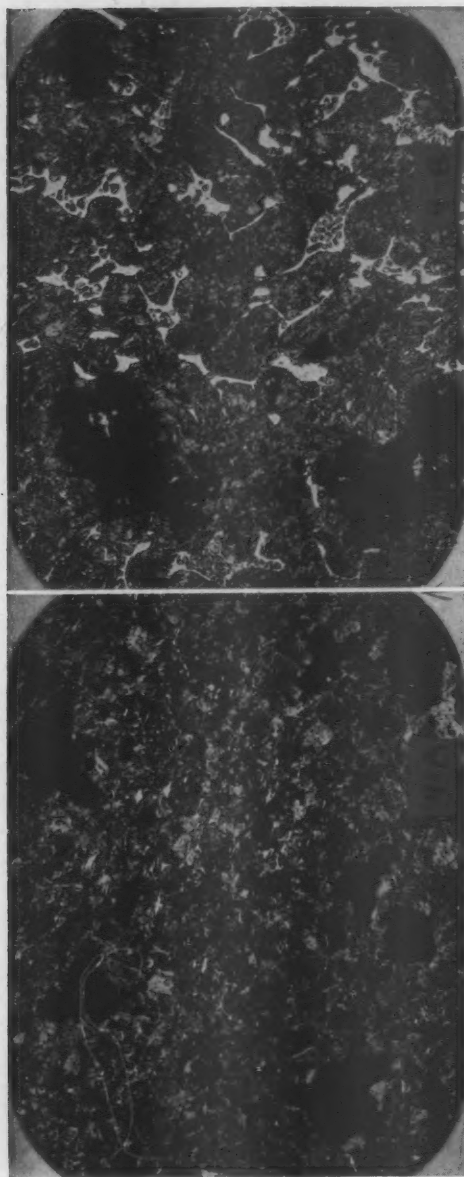
FIG. 4A.—EFFECT OF SECTION SIZE ON TEMPER CARBON SIZE WHEN HEATING TO 1700°F. IN RAPID.—SECTIONS A AND B AFTER HEATING TO 1700°F. AT 228°F. PER HR. AND HOLDING 8 HR. AT 1700°F.—100X.
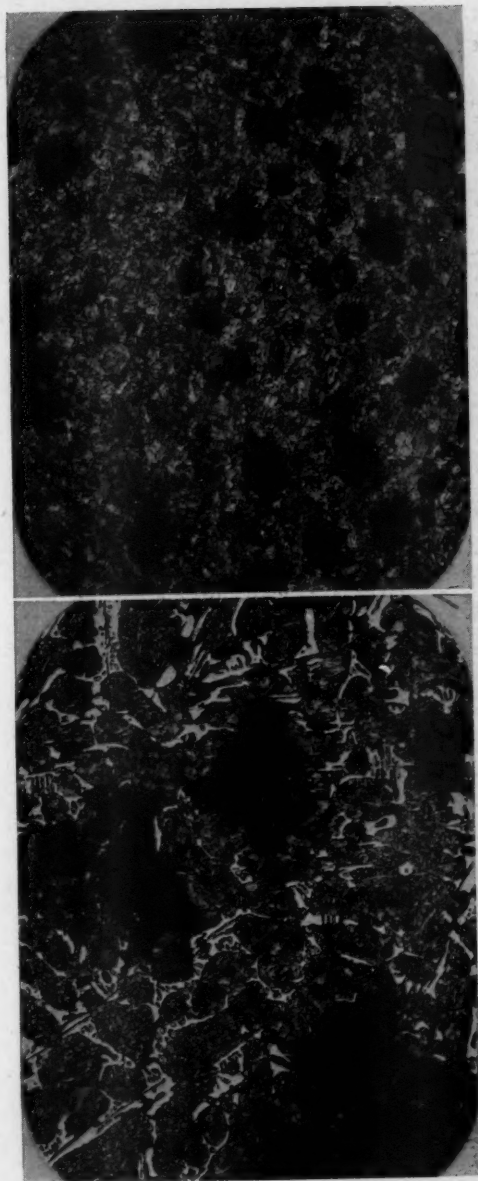
FIG. 4B—SAME AS FIG. 4A BUT SHOWING SECTIONS C AND D WITH SAME TREATMENT AS SECTION A AND B.
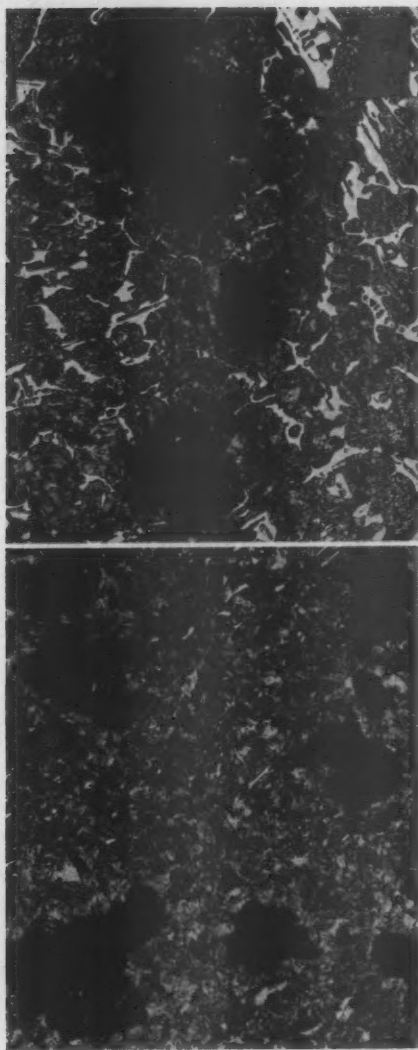
Fig. 5A—Effect of Section Size and Heating Rate on Temper Carbon Size.—Photomicrographs 1 and 2 are of Sections A and C Placed Into Furnace at 1700°F. and Held 7 Hr.
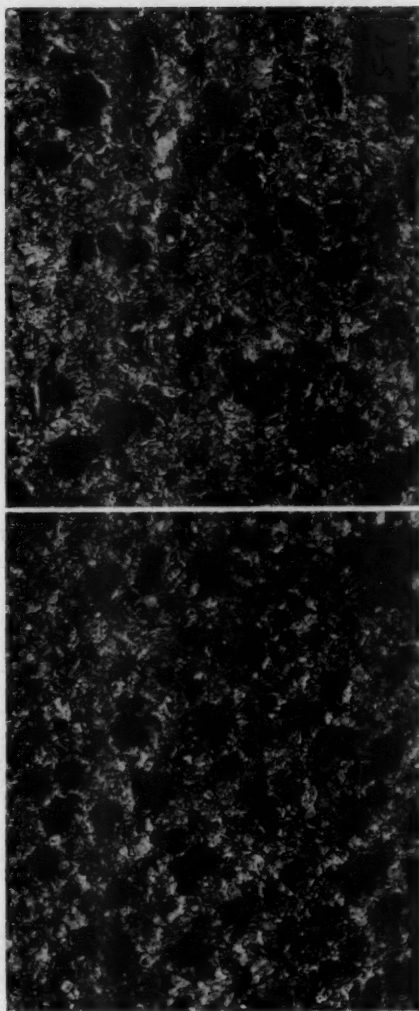
FIG. 5B—EFFECT OF SECTION SIZE AND HEATING RATE ON TEMPER CARBON SIZE.—PHOTOMICROGRAPHS 3 AND 4 ARE OF SECTIONS A AND C HEATED TO 1700°F. AT 62.5° PER HR., AND HELD 8 HR.

19. It is obviously a waste of time to keep the heating rate all the way from room temperature to 1700°F. as slow as 62.5°F.

16. Also, as previously noted, the progress of annealing is more rapid with decrease in distance between nodules so that the first stage of graphitization in section D is complete in less than 8 hr. and, as will be seen further on, was complete in 4 hr. Section A, although from a part of the casting 17/32-in. thick was at the end of the tubular section and therefore next to D, in order of solidification rate had almost completed the first stage in 8 hr. Section B, 15/32-in. thick, but adjacent to a heavy flange and obviously slower in solidification than section A, showed considerable massive cementite after 8 hr. and section C, 13/16-in. thick, showed large quantities of massive cementite. Section A had completed first stage graphitization in 10 hr. Section B was complete in 21 hr. and section C was almost complete in 23 hr. These results then agree with our previous experience, that fast heating produces coarse temper-carbon spots and also agree with previously cited publications by Bean and co-workers[2] and by Phillips and Davenport[3] regarding relation between section size, cast structure and temper-carbon size.

17. Samples of sections A and C were then heated to 1700°F. at 62.5°F. per hour and in 6 hr. after reaching 1700°F., section A had completed first stage graphitization and C had completed first stage graphitization in eight hr. The temper-carbon size shown in Fig. 5, after heating slowly at 62.5°F. per hour to 1700°F. is in striking contrast with temper-carbon size when heated at 228° per hr. to 1700°F. and shows no relation whatever to section size or to austenite-cementite distribution in the white iron. Therefore, we now must modify the theory formerly held, and say that temper-carbon nodule size varies directly with section size when heating to temperature for first stage graphitization is too rapid and is not appreciably affected by section size when heating to temperature for first stage graphitization has been slow enough.

18. Samples of sections A and C, from the same casting heated rapidly to 1700°F., showed the same size of temper-carbon particles as shown in Fig. 4.

Temperature At Which Carbides Dissolve

19. It is obviously a waste of time to keep the heating rate all the way from room temperature to 1700°F. as slow as 62.5°F.

per hr. Therefore, a series of samples were removed from the furnace at different temperatures on the heating part of the cycle to determine when iron carbide dissolved at the Ac_1 metastable and when temper-carbon began to appear. It is recognized that other investigators have determined these points with iron of other composition and origin but it was desired that an independent check be made on the particular iron under consideration.

20. Critical temperature curves, obtained by the temperature differential method, show a range of temperature difference. Bean[13], in discussing a paper by Schwartz, shows such a curve for a white iron containing 2.58 per cent carbon and 0.87 per cent silicon, where the curve inflects at 1405°F. and returns to normal at 1525°F. Phillips and Davenport[9] show a similar curve where the solution of carbon begins apparently at 1409°F. and continues till 1454°F.

21. The author made similar curves about 15 years ago and one, for an iron with 2.34 per cent carbon and 2.98 per cent silicon made white by casting in a chill mold, showed the beginning of the Ac range at 1500°F. and the end at 1620°F. The location of a temperature range where alpha iron, gamma iron and cementite exist simultaneously is shown in the iron-silicon diagrams by
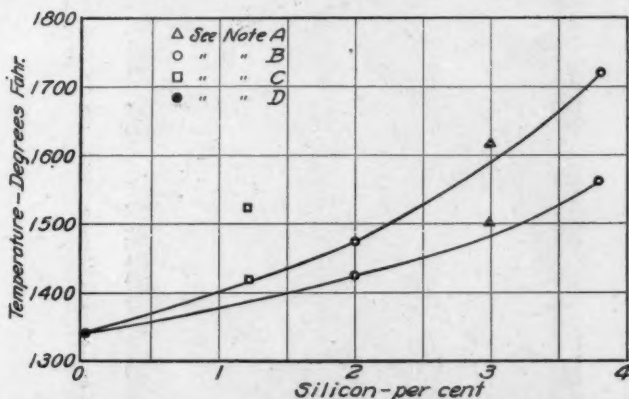
FIG. 6—RANGE FOR SIMULTANEOUS EXISTENCE OF ALPHA IRON, GAMMA IRON AND CEMENTITE.

NOTE A—CRITICAL POINTS DETERMINED BY TEMPERATURE DIFFERENTIAL METHOD ON WHITE IRON WITH 2.84 PER CENT CARBON.

NOTE B—GREINER, MARSH & STOUGHTON—ALLOYS OF IRON AND SILICON, PAGES 140 AND 141—FOR 2.75 PER CENT CARBON.

NOTE C—2.75 PER CENT CARBON, 1.20 PER CENT SILICON METALLOGRAPHIC DETERMINATION ON SAMPLES HEATED AT 100°F. PER HR.

NOTE D—A.S.M. HANDBOOK.
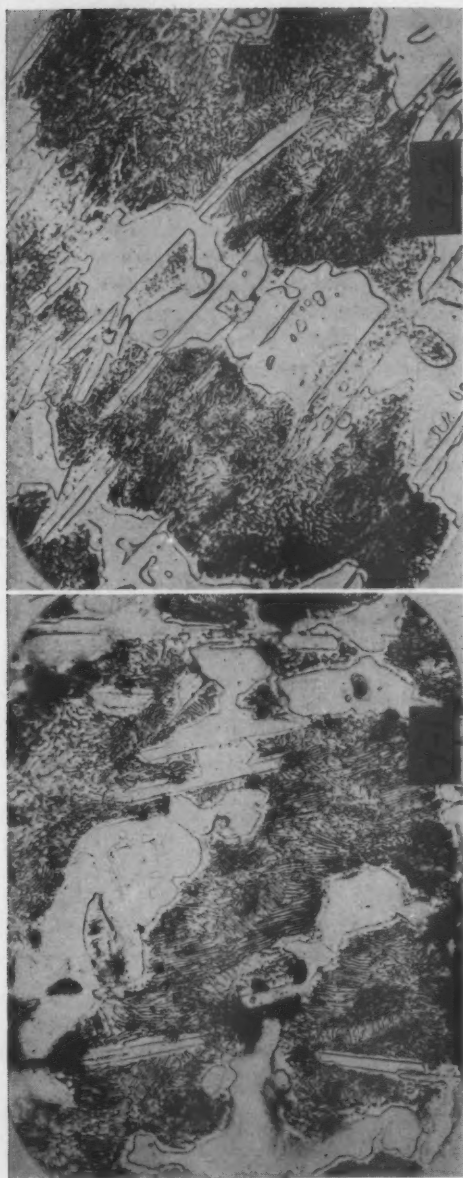
FIG. 7A—STRUCTURE OF SECTION C HEATED AT 100°F. PER HR. THROUGH CRITICAL RANGE.—(1) HEATED TO 1400F.—AIR COOLED—500X. (2) HEATED TO 1425°F.—AIR COOLED—500X.
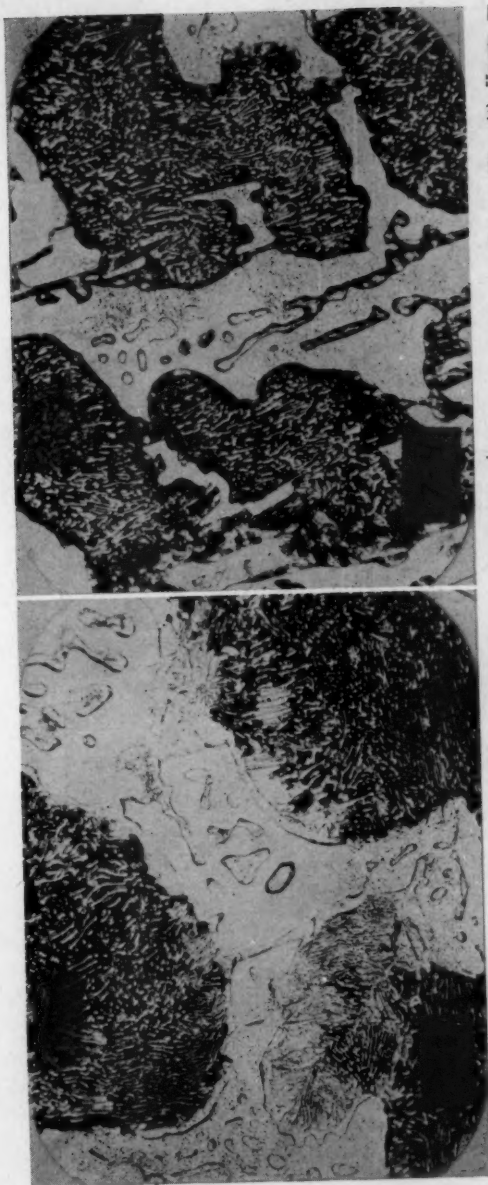
FIG. 7B—STRUCTURE OF SECTION C HEATED AT 100°F. PER HR. THROUGH CRITICAL RANGE—(3) HEATED TO 1475°F.—AIR COOLED—500X. (4) HEATED TO 1500°F.—AIR COOLED—500X.
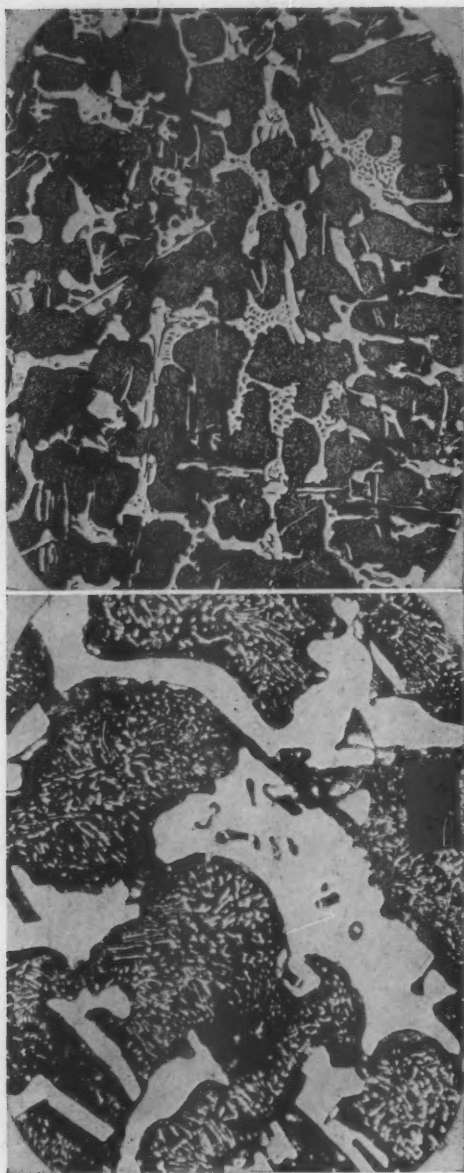
FIG. 7C—STRUCTURE OF SECTION C HEATED AT 100°F. PER HR. THROUGH CRITICAL RANGE—(5) HEATED TO 1525°F.—AIR COOLED—500X. (6) HEATED TO 1525°F.—AIR COOLED—100X.

Greiner, Marsh and Stoughton[14]. For 2 per cent silicon and 2.75 per cent carbon, this range extends from 1425°F. to 1475°F. and for 2.75 per cent carbon and 3.8 per cent silicon from 1562°F. to 1720°F.

22. By locating a point at 1337°F. for zero silicon Ac_1 and drawing lines through the points mentioned for 2.0 per cent silicon and 3.8 per cent silicon irons, as shown in Fig. 6· we find that, for an iron containing 2.75 per cent carbon and 1.20 per cent silicon such as we are dealing with, this range would be 1380°F. to 1410°F. But micrographs in Fig. 7, showing results of our effort to determine this range, indicate that there has been no solution of carbide at 1400°F. At 1425°F., there is some evidence of solution beginning. At 1475°F., solution of carbide is advanced and there is still some undissolved pearlitic carbide and some ferrite present between the austenite and the massive carbides. At 1500°F., the alpha iron has all disappeared except in spots in the eutectic structure. At 1525°F., this remaining ferrite has been obliterated.

Simultaneous Existence of Alpha Iron, Gamma Iron and Carbide

23. According to determination by the metallographic methods used, the range of temperature for simultaneous existence of alpha iron, gamma iron and carbide for the iron being examined would be 1420°F. to about 1520°F. That this does not agree with the published diagrams, we may assume to be because the published diagram represents equilibrium conditions whereas the test conditions under which this range, for our iron, was determined do not represent equilibrium conditions nor even an approach to equilibrium—so the range from 1420 to 1525°F. given for the occurrence of alpha iron, gamma iron and carbide in an iron of 1.20 per cent silicon and 2.75 per cent carbon applies only where the heating rate is about 100°F. per hr. which is a very fast heating rate compared to the time required to obtain equilibrium in an iron-carbide-silicon alloy of the composition under discussion. All this helps to illustrate why the information in the books must be supplemented before an industrial application of the information can be successful. The work must all be done over again to discover what happens under industrial conditions which are far from those of equilibrium. Knowledge of what occurs at various heating and cooling rates is necessary to solve the industrial problem.

24. The fact that carbides begin to dissolve at some spots but not in others until over 100°F. higher, may be due to microscopic
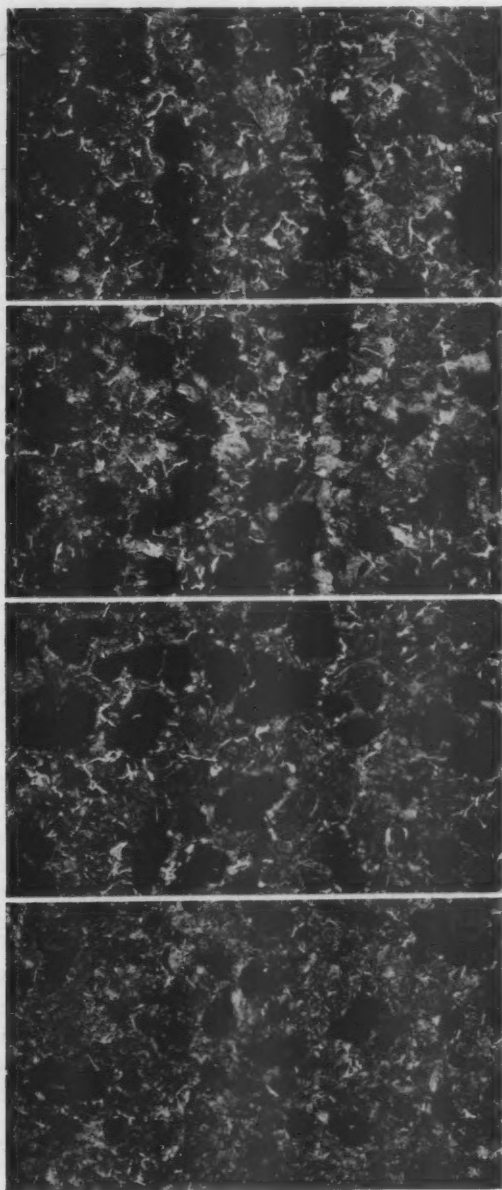
FIG. 8A—DECOMPOSITION OF PEARLITE WHILE COOLING AT 7° PER HR. FROM 1400 TO 1810°F. SECTION C—HEATED TO 1550°F. AT 100° PER HR., HELD 8 HR., HEATED TO 1700°F. AT 300° PER HR., COOLED TO 1400°F. AT 150° PER HR. AND SAMPLES REMOVED AT INTERVALS WHILE COOLING FROM 1400 TO 1810°F. AT 7° PER HR. (1) REMOVED AT 1400°F.—(2) REMOVED AT 1835°F.—(3) REMOVED AT 1877°F.—(4) REMOVED AT 1870°F.
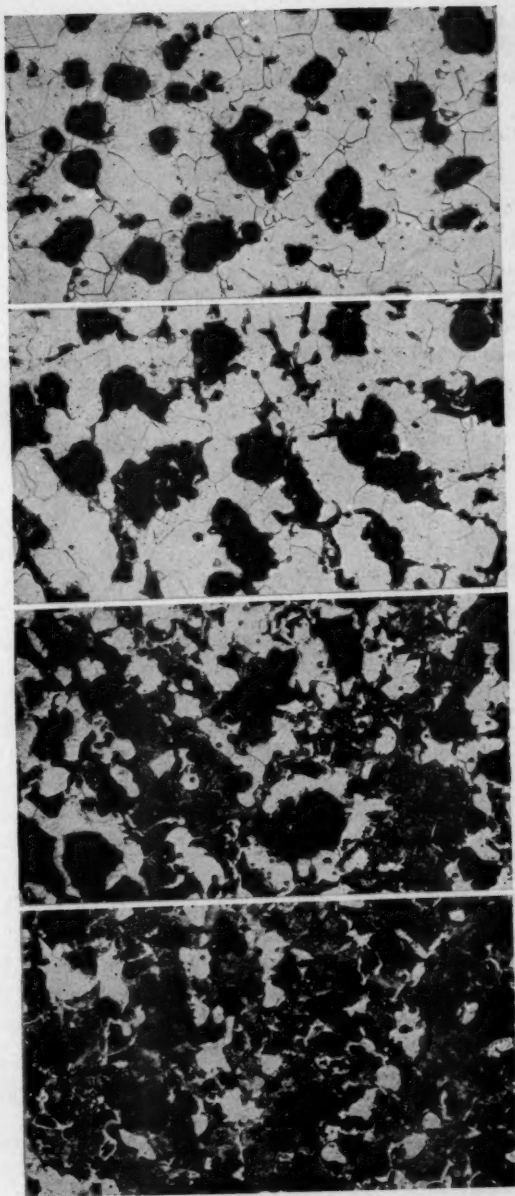
FIG. 8B—SAME AS FIG. 8A BUT (5) REMOVED AT 1860°F.—(6) REMOVED AT 1835°F.—(7) REMOVED AT 1835°F.—(8) REMOVED AT 1810°F.

segregation of silicon, manganese, phosphorus and sulphur in sufficient concentrations to alter the solubility of the carbides or the solvent action of the ferrite. The existence of such microscopic segregation was mentioned by Schwarz and Junge[15] as an explanation for structural phenomena during cooling through the critical range.

Beginning of Temper-Carbon Formation

25. The micrograph in Fig. 7 at 100 diameters of the iron at 1525°F., shows the beginning of temper-carbon formation at frequent intervals. At 1550°F., the iron was held for 3 hr. to allow the temper-carbon particles to develop further, the temperature was increased to 1700°F. and held 11 hr. to complete first stage graphitization,* cooled to 1400°F. at 150°F. per hr. and then cooled to 1310°F. at 7° per hr. During slow cooling from 1400°F. to 1310°F., samples were withdrawn at intervals to observe the structural transition.

COOLING RATES

26. Fig. 8 shows the change in microstructure during cooling from 1400°F. At 1400°F., ferrite begins to appear in very small quantities. In successive samples, withdrawn at 1385, 1377 and 1370°F., the amount of ferrite increased very slightly. At 1360°F., ferrite appeared in larger quantities in lakes as contrasted to a network or grain boundary shape in samples removed between 1400 and 1370°F. At 1335°F., the amount of ferrite has increased considerably so that the amounts of ferrite and pearlite are about equal and at 1325°F., only 10° lower, the reduction in amount of pearlite is remarkable. In this 10° interval, the amount of pearlite drops from 50 per cent to 10 per cent of the area. At 1310°F., all pearlite is eliminated.

27. Schneidewind and White[10] indicate in their Fig. 12 the existence of a sudden decrease in carbon solubility but their location of the temperature range at which this occurs for a 1.26 per cent silicon iron is higher than we found when cooling at 7° per hr. and also, contrary to our findings shows persistence of iron carbide to the extent of 0.12 per cent carbon at 1300°F.**

28.   With the information obtained from metallographic ex-

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* Subsequent experiments showed that when section C was heated from 1520 to 1700°F. at 36° per hr., it required only 5 hr. at 1700°F. to complete first stage graphitization.
** Discussion with Dr. Schneidewind on this point revealed that no pearlite was visible in the microstructure at this point but that the chemical laboratory reported 0.12 per cent combined carbon.
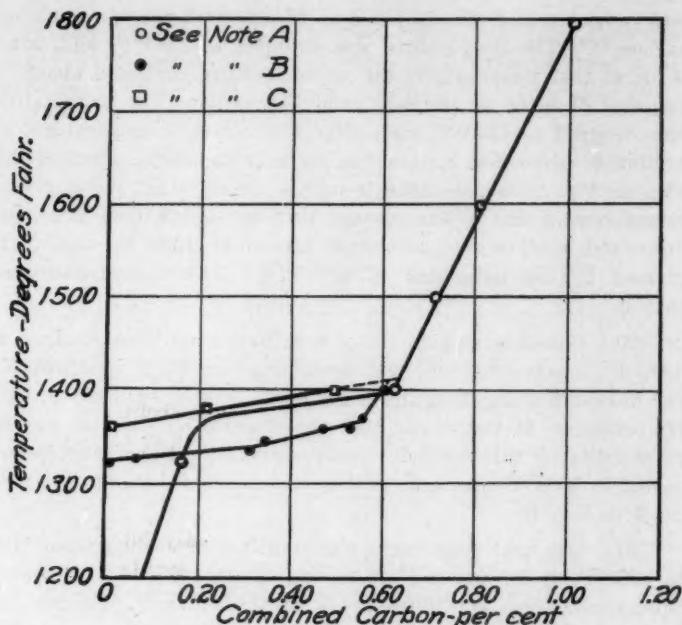
FIG. 9—COMBINED CARBON AFFECTED BY COOLING RATE

NOTE A—EQUILIBRIUM CURVE—SCHNEIDEWIND AND WHITE.

NOTE B—COMBINED CARBON WHILE COOLING AT 7° PER HR. ESTIMATED METALLOGRAPHICALLY.

NOTE C—COMBINED CARBON WHILE STEP COOLING 4 HRS. AT 1400°F., 3 HRS. AT 1380°F., AND 4 HRS. AT 1360°F.

ESTIMATED METALLOGRAPHICALLY IN PER CENT OF EUTECTOID. TO LOCATE POINTS, 0.02% CARBON, ASSUMED TO BE EUTECTOID COMPOSITION, WAS MULTIPLIED BY THE ESTIMATED PERCENTAGE OF THE TOTAL AREA OF MICROSTRUCTURE OCCUPIED BY COMBINED CARBON DISREGARDING AREA OF TEMPER CARBON SPOTS.

amination of specimens removed at various stages during cooling at 7° per hr. from 1400 to 1310°F., we would plot a solubility curve like that shown in Fig. 9, which includes Schneidewind and White's curve. These results indicate that for a cooling rate of 7° per hr., the combined carbon retained at temperatures above 1325°F. is greater than the combined carbon shown by them to be in equilibrium at these temperatures, which is as it should be but, at temperatures below 1325°F., the carbon remaining appears to be lower than they give for equilibrium conditions.

29. In another series of tests, samples of this same iron were carried through complete first stage graphitization and cooled to 1400°F. at 170° per hr. After 4 hr. at 1400°F., the microstructure

was judged to indicate 80 per cent of eutectoid percent combined carbon.*** The temperature was dropped to 1380°F. and, after 3 hr. at that temperature, the microstructure indicated about 35 per cent of eutectoid per cent combined carbon. The temperature was dropped to 1360°F. and, after 4 hr. at that temperature, no combined carbon was apparent in the microstructure. Plotting this also in Fig. 9, we see that it agrees closely with the curve of Schneidewind and White, except that the curve does not bend down but reaches zero combined carbon at 1360°F.—again explained by the difference in method of determining combined carbon.

30.   Complete elimination of pearlite having been obtained at 1360°F., it was presumed that by cooling to 1360°F. at about 20° per hr. and holding, that annealing would complete rapidly at that temperature. It was found that the pearlite elimination was no more rapid by this method. The progress of graphitization when cooled to 1360°F. and held at that temperature is shown by curve no. 4 in Fig. 10.

31.   The next experiment also consisted of cooling from 1400 to 1360°F. at 20° per hr. but instead of holding at that temperature, cooling was continued at 7° per hr. 1318°F. was reached in 8 hr. at which time annealing was complete in one sample but not in another. This method of cooling apparently borders between being slow enough and too fast for complete pearlite elimination. Curve no. 5, Fig. 10, shows the progress of annealing at each temperature. This cycle was repeated with another group of samples and 7.5 hr. after starting from 1400, 1322°F. was reached when annealing was complete. The progress of annealing is shown in curve no. 8, Fig. 10.

32.   Curve no. 5 in Fig. 10 represents the progress of annealing when cooling at 7° per hr. from 1400°F. Annealing was complete at 1315°F., 12 hr. after crossing 1400°F.

33.   Two tests were made in which cooling from 1400 to 1360°F. was at 20° per hr. and from 1360 to 1300°F. at 15° per hr. In neither case was annealing complete, there being about 20 per cent of the pearlite undecomposed. The course of pearlite decomposition is represented in curves nos. 7 and 9, Fig. 10. The microstructure of samples removed at decreasing temperatures on cool-

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*** By assuming that solubility of 0.62 per cent carbon at 1400°F. shown on Schneidewind and White's solubility curve represents the eutectoid composition, 80 per cent completion of second stage annealing would correspond to a combined carbon content of 0.8 x 0.62 = 0.496 per cent.

ing curve no. 7 show that the cooling rate of 15° per hr. causes higher percentages of pearlite to remain at all temperatures than occurs when cooling it at 7° per hr.—as would be expected.

34.   The sample removed at 1315°F. shows about 20 per cent of the ground mass composed of fine pearlite, indicating that austenite was present when the sample was removed from the furnace and air cooled. The sample removed at 1300°F. shows about the
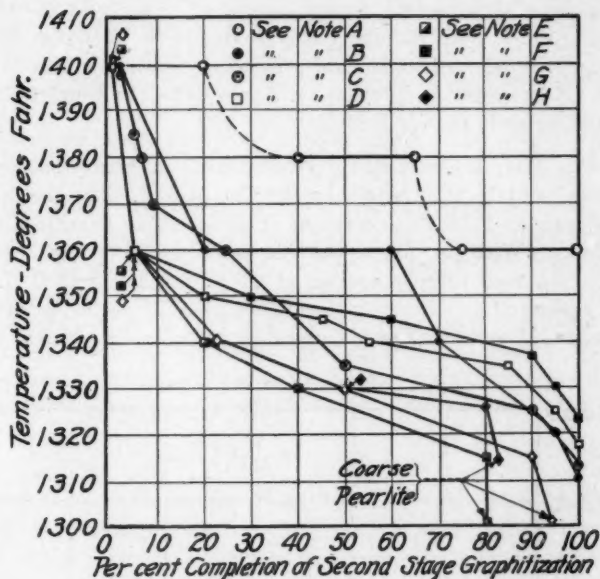
FIG. 10—COOLING CYCLE AFTER COMPLETING FIRST STAGE GRAPHITIZATION AT 1700°F.

NOTE A—CURVE No. 3—172° PER HR. FROM 1700-1400°F.—4 HR. AT 1400°F., COOL TO 1380°F.—3 HR. AT 1380°F., COOL TO 1360°F., 4 HR. AT 1360°F.

NOTE B—CURVE No. 4—150° PER HR. FROM 1700 TO 1400°F., 20° PER HR. 1400 TO 1360°F., HOLD 1360°F. FOR 7 HR., 2 HR. AT 1340°F., 1½ HR. AT 1320°F., ½ HR. AT 1310°F.

NOTE C—CURVE No. 5—150° PER HR. FROM 1700 TO 1400°F., 7° PER HR. 1400 TO 1310°F.

NOTE D—CURVE No. 6—172° PER HR. FROM 1700 TO 1400°F., 20° PER HR. FROM 1400 TO 1360°F., 7° PER HR. FROM 1360 TO 1310°F.

NOTE E—CURVE No. 7—150° PER HR. FROM 1700 TO 1400°F., 20° PER HR. FROM 1400 TO 1360°F., 15° PER HR. FROM 1360 TO 1300°F.

NOTE F—CURVE No. 8—150° PER HR. FROM 1700 TO 1400°F., 20° PER HR. FROM 1400 TO 1360°F., 7° PER HR. FROM 1360 TO 1320°F.

NOTE G—CURVE No. 9—60° PER HR. FROM 1700 TO 1400°F., 20° PER HR. FROM 1400 TO 1360°F., 15° PER HR. FROM 1360 TO 1300°F.

NOTE H—CURVE No. 10—60° PER HR. FROM 1700 TO 1400°F., 20° PER HR. FROM 1400 TO 1330°F., 3° PER HR. FROM 1330 TO 1315°F.

same proportion of the ground mass composed of coarse pearlite indicating that the Ar_1 metastable was crossed while the sample was cooling at 15° per hr. in the furnace between 1315 and 1300°F. Therefore the Ar_1 metastable was crossed somewhere between 1315 and 1300°F. So if we halve the difference, we may select 1307°F. as the Ar_1 metastable and we will be within less than 8°.

35. Microstructure of samples taken at points on curve no. 9 are similar to those on curve no. 7, except that at 1315°F. annealing is about 90 per cent complete instead of 80 per cent in the case of curve no. 7. This difference may be due to the fact that greater homogeneity of composition was attained by cooling at 60° per hr. from 1700 to 1400°F. in cycle no. 9 instead of the usual 150° per hr. used in cycle no. 7.

36. The Ar_1 metastable estimated from microstructure of samples in cycle no. 9 would be close to 1315°F. since the sample withdrawn at 1312°F. shows some spots of coarse pearlite already developed. With the Ar_1 metastable established at about 1315°F., it was reasoned that by cooling at 20°F. per hr. to 1330°F. and then cooling at an extremely slow rate such as 3° per hr. from there down, rapid annealing would occur by being below the Ar_1 stable
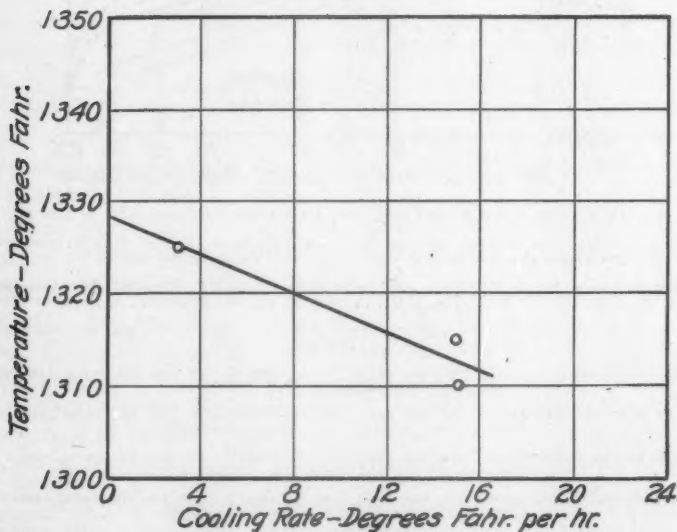
FIG. 11—AR_1 METASTABLE LOCATION

and above the Ar_1 metastable. The progress of annealing on such a cycle is represented in curve no. 10, Fig. 10. Cooling from 1700 to 1400°F. was at 60° per hr., the same as in cycle no. 9, and after cooling from 1400°F. to 1330°F. at 20° per hr., annealing was about 50 per cent complete, exactly the same as in cycle no. 9, but in one hour while cooling from 1330°F. to 1327°F., second stage annealing became about 85 per cent complete with a number of areas of coarse pearlite appearing indicating that at 3° per hr. cooling the Ar_1 metastable is at about 1327°F. In four more hours, annealing progressed only slightly beyond 85 per cent complete — apparently on account of having passed below the Ar_1 metastable at which point the iron carbide was precipitated from austenite forming coarse pearlite. This evidence supports the theory that graphitization from austenite is more rapid than graphitization directly from carbide.

37. In Fig. 11, the information obtained regarding the location of the Ar_1 metastable is plotted to show how the point is depressed, depending upon the rate of cooling. The location of the Ar_1 metastable, by metallographic methods at 1312°F., is not in agreement with the results of Hayes et al[16] who found the Ar_1 metastable when cooling at 15° per hr. to be at 1365°F. Their determination was made by making thermal transformation curves with a Leeds & Northrup Critical Point apparatus.

LIGHT SECTIONS 3/8-IN. THICK

38. Having found that heavy sections represented by C section when heated to 1700°F. at 228° per hr. required 25 hr. to complete first stage graphitization as compared to 5 hr. when heated at 36° per hr., similar tests were made on section D with a thickness of 3/8-in. When heated to 1700°F. at 267° per hr., it required 4 hr. to complete first stage graphitization, whereas when heated from 1525 to 1700°F. at 36° per hr., first stage graphitization was complete in one hour after arriving at 1700°F. The influence of rapid heating obviously is much less damaging to light sections than it is to heavy sections. Table 1 gives the comparison.

Table 1

Time to Complete First Stage — Hours

Heating Rate to 1700°F.	3/8-in Section	13/16-in. Section
About 250° per hr.	4	25
36° per hr.	1	5
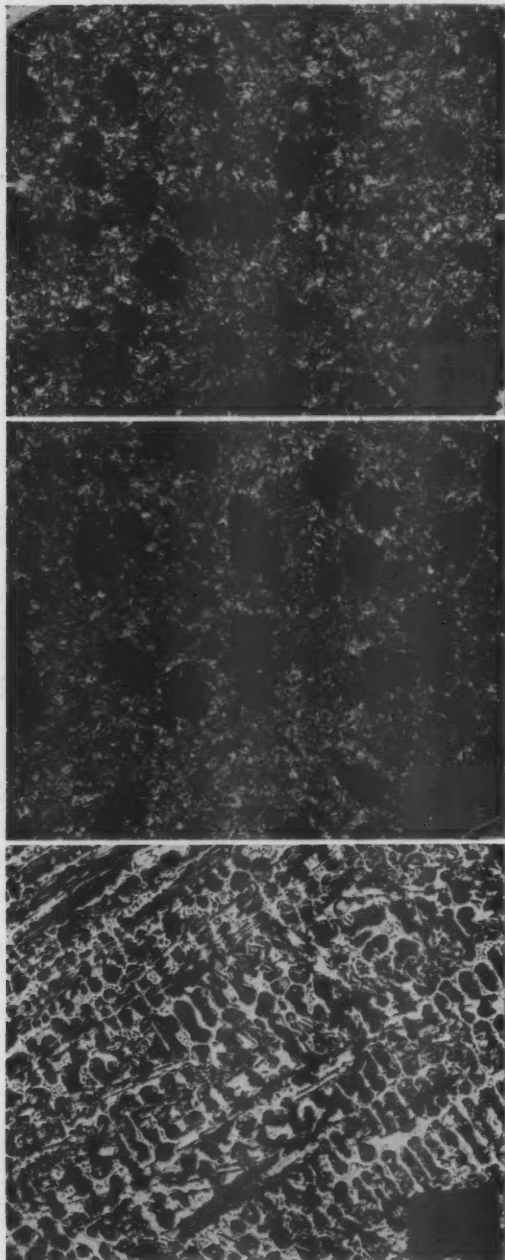
Fig. 12A—White Iron Structure of Light and Heavy Sections and Temper Carbon Size Affected by Section Size and Heating Rate to 1700° F.

(1), (2) and (8) Are Section D—(4), (5) and (6) of Fig. 12A Are Section C.

(1) Shows the As-Cast Structure.

(2)—Section D Heated at 207° Per Hr. to 1700°F, and Held 4 Hr.—Air Cooled.

(3)—Section D Heated at 250° Per Hr. to 1525° F., 36° Per Hr. from 1525 to 1700 and Held 1 Hr.
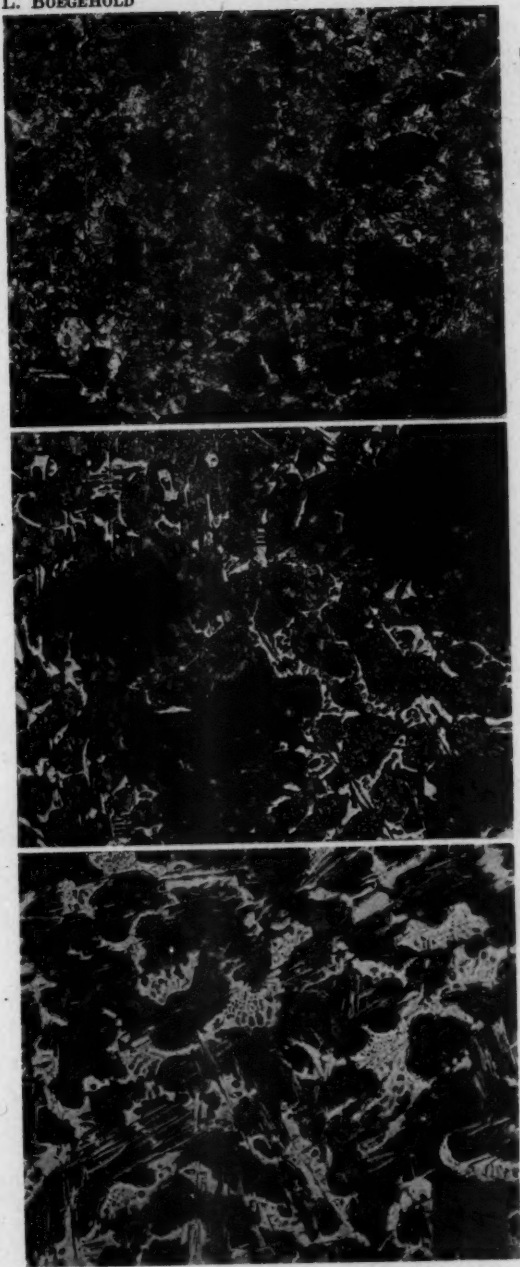
FIG. 12B—WHITE IRON STRUCTURE OF LIGHT AND HEAVY SECTIONS AND TEMPER CARBON SIZE AS AFFECTED BY SECTION SIZE AND HEATING RATE TO 1700°F.

MICROGRAPHS (4), (5) AND (6) ARE SECTION C (SEE FIG. 12A FOR SECTION D).

(4) SHOWS THE AS-CAST STRUCTURE.

(5)—SECTION C HEATED AT 228° PER HR. TO 1700°F. AND HELD 8 HR.—AIR COOLED.

(6)—SECTION C HEATED AT 250° PER HR. TO 1535° F., 36° PER HR. FROM 1525 TO 1700 AND HELD 5 HR.

39. When we compare the relative size and distribution of temper-carbon spots in these samples with the time for completion of first stage graphitization, we suspect that temper-carbon size and distribution is not the sole influence in determining time required for first stage graphitization. The distribution of temper-carbon spots in the 13/16-in. section when heated from 1525 to 1700°F. at 36° per hr. does not appear sufficiently more disperse than in the ⅜-in. section when heated at the same rate to account for the difference between one hour and 5 hr. required in the ⅜ and 13/16-in. sections respectively for completion of first stage graphitization.

40. The temper-carbon spots in the ⅜-in. section heated at 250° per hr. to 1700°F. are distributed about the same as those in the 13/16-in. section heated slowly and the time for completion of first stage graphitization is about the same in both cases. It is presumed that the smaller carbide eutectic structure and more uniform distribution in the white iron exerts some influence upon the speed of graphtization, supplementing and modifying the influence of distance between temper-carbon spots.

41. The relative size temper-carbon spots in the ⅜ and 13/16-in. sections at the two heating rates, and also the carbide structure in the different size sections· is shown in Fig. 12. The distribution of temper-carbon spots and distribution of carbides in the white iron structure, however, appears to have little influence upon the rate of graphitization in second stage graphitization.

42. Fig. 13 shows the progress of second stage annealing at two different cooling rates for both light and heavy sections and also for the light sections when heated to 1700°F. at both high and low rates.. The location of the Ar_1 metastable (Table 2) determined by observing the temperature for occurrence of fine and coarse pearlite in the ⅜-in. samples is the same as in the 13/16-in. samples.

Table 2

Annealing Cycle	Cycle No.	Section Size, Inches	Ar_1 Metastable Temp. of
1525-1700 — 36° per hr. 1700-1400 — 150° per hr., 1400-1360 — 20° per hr. 1360-1300 — 15° per hr.	7	13/16	1310
1525-1700 — 36° per hr. 1700-1400 — 60° per hr., 1400-1360 — 20° per hr. 1360-1300 — 15° per hr.	9	13/16	1315
1525-1700 — 36° per hr. 1700-1400 — 60° per hr., 1400-1330 — 20° per hr. 1330-1315 — 3° per hr.	10	13/16	1327
1525-1700 — 36° per hr. 1700-1400 — 150° per hr., 1400-1360 — 20° per hr. 1360-1300 — 15° per hr.	11	3/8	1310
1525-1700 — 267° per hr. 1700-1400 — 172° per hr., 1400-1360 — 20° per hr. 1360-1300 — 15° per hr.	12	3/8	1315
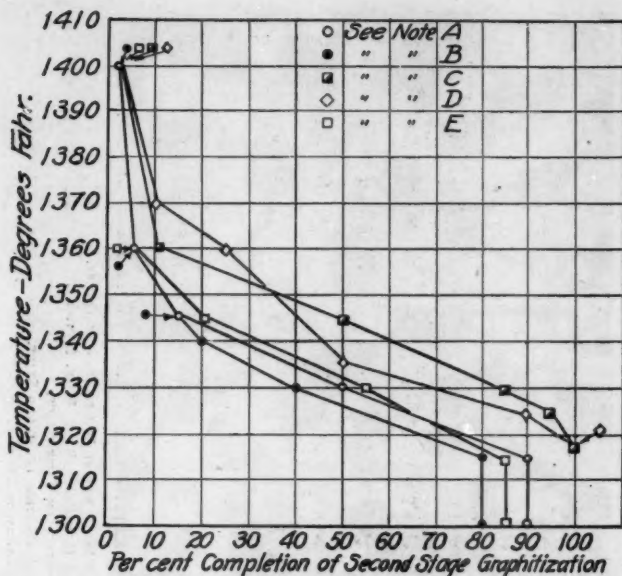
FIG. 18—PROGRESS OF SECOND STAGE GRAPHITIZATION IN LIGHT AND HEAVY SECTIONS.

NOTE A—⅜-IN. SECTION No. 11⎤ 150° PER HR. 1700-1400°F., 20° PER HR.   1400-1360°F.
NOTE B—1¾-IN. SECTION No. 7⎦ 15° PER HR. 1360-1300°F.

NOTE C—⅜-IN. SECTION No. 14⎤ 172° PER HR. 1700-1400°, 7° PER HR. 1400-1315°F.
NOTE D—1¾-IN. SECTION No. 5⎦

NOTE E—⅜-IN. SECTION No. 12—HEATED TO 1700°F. AT 267° PER HR. AFTER FIRST STAGE
        ANNEALING, COOLED SAME AS SECTION No. 11 (SEE NOTE A).

42. Both light and heavy sections, regardless of original carbide eutectic distribution, show the same degree of incomplete second stage annealing when cooled at 15° per hour from 1360°F. and both sections show complete annealing and the same progress at successive stages of cooling from 1400 to 1300°F. when cooled at 7° per hour.

43. In the microstructure at some of the intermediate stages in the secondary stage of graphitization, the arrangement of the ferrite, fine pearlite and temper-carbon, point strongly to the fact that the last pearlite to decompose is always located where the carbide eutectic existed in the cast condition. Photomicrographs in
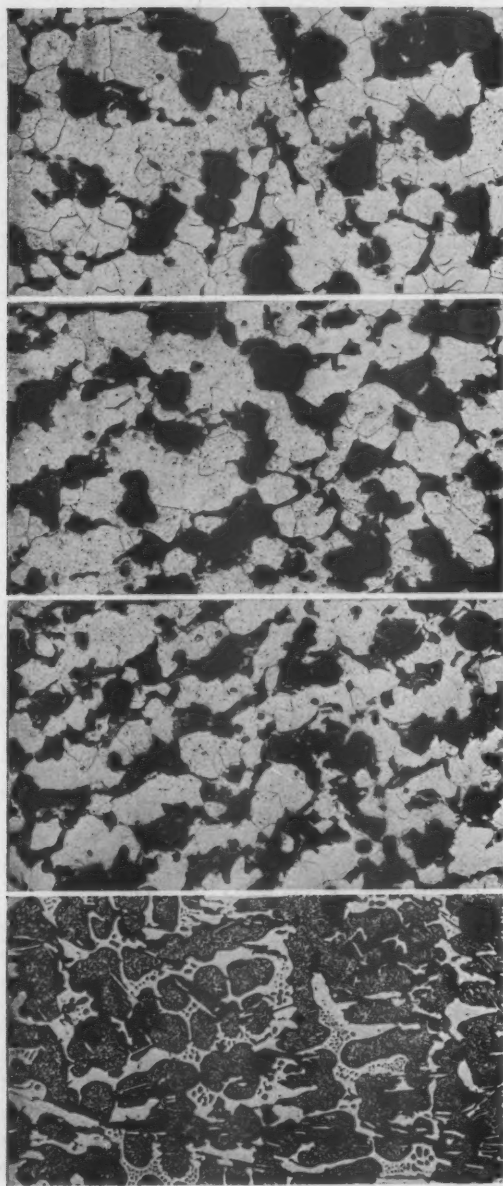
Fig. 14A—Ferrite and Pearlite Areas Reflect Influence of White Iron Structure.
(1), (2), (3) and (4) from 13/16-in. Section (See Fig. 14B for ⅜-in. Section).
(1)—Structure Before Completion of First Stage.
(2), (3) and (4)—Structure Before Completion of Second Stage.
(2), (3) and (4)—Heated Slowly to 1700° F.
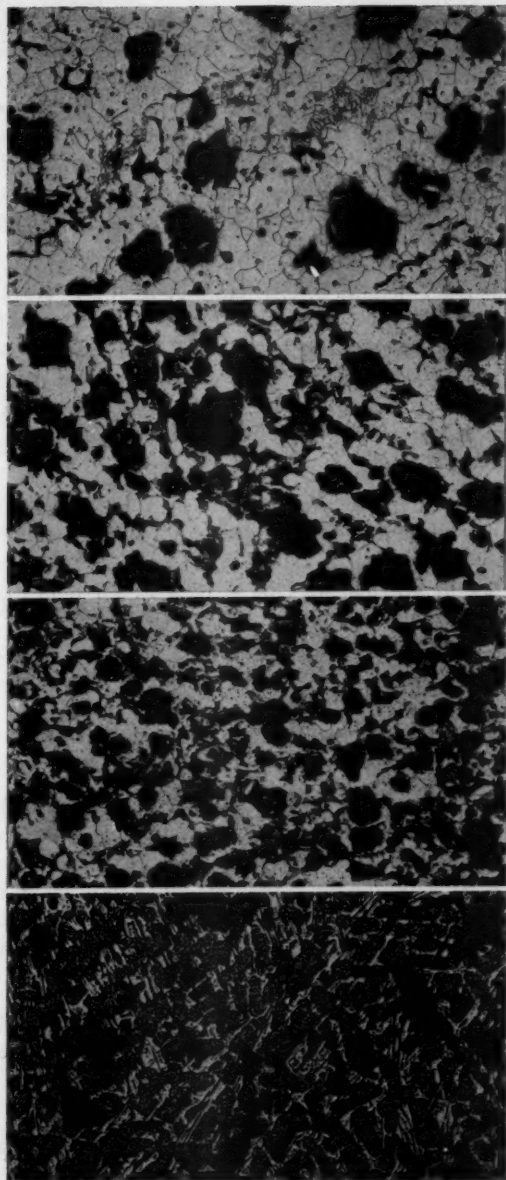
FIG. 14B—FERRITE AND PEARLITE AREAS REFLECT INFLUENCE OF WHITE IRON STRUCTURE.
(5), (6), (7) AND (8) FROM 3⁄4-IN. SECTION (SEE FIG. 14A FOR 13/16-IN. SECTION).
(5)—STRUCTURE BEFORE COMPLETION OF FIRST STAGE.
(6), (7) AND (8)—STRUCTURE BEFORE COMPLETION OF SECOND STAGE.
(6)—HEATED SLOWLY TO 1700° F.
(7) AND (8)—HEATED RAPIDLY TO 1700° F.

Fig. 14 illustrates this point well. The ferrite areas in no. 4 correspond in shape and arrangement to the primary austenite areas in no. 1 between the carbide eutectic areas. It will be noted in no. 1 that temper-carbon is forming most frequently in contact with carbide eutectics and the temper-carbon in no. 4 also is incontact usually with the residual fine pearlite areas. Nos. 2 and 3 also show this type of structure indicating that it is not necessary for the carbon to migrate through ferrite to precipitate in a temper carbon spot.

44.   Photomicrographs nos. 5, 6, 7 and 8 of light sections also show striking similarity between the "as-cast" and "semi-annealed" structure. The presence in no. 8 of both coarse pearlite and fine pearlite indicates that part of the iron has passed through the Ar_1 metastable while other parts have not, which is possibly evidence of the microscopic segregation mentioned by Schwarz and Junge.[15]

45.   The net result of the work described above is that the preferred annealing cycle for the iron in question is as follows: Heat to 1500°F. at a convenient rate, heat from 1500 to 1700 or 1725°F. at 36° per hr. — Hold at 1700 — 1725°F. for 8 hr., cool to 1400°F. at 150° per hr., cool from 1400 to below 1310°F. at 7° per hr.

SUMMARY

46.   Some observations which might be made as a result of this work in connection with this particular kind of iron are:

(1)   The usually accepted idea that large temper-carbon spots always are associated with heavy sections is not always correct. By heating to the primary graphitization temperature at a slow enough rate, the temper-carbon spots in a heavy section may be as small and numerous as those in a light section, thus removing any relationship between as-cast structure and temper-carbon distribution.

(2)   First stage graphitization time increases greatly with increased distance between temper-carbon spots. Carbide decomposition is further retarded by coarse arrangement of iron carbide eutectic and austenite areas.

(3)   Fast heating to the temperature for first stage graphitization causes larger and fewer temper-carbon spots

than when heating is slow. The difference in size of temper-carbon spots as a result of slow and fast heating is much more pronounced in heavy sections than in light sections.

(4) The temperature range in which alpha iron, gamma iron and cementite exist simultaneously when heating at 100° per hr. is from about 1425 to about 1525°F. compared to the published equilibrium diagram of 1380 to 1410°F. for this range.

(5) Complete elimination of pearlite in second stage annealing may be accomplished at 1360°F.

(6) Second stage annealing by cooling from 1360 at 15° per hr., results in incomplete annealing. About 0.10 to 0.12 per cent carbon remains in the combined form.

(7) Ar_1 metastable for this iron was located at about 1325°F. for cooling at 3° per hr. and at about 1312°F. when cooling at 15° per hr.

(8) Primary graphitization of light sections at 1700°F. after slow heating to that temperature is complete in one hour. For heavy sections 5 hr. is required. When heating is rapid the time required is 4 hr. and 25 hr. for light and heavy sections.

(9) Second stage graphitization apparently is not influenced by size and distribution of carbide eutectic and austenite areas in the cast structure. The rate of pearlite decomposition proceeds at the same rate in light and heavy sections.

(10) Evidence is shown supporting the theory that graphitization proceeds most rapidly from solid solution and that temper-carbon precipitates directly from solid solution without migrating through ferrite.

Bibliography

1—Mayne, B. R.: "*The Continuous Process as Applied to Malleable Practice*," TRANS. A.F.A., vol. 38 (1930) pp. 561-572.

2—Bean, W. R., Highriter, H.W. and Davenport, E. S.: "*Fractures & Microstructures of American Malleable Cast Iron*," TRANS. A.F.A., vol. 29 (1920), pp. 306-340.

3—Phillips, A. and Davenport, E. S.: "*Malleablizing White Cast Iron*," TRANS. A.I.M.M.E., January 1922.

4—White, A. E. and Schneidewind, R.: "*The Metallurgy of Malleablization*," TRANS. A.F.A., vol. 40 (1932), pp. 88-117.

5—General Motors Research Laboratory Report MR-1-52, 1931.

6—General Motors Research Laboratory Report MR-1-57, 1932.

7—Hayes, A., Diederichs, W. J. and Dunlap, W. M.: *"Experiments on the Rapid Graphitization of White Cast Irons and a Theory for the Mechanism of Graphitization,"* AMERICAN SOCIETY FOR STEEL TREATING, TRANS., March, 1923.

8—Hayes, A., Diederichs, W. J. and Flanders, H. E.: *"The Mechanism of Graphitization of White Cast Iron and Its Application to the Malleablizing Process,"* TRANS. A.F.A., vol. 32, pp. 1 (1925), pp. 601-623.

9—White, A. E. and Archer, R. S.: *"The Annealing of Malleable Castings,"* TRANS. A.F.A., vol. 27 (1919), pp. 351-365.

10—Schneidewind, R. and White, A. E.: *"The Malleabilization of White Cast Iron,"* UNIVERSITY OF MICHIGAN ENGINEERING RESEARCH BULLETIN, No. 24, August, 1933.

11—Kluytmans, C.: *"Notes on Annealing Black Heart Malleable,"* FOUNDRY TRADE JOURNAL, vol. 34, pp. 550-554, 1926.

12—Smith, C. S. and Palmer, W. E.: *"Some Effects of Copper in Malleable Iron,"* METALS TECHNOLOGY, February, 1935

13—Schwartz, H. A.: *"Some Physical Constants of American Malleable Cast Iron,"* PROCEEDINGS A.S.T.M., vol. 19, part 2, 1919.

14—Greiner, Marsh and Stoughton: *"Alloys of Iron and Silicon,"* THE ENGINEERING FOUNDATION MONOGRAPH, McGraw Hill Book Company.

15—Schwartz, H A. and Junge, C. H.: *"Metallographic Changes During Cooling Between the First and Second Stages of Annealing,"* TRANS. A.F.A., vol. 36 (1936), pp. 507-526.

16—Hayes, A., Flanders, H. E. and Moore, E. E.: *"Identification of the Thermal Effect of the Iron-Carbon Eutectoid and Extrapolation of the Heating and Cooling Curve Values to Zero Rates,"* TRANSACTIONS A.S.S.T., vol. 5, 1924, pp. 183-194.

DISCUSSION

Presiding: L. N. SHANNON, Stockham Pipe Fitting Co., Birmingham, Ala.

H. A. SCHWARTZ[1] *(Written discussion)*: The following comments are offered in elaboration rather than in criticism of Boegehold's paper since there appear to this commentator to be points of interest beyond the somewhat severely practical field to which the author confined himself.

The author's work parallels in part work done by Schwartz and Junge in the paper referred to by Mr. Boegehold in his paper. In that paper the initial appearance of ferrite in network and at a lower temperature in crystallites is recorded (Figs. 1 and 2) and a temperature range of about 25°C. (45°F.) is found for the range in which gamma iron, alpha iron and *graphite* can coexist. The corresponding

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[1] Manager of Research, National Malleable and Steel Castings Co., Cleveland, O.

interval in the author's work seems to have been 90°F., a difference
which is not surprising having in mind differences in the character of
the iron and also that our work was intended to approach equilibrium
conditions as closely as possible, whereas Boegehold concerns himself,
deliberately, with certain cooling rates.

In a reaction as slow as that here under consideration plainly the
amount of an excess phase which will be rejected is a compromise
between what wants to come out and what can come out in a given
time.

That the graphitizing rate of malleable cast iron is largely a
function of nodule number is a necessary mathematical consequence
of the mechanism described by Schwartz in 1926*.

We can confirm the author's observation that nodule number
varies with the heating rate in annealing and may add that this effect
is noticeable even at relatively low temperatures.

In the light of the author's table, "Time to Complete First Stage,"
the immediately preceding statement, "The influence of rapid heating
obviously is much less damaging to light sections than to heavy sec-
tions," is true if one considers the matter solely from the viewpoint
of commercial saving of time, but misleading from a metallurgical
viewpoint of the *ratio* of the time of completion of first stage
graphitization with slow and fast heating is nearly the same (perhaps
as closely the same as the experimental procedure warrants) for both
the thick and thin sections.

This commentator does not wish to assert that graphitizing rate is
solely determined by nodule number. Indeed work to be published
elsewhere has been undertaken to determine this relationship. It is,
however, his opinion that no adequate idea of nodule number can be
obtained by mere visual inspection of a specimen or a few micrographs
and that for an interpretation of any quantitative value a mathematical
evaluation of the nodule number per unit volume must be undertaken.
The enormous differences in nodule number in specimens which look
approximately alike is sometimes surprising.

Such methods of investigation of course are justified only if one
wishes to delve into causes more deeply than is necessary if only the
time for completion of anneal is of interest.

This commentator is also inclined to hold a reservation regarding
the last phrase in the author's first conclusion "thus removing any rela-
tionship between as cast structure and temper carbon distribution."
So long as there is a ratio of some 5 to 1 in the first stage of anneal-
ing whether heating be slow or fast it appears that *something* has
survived which had a relationship to the as cast structure. It is also the
invariable experience in this laboratory that temper carbon begins to
form at some sort of interface**.

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* *Graphitization at Constant Temperature*, Transactions of American Society for
Steel Treating, Vol. 10, page 53.

** *Note on the Metallography of Ferrite in Malleable Cast Iron*, Schwartz and
Junge, Transactions of the A.F.A., Vol. XLII, P. 94 (1934).

Since an entire interface does not cover itself with graphite, obviously some other factor determines the number of nuclei in a given interface. Heating rate, prequenching and other factors may influence this number but the *location* of nodules is still importantly related to freezing structure. Their *number* is a function of many variables of which freezing rate is only one.

That time of completion of the second stage of anneal is not so much related to freezing structure as is that of the first stage is not surprising for the only survival of the freezing structure at the time when second stage annealing begins is the *location* of temper carbon nodules. It would seem likely as a first approximation that the second stage rate should be proportional to their number which is, as has been said, *only in part* determined by the freezing pattern.

Within the rigidly utilitarian scope of this paper the author's conclusions seem thoroughly well established. We will in general learn most from a given practice if we consider not only what is done but why it works. The preceding comments are offered to supply a setting for Mr. Boeghold's specific observations.

C. H. JUNGE[2] *(Written discussion):* An important item brought out in this interesting paper is that white cast iron, as well as malleable cast iron, does not have a transformation temperature at A_1, but rather a range of temperature[3] over which the A_1 transformation takes place. With a heating rate of 100°F. per hour, this spread is 100°F. for the white iron used in this investigation. Upon cooling after first stage annealing has been completed, it is again evident from the micrographs that the Ar_1 transformation also is spread out over a range of temperature, and that Ar_1 stable and Ar_1 metastable may be different for different microscopic areas in the same iron.

Another significant finding is that complete second stage anneal can be accomplished very rapidly by direct decomposition of solid solution to form ferrite and graphite, without ever passing through the pearlite transformation.

Mehl and Wells[4] have shown that the eutectoid concentration in Fe-C alloys is not of a fixed composition, but is dependent upon the rate of cooling. It is thus possible in an alloy to have a pearlite which is either of higher or lower combined carbon than the true equilibrium pearlite composition. A low carbon steel cooling in a mold would have a pearlite lower in combined carbon, while a white iron would have a pearlite higher in combined carbon than the equilibrium value.

Upon reheating the white iron to above the critical, only cementite in the amount necessary to give the equilibrium concentration of solid solution will dissolve, leaving a portion of the cementite plates of the

[2] Research Staff, National Malleable and Steel Castings Co., Cleveland, O.
[3] See also Andrew and Dickie, "The Ac_1 Range in Special Steels," Journal of the Iron and Steel Institute, vol. CXV, No. 1 (1927), p. 647.
[4] R. F. Mehl and Cyril Wells, "Constitution of High-Purity Iron-Carbon Alloys," Trans. A.I.M.E., vol. 125 (1937), p. 429.
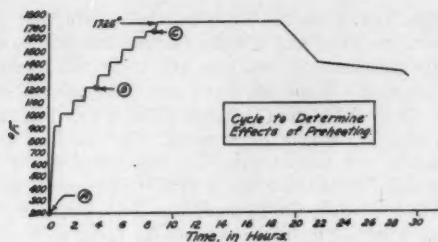
Heat No.	Desig-nation	T. S. (Lbs/sq.in)	Elong. (% in 2 in)	BHN	Estimated % pearlite in matrix	Estimated % massive cementite
832	A	58,200	12½	149	5-10	0
"	B	64,100	10½	163	25	0
"	C	68,900	8	187	60	5
842	A	52,800	9½	143	0	0
"	B	64,800	10½	148	0	0
"	C	54,000	8	174	40	1

Chemical Analyses		
Heat No.	T.C.	Si
832	2.39	1.01
842	2.55	1.12

FIG. 15—CYCLE TO DETERMINE EFFECTS OF PREHEATING.

original pearlite in excess. This situation is readily visible in Fig. 7C (5), where undissolved fragments of the cementite laminations of the pearlite remain in the dark matrix of solid solution.

There seems to be considerable evidence in this paper and elsewhere[5], that temper carbon spots form at cementite-austenite interfaces. We have just seen above that a white iron when heated just above its critical has many small cementite fragments throughout the matrix. Presumably these fragments tend to go into solution more and more as the temperature is raised to that of first stage anneal, say 1725°F. Now if the iron is held for some time just above Ac_1, or is heated *slowly* from Ac_1 to 1725°F., there is obviously a longer time interval in the range where the fragments of pearlitic cementite persist in the matrix, and hence a greater opportunity for temper carbon spots to form at these cementite-austenite interfaces within the matrix, than if the iron is heated *rapidly* to 1725°F. This is suggested as a possible mechanism to explain the observations of this paper that nodule number can be widely different for the same iron, dependent upon the heating rate used at the start of anneal.

H. BORNSTEIN[6]: We have been experimenting with the annealing of malleable iron by putting in a radiant tube furnace and carrying

[5] H. A. Schwartz and C. H. Junge, "A Note on the Metallography of Ferrite in Malleable Cast Iron," Trans. A.F.A., vol. XLII (1934), p. 94.

[6] Director of Laboratories, Deere & Co., Moline, Ill.

on, as one might say, a partial experimental production program. And along with that, we have had a little electric furnace in the laboratory where we have run through test lots and tried to correlate the results in the two furnaces. What we have run into can be shown by two charts (Figs. 15 and 16) and a table (Table 3). Fig. 15 shows data from the laboratory furnace experiment. We put our specimens in at room temperature and heated them up rapidly to 900° F., and then at about 100° an hour from there up to 1725° F. In one of the experiments, we put a test bar in when the furnace was cold, another at about 1200°, and another at 1700°. Fig. 15 shows results from such test bars of two heats. Heat No. 842 was a higher silicon (1.12 per cent) and breaks down easier, while heat 832 has one per cent. On the lower silicon heat we found the best results in putting it in the cold furnace, poor by putting it in at 1200°, and still poorer by putting it in the hot furnace at 1700°, as is shown by the Brinell figures and the amount of pearlite that is left at the end of the annealing period. And in the case Heat No. 832, we get around 5 to 10 per cent of pearlite putting it through the regular cycle, but the pearlite increases up to 60 per cent by putting it into the hot furnace.

The second set of bars being easier to anneal, we get no pearlite left by putting bars in cold furnace or at 1200° F., but we get quite a lot of pearlite in putting it into the hot furnace.
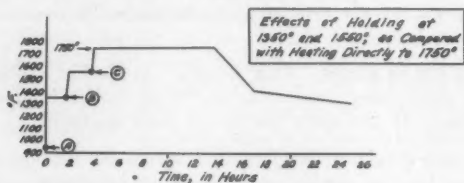
Heat No.	Desig- nation	T.S. (Lbs./sq. in.)	Elongation (% in 2 in.)	Estimated % pearlite in matrix	Massive Cementite
700	A	54,500	15	0	0
"	B	54,000	11	25	0
"	C	54,100	10	25	Tr
716	A	58,700	10	0	0
"	B	61,600	11½	40	0
"	C	61,900	10½	40	Tr
735	A	57,600	16	0 - Tr	0
"	B	60,900	12½	25	0
"	C	61,800	8	50	Tr +
737	A	60,600	21	5	0
"	B	67,200	15	30	0
"	C	67,400	12½	50	Tr +

Chemical Analyses		
Heat No.	TC	Si
700	2.76	1.19
716	2.47	1.20
735	2.45	1.22
737	2.37	1.19

FIG. 16—EFFECTS OF HOLDING AT 1350°F. AND 1550°F., AS COMPARED WITH HEATING DIRECTLY TO 1750°F.

Fig. 16 shows something similar excepting that we go on a little bit further. In the laboratory furnace we held for periods of about two hours at 1350° and 1550°. The idea originally was that it took about six hours to come up to temperature in the one-ton furnace and we wanted to take about the same length of time in coming up to temperature in the laboratory furnace, so we more or less arbitrarily held for two hours at 1350° and two hours at 1550°. And then we found that it made considerable difference as to whether or not we had this holding period at 1350°. Fig. 16 shows data obtained from test bars from each of four heats.

"A" in each case represents the bars that were through the complete cycle; "B," the bars that were put in at 1350°; and "C," bars that were put in at 1550°. Now the bars put in at 1350° did not remain at that temperature for any length of time because they were put in when the furnace was being changed to heat up to 1550°, and we find a material difference there. In each case the bars having treatment "A" show very much better results than bars having treatments "B" or "C." This would indicate to us there was considerable benefit in the hold at 1350°. That point is further confirmed by the data of Table 3.

We thought some of that difference might be due to the fact when you put it in a little later on you have not as many hours of temperature. So we ran several experiments, made them both the same length of time.

On this particular iron of Table 3, 23 hours was about the best annealing cycle we could get. In one case (AV) we heated and held at 1350°. In the other case (AU), we put the bars in after the 1350°. But in the second case, the two hours that we had taken off at the start we got back by holding two hours longer at 1750°, and the dif-

Table 3

EFFECT OF PREHEAT ON ANNEAL

Annealing Cycle	AV	AU
Heat to 1350°F	2 hrs.	_ hrs.
Heat to 1550	2	2
Heat to 1750	2	2
Held at 1750	6	8
Cool 1750-1400	2½	3¼
Cool 1400-1300	8½	8
		23¼

Annealing Cycle	Heat No.	Per Cent Silicon	Tensile Strength lb. per sq. in.	Elongation Per Cent	Pearlite in Matrix Per Cent
AV	700	1.19	55,750	19	None
AU	700	......	59,000	11	25
AV	716	1.20	56,500	19	None
AU	716	......	61,600	11.5	40
AV	735	1.22	58,600	18.5	None
AU	735	......	60,900	12.5	25
AV	737	1.19	62,400	19	5
AU	737	......	67,200	15	30

The temper carbon particles in AU were larger and more widely separated than in AV. Fractures of test bars from AV showed normal malleable while bars from AU showed coarser fracture sparklers.

ferences in results are shown in Table 3. It will be noticed that in each case we get substantially better annealing where we have had the 1350° hold and, furthermore, there is a definite effect on the size of the carbon particles. The temper carbon particles in "AU" were larger and more widely separated than in "AV." Fractures of test bars from "AV" showed normal malleable, while bars from "AU" showed coarser fractures with sparklers.

All this simply indicates to us that it may be possible, for each particular composition of iron, to save time perhaps by finding the temperature at which you should hold these irons and, after you have the holding period, whether it is 1350° or 1400° or 1450°, you can probably heat rapidly from that point on to your regular holding temperature, which may be 1700 or 1750°.

The experimental work in our laboratory is being continued to find the optimum range of holding temperatures.

R. J. COWAN[7]: This paper brings to mind a number of disconnected experiments which seem to me to have a bearing on the subject. I would like to bring them in as a series of questions regarding the effects that have been noted.

The first is the case of a test sprue, from a heat that is in the process of melting down. If that test sprue be poured cold, of course, we know it is mottled, but if you get it to the point, (or if you have a composition) which will not be mottled at a low temperature, so that you get a clear sprue, and then periodically, as the metal comes to temperature, pour another sprue and another sprue and another sprue, the metal will be clear all the way through. If these pieces are annealed together, it will be found that the temper carbon size in the annealed metal is all different, and the grain size of that metal will be different also.

This suggests also another test which was run in which hard iron was quenched from a variety of temperatures, progressively, and then annealed together. Obviously it had the same composition. But the grain size of the resulting metal was decidedly different. And the temper carbon size was decidedly different after annealing, and they were all annealed alike.

In the third test to be mentioned the hard iron was held at a temperature of approximately 1000°F. for a period of something like 20 hours. It was then given the usual annealing treatment. After the 20-hour hold at 1000°, it was found that first-stage graphitization could be completed in less than two hours, whereas in the same metal not so held, it took ten to twelve hours at a temperature of 1750° to complete first-stage graphitization.

This matter all comes back not only to graphite or temper carbon sizes, but it seems to me to have some relationship to grain size, because as we get small temper carbon particles we have fine grain size. And in this experiment which I just mentioned, where the metal was

held at low temperature, 1000°, the thing that was most noticeable was the fact that the temper carbon particles were extremely small and the grain size also was small.

There has been a question in my mind whether or not we may be paying too much attention to graphite and the mechanism of graphitization. Would it not be well for us to pay attention to the inherent grain size? Let's take a leaf from the experience of the steel mills and consider the inherent grain size or what would be the equivalent of inherent grain size in malleable practice, considering the size of the ferrite grain that results after the anneal, for that, I believe, is the secret or one of the secrets of rapid malleablization.

There is another aspect of that matter. We all know that aluminum has a tremendous effect upon the graphitizing rate. We know also that aluminum has a tremendous effect upon grain size. Ordinarily it is a good practice to keep away from aluminum because it is hard to control. But if it were properly controlled, it seems to be working along the same line as this idea of controlled grain size with accelerated graphitizing rate.

Now, I am wondering if there isn't some connection between these various scattering remarks and the work of Mr. Boegehold.

H. A. SCHWARTZ: Would the speaker care to elaborate on his statement "graphitization can take place without the migration of carbon through ferrite"?

MR. BOEGEHOLD: My thought on that subject is that during all times of second stage graphitization there appeared always to be some combined carbon in contact with the temper carbon spots, so there was a ready avenue for the carbide to decompose right into the temper carbon spots without having to go around through any ferrite.

A. L. BOEGEHOLD (Written closure): Referring to Mr. Cowan's experiment where sprues were poured at different temperatures starting with a temperature so low that the sprue was mottled, I can see that the annealed sprues would not all have the same temper carbon distribution because of some difference in solidification rate—as explained in the paper just abstracted and possibly because of a difference in non-metallic nuclei not yet in solution or agglomerated, having a particular influence upon the crystal formation during solidification. In one experiment we conducted, we examined the annealed structure of the same iron in one case heated to 2850°F. and in the other to 3000°F. before casting and found no difference in the temper carbon distribution. It was closely packed in both when heated slowly to 1700°F. and coarse when heated rapidly to 1700°F. in the anneal. It didn't in this particular iron.

Quenching the casting from various temperature and its effect upon annealing has been reported in the literature. The result is similar to that obtained from the rapid cooling in light castings only more pronounced. The temper carbon spots are much smaller and more numerous consequently annealing proceeds more rapidly.

Mr. Cowan's experience with preheating at 1000°F. agrees with some work we have done with preheating at temperatures as low as 800°F. This treatment does produce smaller temper carbon spots with certain particular varieties of malleable iron. The mechanism is probably something akin to age or precipitation hardening. By providing sufficient time for the microscopic or sub-microscopic particles to precipitate, we obtain nuclei for temper carbon precipitation. As Mr. Cowan says—we probably can take a leaf from the steel makers' book and go through certain oxidizing and reducing operations in order to form certain non-metallics in the melt which will behave later on as nuclei for graphitization.

Mr. Bornstein has presented some very interesting information. He has devoted more attention to the heating up phase of annealing than we have and has obtained some valuable results. I am glad to see that in general they confirm the results presented in the paper.

Mr. Junge's explanation of the possible formation of temper carbon spots at the many austenite-carbide interfaces within the areas bounded by the massive cementite areas is plausible and there seems to be no reason for thinking that it does not occur in that way. It should be pointed out, however, that the carbide particles remaining out of solution after the pearlite has gone in solution are probably not part of the pearlite but represent the pro-eutectoid cementite portion of what was a saturated austenite containing approximately 1.7 per cent carbon. That this pro-eutectoid carbide did not precipitate out in network form the way that constituent usually appears makes it somewhat more difficult to recognize. It is hardly reasonable to assume that the large quantity of undissolved carbide within the austenite in Figure 7C is part of the eutectoid carbide.

Dr. Schwartz' comments regarding the difference between the Ar temperature range of 45°F. for equilibrium conditions and 90°F. for certain cooling rates is timely. This is a difference which the author wishes to be clearly appreciated. It has come to be considered scientific to determine equilibrium conditions. Industrially, however, this kind of information is often worse than useless because it may be misleading to some people not sufficiently informed. The author contends therefore a determination of conditions at various cooling rates is equally scientific and infinitely more useful.

Referring to the statement that "The influence of rapid heating obviously is much less damaging to light sections than to heavy sections," the author does not know of any authority for Dr. Schwartz' assumption that a conclusion drawn from a mathematical difference between two numbers is not as much a metallurgical viewpoint as a conclusion drawn from the ratio of two numbers. We are interested in time differences so this author still prefers the conclusion drawn from the difference in time required by the light and heavy sections for first stage annealing.

Spheroidized Pearlitic Malleable

By Duncan P. Forbes,* Rockford, Ill.

Abstract

The paper discusses the fact that malleable iron containing pearlite has been produced accidentally by malleable foundrymen for many years. Only recently has much attention been given by malleable metallurgists to the production of commercially valuable pearlitic malleable. The process of manufacture and the microstructure of spheroidized pearlitic malleable are described. The paper outlines the method and results of a series of tests to determine the influence of manganese added to the metal as a graphitization retardant under various conditions of heat treatment. The author concludes that manganese is not a great factor in retarding graphitization above the critical temperature, but that below the critical temperature its influence is great in controlling the combined carbon content of the final product. The physical properties of the metal are dependent upon the final combined carbon content and, therefore, also upon the percentage of manganese present.

1. Before considering certain phases of the type of metal known as spheroidized pearlitic malleable, or more simply by the trade name "Z-Metal," under which name it is quite generally marketed, it is profitable to make a quick survey of the entire class of ferrous products characterized by the presence of both temper carbon and combined carbon in the microstructure. This material is rather inadequately named "pearlitic malleable iron."

2. From the earliest days of the black heart malleable process malleable foundrymen have made inadvertently, and to their great regret, large quantities of pearlitic malleable. This generally has been of two forms:

   (1) Where the entire microstructure contained appreciable quantities of pearlite.

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* President, Gunite Foundries Corporation.

(2) Where the partly decarburized surface layer of the cast-
ings contained appreciable quantities of combined car-
bon, but where the core of the casting showed a per-
fectly normal malleable iron structure.

3. The production of these materials usually was accidental
and the metallurgy of malleable iron has been concerned largely
with ways of avoiding the occurrence of combined carbon in the
malleable iron structure. In light of present knowledge, it is fairly
certain that such malleable iron contained combined carbon after
annealing because of:
(1) Incorrect analysis,
(2) Incorrect heat treatment, or,
(3) A combination of both.

4. It is now understood that within certain limits, an analy-
sis which tends to promote combined carbon under one set of heat
treating conditions, may give perfectly normal malleable if heat
treated by a proper heat treating cycle.

5. The malleable foundryman has learned to use metal of a
chemical analysis within certain limits, which is then heat treated
in conformity with a certain practice he has developed, and he gets
a surprisingly uniform product. However, it is more surprising
that the malleable metallurgist has devoted his attention to learn-
ing how to avoid combined carbon in his product, and has neglected
to learn how to produce combined carbon under such conditions
that will give materials of commercial value.

6. The first serious attempt to correlate and publish informa-
tion on various types of pearlitic malleable occurred at a meeting
sponsored by the Cleveland District Committee of the American
Society for Testing Materials, held in Cleveland, Ohio, January 27,
1936, at which a symposium on pearlitic malleable cast iron was
presented. This symposium* classified so exhaustively the various
types of material possible in this field that it is unnecessary to
repeat the classifications in this paper.

7. However, it is one thing to outline and explain the broad
principles involved in production of certain types of metal and
another thing to study completely each potential product so that
the manufacturing problems are clearly understood permitting full

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* Published as a reprint by The American Society for Testing Materials, 260
Broad St., Philadelphia, Pa. Also available through the American Foundrymen's
Association, 222 West Adams St., Chicago.

advantage to be taken of the physical properties inherent in the metal.

SCOPE OF PAPER

8. It is the purpose of this paper to consider the effect of the graphite retarding element manganese in that type of product which is classified in the symposium as Alpha Family, Division I, Class *B*, Group 5, Type *d*. This metal is characterized by being produced from white cast iron containing elements retarding graphitization, in which metal the massive cementite has been decomposed and the carbide incompletely graphitized, resulting in a microstructure containing temper carbon and spheroidized cementite in a matrix of ferrite so distributed as to be substantially homogeneous. Such a metal is known as "spheroidized pearlitic malleable."

METHOD OF MANUFACTURE

9. As manufactured commercially, a suitable white iron metal is prepared, usually in an air furnace, of an analysis suitable for the production of malleable iron castings. A typical analysis could be as follows:

	Per Cent
Total Carbon	2.43
Silicon	1.03
Manganese	0.34
Sulphur	0.08
Phosphorus	0.15

10. At the time the iron is tapped into the pouring ladle, an addition of ferromanganese is made to increase the manganese content to the percentage desired.

11. A high manganese content can be produced in the melting furnace by a suitable selection of raw materials, but there is evidence that ladle additions of manganese give more beneficial results than would be the case if the entire amount of manganese were melted in the original charge. Commercially it is desirable to make ladle additions in any case, because producers of this material usually produce more than one grade of metal from each furnace melt and this is done more easily by varying the percentage of manganese than it is by varying the heat treatment, as will be explained later.

12.   Contrary to the usual metallurgical belief, additions of ferromanganese to white iron intended for heavy sections are more apt to produce mottled iron than would the same base iron poured into the same castings without the ferromanganese addition.

13.   It appears in practice that an 80 per cent ferromanganese with a 4.00 per cent carbon content is more apt to cause primary graphite precipitation than an 80 per cent ferromanganese with 1.00 per cent carbon or less. This hardly can be explained by the increased percentage of carbon going into the castings because the amount of carbon added in the ferromanganese can not increase the carbon in the castings by more than three or four points. Probably the tendency toward primary graphitization can be explained more easily by some theory of nucleus formation than by actual increase in carbon content.

14.   The foundryman, however, must take into consideration this action in pouring castings of heavy section and usually he solves his problem by keeping the total carbon of the melt at as low a point as possible, without incurring difficulties in the castability of the metal. The lower carbon content, incidentally, improves the physical properties of the resulting product by reducing the amount of temper carbon present.

HEAT TREATMENT

15.   After the castings have been poured from a suitable mixture, they are cleaned and subjected to heat treatment. The heat treatment may follow several different forms but, in general, it consists of five steps:

(1)   Heating to a holding temperature (approximately 1700°F.).
(2)   Holding at this temperature for a time sufficient to decompose massive cementite completely.
(3)   Cooling rapidly to a temperature below the critical range to produce a fine grained pearlite in the matrix.
(4)   Holding the castings at a temperature in the "spheroidizing range."
(5)   Cooling the castings rapidly to room temperature.

16.   During step (1), the time required depends upon the furnace equipment and the rate at which the castings can be brought up to the holding temperature.

17. Time required for step (2) is dependent principally upon the temperature used, the analysis of the casting, and the grain size of the white iron. This latter depends to a considerable extent upon the rate of cooling of the castings during solidification. It appears that the time required during step (2) is reduced if the massive cementite is present in a large number of small particles, instead of fewer larger particles. Probably the time of decomposition is roughly inversely proportional to the surface area of the cementite crystals.

18. The time required for step (3) is important because, if the castings are cooled slowly from the holding temperature to the spheroidizing temperature, there is a tendency toward the formation of coarsely laminated pearlite. These coarse laminations tend to persist during the spheroidizing treatment, resulting in decreased ductility and resistance to shock.

19. There has been a fear that if the cooling rate in the third step were too rapid, massive cementite would be rejected at the grain boundaries by the austenite in cooling, which would result in unfavorable spheroidizing action. It appears, however, that this condition does not occur readily in castings cooling in air and it is, therefore, good practice to accelerate this cooling rate as much as possible, without actually quenching in a liquid medium.

20. After the completion of step (3), the metal is in a true "pearlitic malleable" condition, and such a metal has wide commercial possibilities. It is characterized by high ultimate and yield strength and a fair measure of ductility.

21. The spheroidizing treatment of step (4) serves the purpose of greatly increasing the ductility and resistance to shock and also providing machinability approaching that of malleable iron. The time of the spheroidizing treatment in step (4) depends on the condition of the pearlite as a result of step (3) and upon the degree of spheroidization desired. In commercial practice, considerable time is allowed for this step because much is to be gained by promoting the maximum ductility.

22. The influence of various alloys on the spheroidizing tendency of this material has not been studied widely. It appears that the action of surface tension of the cementite in transforming fragments of the cementite laminations of pearlite into the spheroidal form, results in a fairly homogeneous matrix.

23. Step (5), cooling rapidly from the spheroidizing temperature to room temperature, is a matter largely of convenience. There is evidence, however, that rapid cooling prevents the embrittlement of the ferrite grains in which the spheroidized cementite is imbedded.

Microstructure

24. Before going further with the particular tests concerning which this paper was written, it would be well to touch on the microstructure characteristic of spheroidized pearlitic malleable iron of the Z-metal type. Temper carbon appears to be present in substantially the same form and distribution as found in commercial malleable iron. One may assume that the temper carbon nodules are smaller than in malleable because part of the total carbon remains in combined form. There is no reason to believe that there are fewer carbon nodules than in malleable and, therefore, the average must be smaller.

25. If the total carbon is 2.40 per cent and the combined carbon 0.60 per cent, the graphitic carbon would be 1.80 per cent, or 25 per cent less than in completely graphitized material of the same total carbon content. Because of the fact that the temper carbon lies in three dimensions, the average diameter of each nodule, decreases as the cube root of the temper carbon content; so that the apparent diameter of each graphite nodule of 25 per cent less volume is approximately 9 per cent less, which would not be readily apparent on visual examination.

26. The matrix of this type of pearlitic malleable consists of ferrite grains, in which are dispersed spheroids of cementite. The spheroids may be large or small, depending upon the condition of the pearlite before spheroidizing and the extent to which the spheroids dissolve during this treatment. At one time, it was believed that the spheroidized cementite was a metallographic constituent having certain fixed analyses similar to pearlite. It seems now, however, that this is not the case and that this metal can have a widely varying combined carbon content, the chief difference between high carbon and low carbon matrix being the number and size of spheroids present in the ferrite.

27. Under the microscope, it is sometimes difficult to distinguish the grain boundaries of the ferrite because the etching is more pronounced in outlining the spheroids. With care, how-

ever, the grain boundaries can be distinguished and the ferrite grains appear to have approximately the same size as ferrite in malleable iron.

PHYSICAL PROPERTIES

28. In the production of spheroidized pearlitic malleable, it is possible to have a wide variety of properties and this is the case also with practically all other pearlitic malleable irons. In fact, with pearlitic malleable iron, it is possible to duplicate almost any grade of steel, with the presence of temper carbon as the important difference. In the case of this spheroidized pearlitic malleable, made under the trade name Z-Metal, the physical properties depend largely upon the percentage of combined carbon present and the uniformity and fineness of the spheroid distribution.

29. The combined carbon content can be varied either by heat treatment or by percentage of graphite retardant present in the metal, or by a combination of both. There has been a theory that the temperature from which the metal was quenched in step (3) of the heat treatment tended to fix the percentage of combined carbon which was to remain in the castings after step (4).

30. To test this theory, a series of A. S. T. M. standard malleable test bars were prepared of the analyses shown in Table 1. They were all produced from a single air furnace heat. The no. 1 series was of normal malleable composition. The no. 2 series used the same base metal, with the addition of approximately 0.37 per cent manganese. The no. 3 series was of the same analysis, with the addition of 0.66 per cent manganese. The analyses shown in Table 1 were taken from hard iron coupons poured from the same ladle as the test bars.

31. All of these bars were then submitted to the first two steps of the heat treating process, as follows:

> Heat to 1720°F. in 20 hr.
> Hold at 1720°F. for 24 hr.
> Cool in air to room temperature in 1½ hr.

32. The bars then were sorted into four groups of six bars, each group containing duplicate bars from each of the three series shown in Table 1. The groups were identified as groups A, B, C and D, respectively. These bars showed the following combined carbon contents:

	Per Cent Carbon
Series (1)	0.665
Series (2)	0.725
Series (3)	0.710

33. The bars from group A were reserved for later heat treatment by steps (4) and (5). The remainder of the bars were heated to the temperatures indicated below, held at temperature for 20 hr., and cooled in air, so that they reached a black heat in approximately 3 min. after removal from the furnace.

	°F.
Group B Bars	1600
Group C Bars	1500
Group D Bars	1400

34. It was assumed that a 20-hr. period was long enough to bring the combined carbon content of the austenite to substantial equilibrium with graphite at the temperature used. The combined carbon content is shown in Table 2.

35. It is probable that had the bars been quenched in oil or water, instead of being cooled in air, a slightly higher combined

Table 1

ANALYSES OF BARS TAKEN FROM HARD IRON COUPONS

Series	T. C.	Si.	Mn.	P.	S.
	%	%	%	%	%
1	2.43	1.03	0.34	0.15	0.08
2	2.37	1.03	0.71	0.15	0.08
3	2.33	1.02	1.00	0.15	0.08

Table 2

HEAT TREATMENT AND COMBINED CARBON PER CENT OF BARS

Group	Heat Treatment	Series 1 %C.C.	Series 2 %C.C.	Series 3 %C.C.
B	As air cooled from 1600°F.	0.850	0.860	0.835
	After 30 hr. at 1280°F.	0.070	0.255	0.500
C	As air cooled from 1500°F.	0.735	0.810	0.760
	After 30 hr. at 1280°F.	0.140	0.235	0.430
D	As air cooled from 1400°F.	0.570	0.665	0.700
	After 30 hr. at 1280°F.	0.050	0.260	0.545
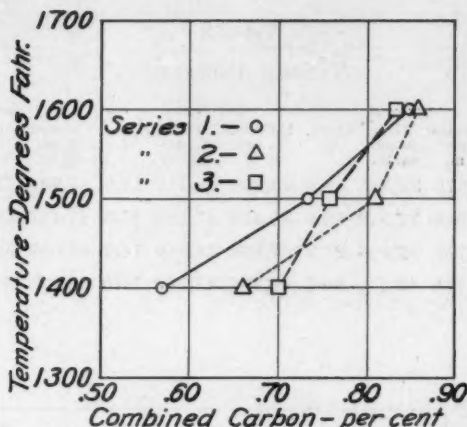
FIG. 1—SOLUBILITY OF CARBON IN AUSTENITE—STABLE SYSTEM.

carbon percentage would have been obtained. It is doubtful, however, if either a longer exposure or a more drastic quench from temperature would have altered greatly the final product.

36. It is noted by an examination of Table 2 that the temperature at which the samples were held prior to air cooling seems to determine the percentage of combined carbon soluble in austenite in the presence of temper carbon. The results are plotted in Fig. 1 and the lines correspond to the A_{gr} line in the stable iron carbon diagram.

37. In view of the fact that the analytical methods used for combined carbon were probably accurate only to plus or minus 0.05 per cent, it is reasonable to believe that the points shown in Fig. 1 may all lie close to a single line and that the percentage of manganese in the metal between the limits of 0.34 per cent and 1.00 per cent has little or no effect on the amount of carbon soluble in the stable equilibrium system.

38. All of the bars were then heated to 1280°F., held for 30 hr. and cooled in air. Comparison of the final combined carbon percentages shown in Table 2 and the tensile properties shown in Table 3 indicate that the holding temperature prior to rapid air cooling has little or no effect on the physical properties or combined carbon percentages after the spheroidizing treatment.

39. Fig. 2 shows the tensile properties of bars from each series with each group plotted against combined carbon content as

Table 3

TENSILE PROPERTIES

Group	Series 1			Series 2			Series 3		
	Ultimate lb. per sq. in.	Yield lb. per sq. in.	Elong. %	Ultimate lb. per sq. in.	Yield lb. per sq. in.	Elong. %	Ultimate lb. per sq. in.	Yield lb. per sq. in.	Elong. %
A	56,670	39,590	16.5	69,290	48,470	16.5	85,990	57,290	10.5
B	55,900	36,940	16.0	70,880	47,120	15.5	87,150	60,310	13.0
C	56,710	38,900	21.3	67,130	46,040	14.7	83,090	58,090	11.5
D	58,930	40,410	20.0	67,800	47,390	16.0	83,520	57,330	12.3
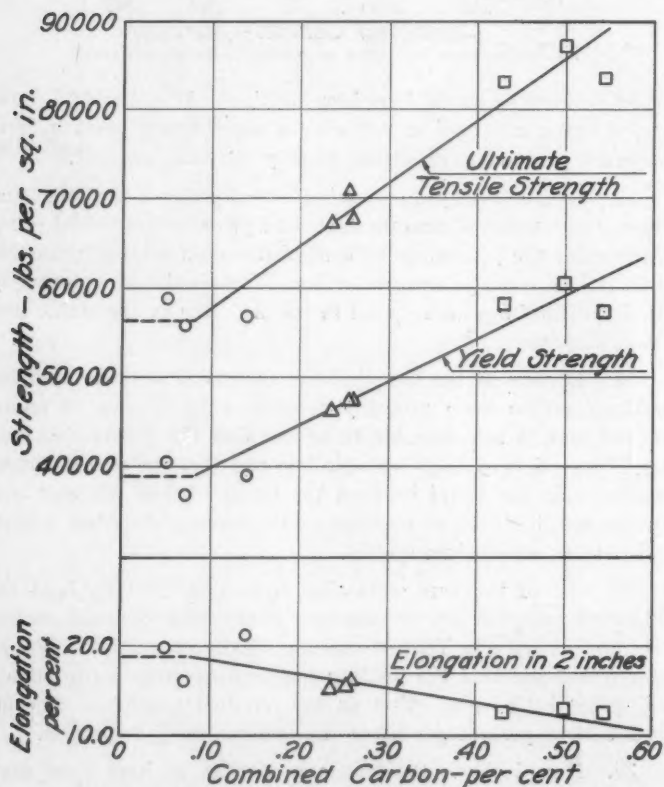
FIG. 2—TENSILE PROPERTIES OF BARS AS AFFECTED BY PER CENT OF COMBINED CARBON.
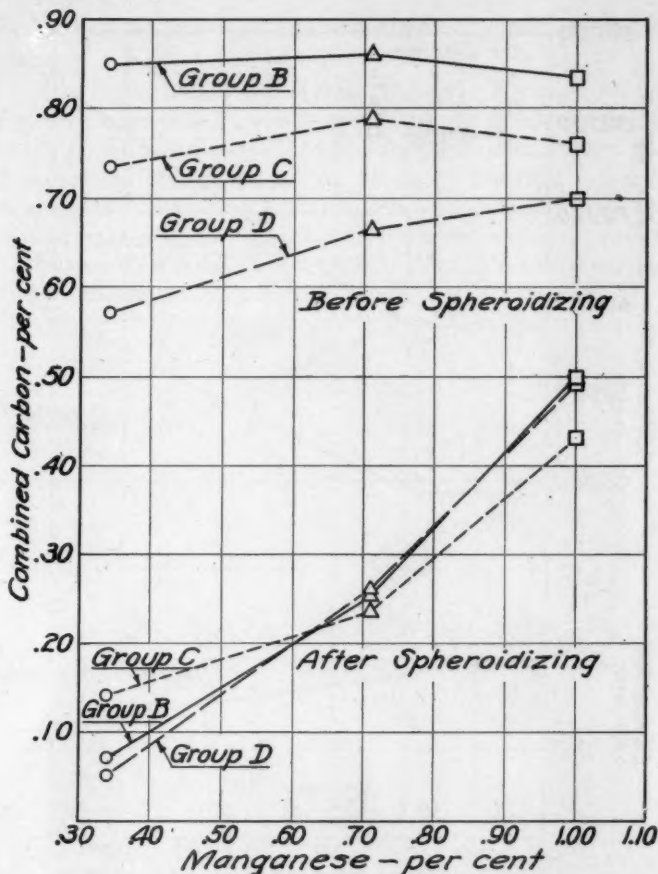
FIG. 2—RELATIONSHIP OF COMBINED CARBON TO MANGANESE CONTENT BEFORE AND AFTER SPHEROIDIZING.

the abscissa. The ultimate tensile strength and yield strength increase and the elongation decreases with increasing combined carbon content. This same relationship, of course, holds true for steel of increasing carbon content.

40. The author has taken the liberty of drawing the curves shown in Fig. 2 so that a break occurs at 0.08 per cent combined carbon, which is taken as the approximate percentage of combined carbon which is soluble in ferrite. In other words, it has been assumed that the physical properties of the metal are not altered,
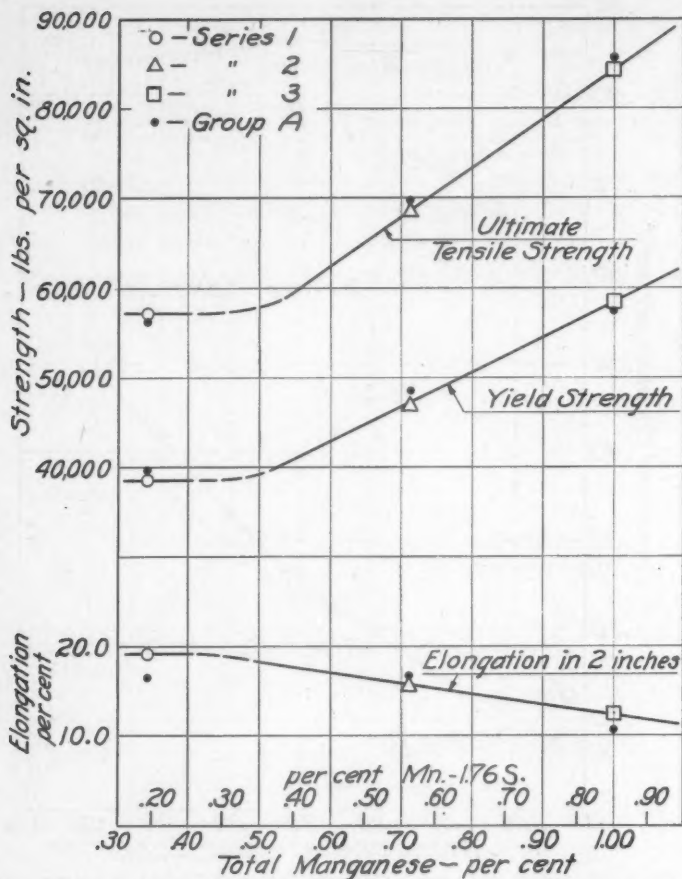
FIG. 4—TENSILE PROPERTIES OF BARS AS AFFECTED BY PER CENT MANGANESE.

except in the presence of spheroidized cementite. This assumption is not supported by evidence uncovered by this series of experiments and is mentioned in this paper simply because general experience seems to indicate that the assumption is correct.

41. In Fig. 3, the combined carbon content of the samples, as given in Table 2, have been plotted against the manganese content of the metal. The curves show both the combined carbon content after air cooling from the respective temperatures above the

critical and the combined carbon content after the spheroidizing treatment.

42.   Fig. 4 shows the average physical properties obtained from all the group *B*, *C*, and *D* bars (after the spheroidizing treatment) plotted as a function of the manganese content. Two scales of manganese content are shown, one being the total percentage of manganese and the other being the percentage of manganese remaining after deducting the 0.14 per cent manganese required to ''balance'' the 0.08 per cent sulphur content, which is supposed to be present in the form of MnS.
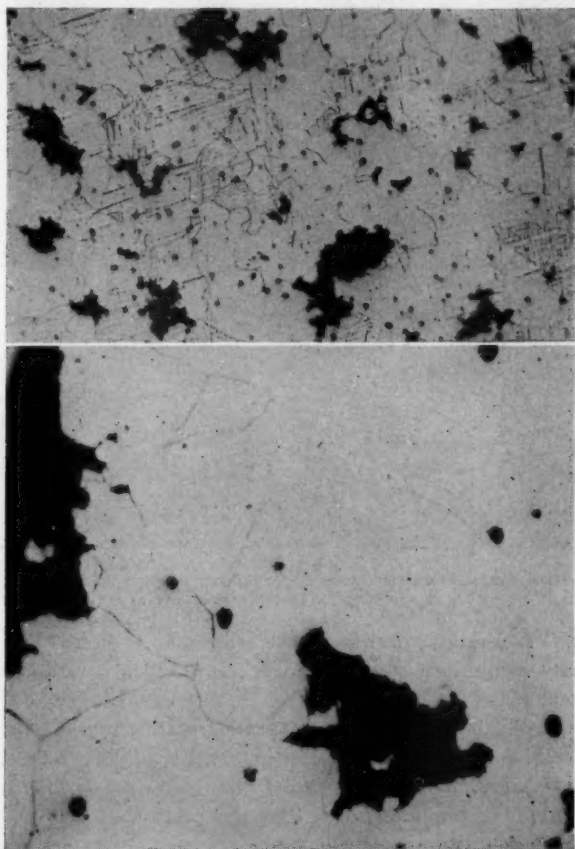
FIG. 5—SAMPLE FROM SERIES 1, GROUP B—MANGANESE CONTENT 0.34 PER CENT—NITAL ETCH—(UPPER) 100X—(LOWER) 500X.
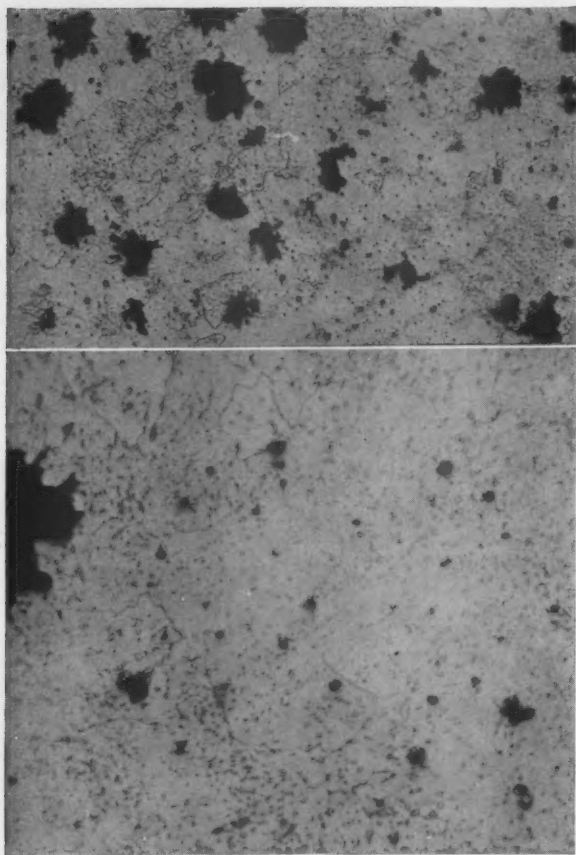
Fig. 6—Sample from Series 2, Group B—Manganese Content 0.71 per cent—Nital
Etch—(upper) 100x—(lower) 500x.

43. It would appear from the shape of the curves that, with
the analysis and heat treatments used, approximately 0.45 per cent
total manganese would be required in the metal before combined
carbon would be present as a separate metallographic constituent
and, therefore, before the physical properties would become greatly
altered. Unfortunately, no samples were prepared with a manga-
nese content of this amount so that accurate location of the point
at which cementite appears as a separate metallographic constitu-
ent will require further tests.

44. For the sake of comparison, the group *A* bars (which had
not received the extra heat treatment given the bars from groups
B, C and *D*) were tested after the spheroidizing treatment and the
results plotted in Fig. 4. They were found to correspond closely
to the averages of groups *B, C,* and *D,* showing that the manga-
nese content is more important than heat treatment above the crit-
ical temperature. Photomicrographs are presented (Fig. 5 to 7)
showing the structure of the metal from the group *B* test bars
from series *1, 2* and *3.* The group *B* bars were representative of
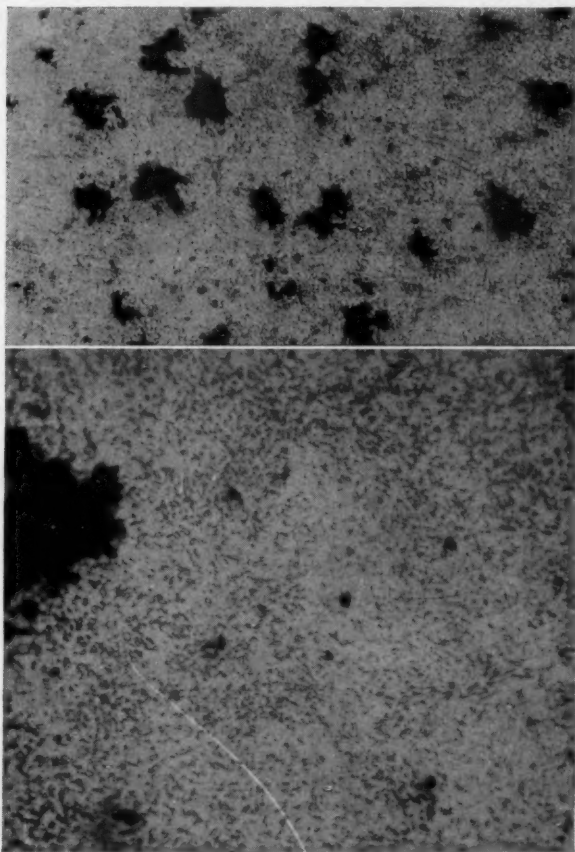
FIG. 7—SAMPLE FROM SERIES 3, GROUP B—MANGANESE CONTENT 1.00 PER CENT—NITAL
ETCH—(UPPER) 100X—(LOWER) 500X.

the metal from each of the other groups. At low magnification, the temper carbon distribution and grain size of the ferrite is shown, and at the higher magnification the size and distribution of the cementite spheroids is illustrated.

CONCLUSIONS

45.   From this investigation, it appears that, for the analysis of the metal used, the following conclusions can be stated:

(1)   After the massive cementite has been completely decomposed by prior heat treatment, the limit of solubility of combined carbon in austenite in the presence of temper carbon is increased with higher temperatures.

(2)   A variation in the manganese content of the material has little if any effect on the limit of solubility of combined carbon in austenite in the presence of temper carbon.

(3)   The combined carbon percentage remaining in the material after the spheroidizing operation, is increased with higher percentages of manganese.

(4)   The combined carbon percentage remaining in the material after the spheroidizing treatment is not affected greatly by the temperature in the austenitic range from which the material was air cooled.

(5)   With increased combined carbon content, or with increase in manganese (which is the cause of increased combined carbon content), the ultimate tensile strength and the yield strength are increased, while the percentage of elongation is decreased.

46.   The author is tempted to reach other broad conclusions which are not entirely supported by the experimental data. For example, the fact that manganese content has relatively little effect on the solubility of combined carbon in austenite indicates probably that, within reasonable limits, manganese has but little effect on the rate of graphitization at temperatures in the so-called graphitizing range.

47.   Also, because in every case the combined carbon content of the material after spheroidizing is less than the combined car-

bon content of the metal as cooled from various temperatures above
the critical, it can be concluded that the function of manganese
as a graphitization retardant is one of slowing up the graphitiza-
tion of combined carbon in pearlite at the spheroidizing tempera-
ture, which is below the critical.

48. The mechanism of spheroidization is well known. In the
case of steels, all of the combined carbon not soluble in ferrite re-
mains in spheroidized form. If, however, graphite is present, then
the combined carbon of the pearlite in the process of spheroidiza-
tion has an opportunity to migrate and precipitate to the stable
form, temper carbon. It is difficult to see how manganese could
stabilize, under equilibrium conditions below the critical, a definite
percentage of combined carbon in the form of cementite, unless it
could be shown that manganese carbide forming a double carbide,
with iron carbide, could become insoluble in ferrite, or in an amor-
phous grain boundary material. Such stability of manganese car-
bide appears doubtful in view of the fact that with manganese of
0.34 per cent, as typified in bars from series no. 1, there was no
stable carbide remaining after the spheroidizing treatment. The
author is, therefore, tempted to conclude that the mechanism by
which manganese acts as a graphitization retardant is that of
either slowing up migration of carbon atoms in either ferrite or
grain boundary material, or by slowing up the rate of solubility of
cementite in these materials.

ACKNOWLEDGEMENTS

49. The author wishes to express his thanks to P. A. Paulson
and George Minert, metallurgist and chemist, respectively, of
Gunite Foundries Corporation, for helpful suggestions in the prep-
aration of these experiments and for assistance in the chemical and
physical testing required.

DISCUSSION

Presiding: HOMER WRIGHT, Michigan Malleable Iron Co., Detroit,
Mich.

L. H. RUDESILL[1], *(Written discussion)*: The outstanding point in
Mr. Forbes' paper is the general accumulation of information, that points
out the value of intentionally producing and controlling a definite
amount of combined carbon in the conventional ferritic malleable struc-
tures. It appears that we have only scratched the surface in developing
knowledge in connection with this important subject. It is becoming ap-

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[1] Metallurgist, Griffin Wheel Co., Chicago, Ill.

parent, that any number of commercial and uniform metallurgical structures of combined carbon and ferrite can be produced by some definite operating practice. The process would consist necessarily of a material, cast white, containing a sufficient amount of carbide stabilizer, and heat treated to produce a definite combined carbon and various definite metallurgical compositions. It may also be possible to use some other carbide forming element other than manganese, such as chromium or vanadium for this purpose. The writer has had experience in producing both gray iron and pearlitic irons, with late ladle additions of ferro-chromium, and the resultant spheroidal formations of combined carbon were remarkably uniform. Mr. Forbes definitely points out, that it is surprising that the malleable metallurgists have neglected to develop information on how to produce and control combined carbon under such conditions, that will eventually result in a material of commercial value.

I wish to confirm Mr. Forbes' assertion that a manganese content up to 0.75 per cent or a surplus of about 0.6 of a per cent over that required to form manganese sulphide, actually acts as a mild graphitizer. As the manganese exceeds this point, however, there is a sharp hardening affect and graphitizing is retarded proportionally. I have found this to be true in attempting to vary the depth of chill on car wheels, where the precision of control is the closest I have ever seen.

H. A. SCHWARTZ[2] *(Written discussion)*: Mr. Forbes' contribution is to be welcomed as representing one of the first public disclosures of the operating details under which any of the many forms of pearlitic malleable are commercially produced. The author is in a position to speak authoritatively and in such detail as he chooses about this process since his corporation was one of those which first exploited this particular variety of malleable iron in which significant amounts of combined carbon were purposely retained.

In reading through the paper this commentator finds occasion now and then to add certain remarks to what the author has said. The mechanism described in paragraph 17 is certainly rather too simple to be adequate. Since graphitization represents the combination of the steps: solution of cementite, decomposition of cementite, migration of carbon, crystallization of carbon (not necessarily in this sequence so far as dissociation is concerned), the reaction as a whole will have its rate determined by that of the slowest step. Solution and crystallization are steps dependent upon area of interface, therefore, in the very beginning of the reaction, when the surface of graphitie is infinitely small or absent, that reaction must be the slowest and govern the process. Just before the cementite disappears, the area of cementite must be small and must govern the process.

Evidence has been offered elsewhere long ago[3], that during most of the reaction, the rate of migration of carbon through iron was the

governing cause. It is, however, true that graphite first appears at intermetallic interfaces. There are more of these interfaces in a "fine grained" white iron than in a coarse grained. There may, therefore, be more graphite nodules and the graphitizing rate may, therefore, be faster.

With regard to the last sentence of paragraph 24, evidence exists[4] that ferromanganese added to molten air furnace metal increases the nodule number for nodules germinated at around 1650°F. (900° C.) as compared with the unalloyed simple malleable. Forbes' sentence is, therefore, probably an under-statement of the facts.

With regard to paragraph 26, it can certainly be shown microscopically that a given spheroidized matrix decreases in number of nodules as it graphitizes. The author's conclusion that the spheroidized matrix is, therefore, not of any particular composition is certainly justifiable. Pearlite is of constant composition only because its composition represents that of austenite in equilibrium with cementite and ferrite. The author's matrices are this composition modified by subsequent graphitization.

With regard to the conclusions in paragraph 45, the first conclusion is, of course, merely a restatement of the well known direction of the $A_{G R}$ line. The second represents a valuable statement regarding the effect of a particular element on the iron-carbon system. The third is a modest statement of certain matter which the author expands perfectly justifiably in his next paragraph. The third conclusion is merely a statement that manganese retards graphitization at spheroidizing temperatures. The fourth conclusion is the equivalent of saying that no matter at what temperature graphitization first occurred, the cooling rates referred to in the paper are such as to bring the solid solution reasonably close to eutectoid composition before the spheroidizing operation has an opportunity to begin. The fifth conclusion as pointed out by the author is the logical result of considering this material as a spheroidized steel in which temper carbon has been embedded.

With regard to paragraph 46, it has been shown[5] that manganese retards graphitization much more at temperatures below the critical point than at temperatures above them. These observations entirely confirm Mr. Forbes' paragraph 46. This commentator, however, has a mental reservation to the fact that the effect of an element on graphitizing rate is not to be judged so simply as by its effect on the $A_{G R}$ line. Indeed the view expressed by the author in paragraph 46 would seem to be inconsistent with the view expressed in paragraph 17 in which he considers the fineness of structure of the white cast iron as so largely a determining factor. Both are probably of significance and, in addition, the fact that the presence of solute atoms in iron may very readily alter the rate of migration of carbon through that iron.

[4] H. A. Schwartz, H. H. Johnson and C. H. Junge, *Retarding Effect of Certain Metallic Elements on Graphitization*, Transactions of the American Society for Metals, vol. 25 (1937).

[5] Tario Kikuta, *On the Malleable Cast Iron and the Mechanism of its Graphitization*, Science Reports of the Tohoku Imperial University, vol. XV (1926).

This commentator is particularly struck and pleased with the clear way in which the thoughts of paragraph 48 have been put forward. He there definitely takes cognizance of the fact of a stranger element on migratory rate which may be a most important factor in determining graphitizing rates at spheroidizing temperatures and, therefore, a matter of grave importance in determining the composition of material suited to the process being described.

In conclusion, one would like to ask a question. It is well known that many tool steels, especially alloy steels, can be spheroidized most effectively by a slow cooling through the critical range. Has any combination of alloying elements with such slow cooling ever been tried in the production of "Z" metal, and if so, with what success?

E. Touceda[6] *(Written discussion)*: As far as the speaker's experience goes, he has always considered the Z-metal type of casting quite temperamental in respect to uniformity of structure, to secure which necessitates infinite pains in heat-treatment.

Then again, experimenting with step castings test specimens, in which the sections were ⅝, ¾ and 1 in., that a considerable difference was found in both structure and fracture, and particularly in toughness.

He has noticed, not only in this paper, but in those that have been written on the product made by the short anneal that while there are numerous tensile tests recorded for comparison, practically no attention has been paid to the matter of toughness. As an engineer, the two properties that he considers of most value is a combination of high yield and toughness to impact.

In estimating the worth of a ferrous product, therefore, he believes that the matter of toughness deserves serious consideration in an estimate of the worth of the metal for most uses.

In reading the paper, the speaker did not notice that any remarks were made in regard to corrosion resistance. In the past a considerable tonnage of bolts for joining the flanges of large water pipe have been sold and he would ask Mr. Forbes if he knows just how these stood up in respect to corrosion resistance.

The speaker believes that quite a fair size field can be developed for the Z-metal type of castings, for there are many parts, the value of which, depends in large measure upon their ability to withstand wear. In such parts where impact stresses are not too severe, the Z-metal type of casting excels. There should be a field for the castings as a substitute, for example, for cylinders for pneumatic tools that now are made of forgings, for obviously the cost of machining would be greatly lessened.

In paragraph 12 Mr. Forbes refers to possibility, when using ferromaganese, of heavy sections being mottled and in the following paragraph he refers to the use of an 80 per cent ferromanganese with a 4 per cent carbon content, and also to one of the same manganese content but with 1 per cent of carbon. The ferromanganese with which the

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[6] Consultant, Malleable Founders' Society, Albany, N. Y.

speaker has come in contact contain much more than 4 per cent of carbon when the manganese is 80 per cent.

Thermit alloys can be made with little or no carbon, while the carbon can be adjusted to certain percentages in the electric furnace, but if reference is made to blast furnace ferromanganese the speaker feels certain that the carbon will be of the order of 6.5 per cent. Even spiegeleisen of around 21 per cent manganese will contain 5.5 per cent of carbon.

The section of the paper beginning with paragraph 28, is headed "Physical Properties." If this has reference to tensile properties, it should be headed "Mechanical Properties."

All together the paper is one that the speaker is very glad to have in his files and is very enlightening in many directions.

MEMBER: Has the author any information of the effect on the ductility with progressive holding, say from two or three hours up to 30 hours, at the spheroidizing temperature? Does the elongation get progressively greater at 30 hours in his operation?

MR. FORBES: In connection with this last question is a loss in combined carbon as the spheroidizing operation continues. The structure starts out entirely pearlitic and ends up entirely spheroidized. The ductility increases naturally as the spheroidizing goes on, from two causes: (1) The spheroidizing effect, and (2) the decrease in combined carbon. But that can be counteracted by a higher manganese content to begin with, so we then have only the effects of spheroidization. And we spheroidize for 30 hours because we found a shorter time than that will very often give erratic results, particularly in ductility, and only with the full time do we feel confident it will be as ductile as it should be.

The written discussion of Dr. Schwartz is very constructive and very greatly appreciated. The descriptive method of graphitization outlined in the paper is not an attempt to explain it completely.

The malleable foundryman has a "rule of thumb" method for determining the so-called balance between manganese and sulphur. The usual rule sets the percentage of manganese at twice the percentage of sulphur plus about 0.15 per cent. In view of the fact that all the sulphur is probably not present in the form of a pure chemical compound of manganese, it is likely the reference in my paper is not of particular scientific value. It illustrates roughly to the malleable foundryman that considerable excess of manganese over that of normal malleable practice is necessary to obstruct the solubility of carbide at spheroidizing temperature and prevent complete graphitization of the carbide.

Dr. Schwartz is entirely correct in his comments on paragraph 46—namely, that manganese has other effects on first stage graphitizing rather than merely the location of the A$_{cm}$ line. However, the effect of manganese on a given size casting being more or less uniform, the foundryman can from a single base analysis of iron and a single heat

treatment obtain resultant material of widely different properties by the simple expedient of varying the amount of manganese addition.

I would like to be able to throw a little more light on Dr. Schwartz' question as to the possibility of spheroidizing by slow cooling through the critical range. This would appear entirely possible if controlled cooling could be maintained commercially on large ovens. Unfortunately, we have had no experience in our laboratory on this point.

In connection with Professor Touceda's remarks, there has been quite a bit of work done on the corrosion of Z-metal bolts, but apparently there is no set definition of what constitutes corrosion resistance. In other words, corrosion in acid mine water is one thing, and corrosion due to exposure to salt air is another. In some reports I have received, based on loss of weight of bolts, assuming the loss of weight of a Z-metal bolt was unity, the loss of weight of a normal malleable bolt was 3, and a steel bolt was 7. Now that statement should be qualified by stating that the Z-metal bolt contained three-quarters of one per cent copper, which probably had some value in corrosion resistance, while the malleable iron and steel, as far as we know, contained no copper.

As to the type of ferromanganese we used, it was electric furnace ferromanganese. I know for heavy section castings we pay a premium for the very low carbon ferromanganese to avoid graphitization in the heavy sections. Normally, we use alloy containing about 1.50 per cent C.

Effect of Blast Pressure in Cupola Melting

J. T. MacKenzie* and C. K. Donoho*, Birmingham, Ala.

Abstract

The authors report on an investigation of the effect of varying blast pressures on cupola melting and on the properties of the resulting iron. Separate heats were made in a 21-in. test cupola using a wide range of blast pressures and two different kinds of metal charges. Results indicate that blast pressure may be varied widely to change the melting rate, and that, except for slight changes in analysis, the melted iron will show no undue variations from normal.

1. In most cupola operations, the blast pressure under which a cupola is operated, is varied to control the melting rate of the cupola. Melting rate is dependent primarily, of course, on the weight of oxygen passing through the cupola in a given time, which is approximately proportional to the volume of air per unit time.

2. This series of test heats was made to determine the effect of varying blast pressure on melting and on the properties of the resulting iron while other conditions were maintained as nearly constant as possible.

Equipment

3. The 21-in. inside diameter cupola used in these tests, is essentially the same in construction and operation as the test cupola previously described[1, 2]. Air was supplied by a variable-speed,

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[1]Johnson, H. V., and MacKenzie, J. T., *"Effect of Tuyere Height on Carbon Pick-up in the Cupola,"* Transactions American Foundrymen's Association, 1936, vol. 44, pp. 178-192.

[2]Bowers, J. A., and MacKenzie, J. T., *"Cast Iron as Affected by Coke Size in Cupola Melting,"* Transactions American Foundrymen's Association, 1937, vol. 45, pp. 293-319.

* Chief Metallurgist and Assistant Metallurgist, respectively, American Cast Iron Pipe Co.

Note: Presented before Gray Iron Session 42nd Annual Convention, Cleveland, O., May 16, 1938.

positive-displacement blower and controlled by a rheostat and bleeder valve. It was found that, due to the small amounts of air used, blast volume could not be measured accurately by the instruments available. Pressure was measured at the wind pipe just before entering the tuyeres by a U-tube mercury manometer calibrated in ounces. Iron temperatures were measured by a Standardized Leeds & Northrup disappearing filament optical pyrometer.

RAW MATERIALS

4. All coke used in these tests was taken from a single car of foundry coke and screened to a uniform size. The coke used was that which passed through a 3-in. screen and remained on a 1½-in. screen, so that the average size of the coke was 2¼-in. Analysis of the coke is as follows: Volatile matter, 1.19 per cent; fixed carbon, 89.23 per cent; ash, 8.5 per cent; moisture, 0.40 per cent; sulphur, 0.614 per cent; shatter test-retained on 2-in. screen, 81 per cent; cell space, 46.8 per cent.

5. One series of heats was made using iron charges composed of slabs broken from cast iron pipe of about ½-in. wall thickness. Average size of the slabs was about 8x4-in. Chemical uniformity of the metal charged was controlled by analyzing each pipe used and allowing a maximum of 20 points variation in silicon or carbon in the metal used for any heat. The proper quantity of iron for each heat was mixed thoroughly before weighing up the charges, and the average analysis of the metal charged was determined by averaging individual analyses of the pipe used and also by analyzing a composite sample.

6. Analyses of the scrap charges used were for all heats within the following limits: Total carbon 3.35 to 3.53 per cent; silicon 1.28 to 1.50 per cent; sulphur 0.06 to 0.07 per cent; manganese 0.39 to 0.45 per cent; phosphorus 0.58 to 0.72 per cent.

7. Two heats were run using silicon steel pigs made in an electric furnace. Analyses of these charges were within the following limits: Total carbon: 0.58 to 0.76 per cent; silicon 1.71 to 1.78 per cent; sulphur 0.040 to 0.042 per cent; manganese 0.74 to 0.76 per cent; phosphorus 0.03 to 0.05 per cent.

8. A dolomitic limestone composed of about 54 per cent $CaCO_3$ and 44 per cent $MgCO_3$ was used as the flux.

Procedure

9. Unscreened coke which averages about 4-in. in size was used in the bed to 12-in. above the tuyeres. The bed was built up to 40-in. with the sized coke, burned through, then built up to 50-in. with sized green coke.

10. Iron charges were 125 lb. where pipe scrap was melted and 100 lb. where the silicon steel was used. Coke splits were in all cases 12.5 lb. of the 2¼-in. sized coke. Four pounds of dolomite were added with each coke charge. Total metal charged per heat was 1200 to 1250 lb. which was ten charges of pipe scrap, or 12 charges of steel pig.

11. Charging was complete before the blast was put on, and the stack was kept full with coke during the heat to maintain the pressure-volume ratio of the air blast.

12. For each heat, the blast pressure was kept constant throughout the heat at the selected value for that heat. The melted iron was caught in seven to ten approximately equal ladles of about 135 to 150 lb. each. Two chill specimens of 1½x1½-in. and 1½x3-in. section, were cast from each ladle and three sets of four American Water Works Association standard 2x1x26-in. test bars were cast from the 3rd, 5th, and 7th ladles.

13. Separate heats were run on 3, 5, 7, 10, and 12 oz. blast pressure when cast iron pipe scrap was used. Heats on 6 and 12 oz. pressure were run when silicon steel pigs were melted.

Test Results

14. Melting temperatures and chemical analyses of each ladle of each heat were obtained and recorded. Total blowing time for each heat was recorded. The three sets of four American Water Works Association standard test bars poured from each heat were tested transversely, and breaking load and deflection were recorded.

15. Fig. 1 shows the temperature data recorded for the pipe scrap heats. These readings were made on the stream at the end of the iron trough, which was 5 ft. long. It was found that more accurate and comparable readings could be made at this point than at any other. However, due to the length of trough and the small trickle of iron, these readings are found to be 75 to 100°F. lower than actual melting temperatures. In obtaining averages, the data recorded on the first two ladles were omitted because it was felt

Blast Pressure- Oz.	Symbol	Avg. Temp. (Omitting Ladles 1 & 2)
3	□	2530 Degrees Fahr.
7	•	2605 " "
10	×	2660 " "
12	△	2690 " "
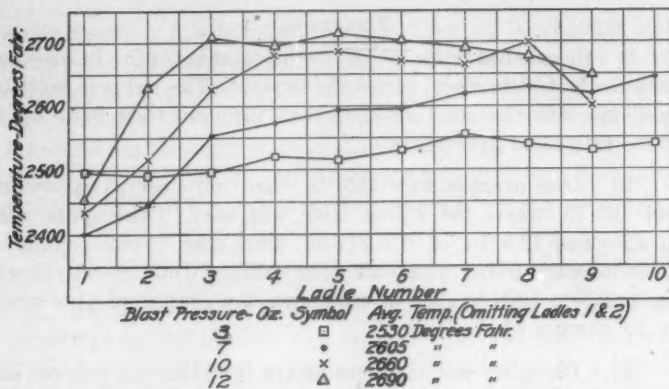
FIG. 1—EFFECT OF BLAST PRESSURE ON STREAM TEMPERATURE IN PIPE SCRAP HEATS.

Table 1

EFFECT OF BLAST PRESSURE ON MELTING OF PIPE SCRAP

	Blast Press. Oz.	1	2	3	4	Ladle No. 5	6	7	8	9	10	Average Omitting 1 and 2
Stream Temperature (°F.)	3	2495	2490	2495	2520	2515	2530	2555	2540	2530	2540	2530
	5	......	......	......	......	......	......	......	......	......	......	......
	7	2400	2445	2555	2575	2595	2595	2620	2630	2620	2645	2605
	10	2480	2515	2625	2680	2685	2670	2665	2700	2600	......	2660
	12	2455	2630	2710	2690	2715	2705	2675	2650	......	......	2690
Carbon Pick-up (Per Cent Carbon)	3	0.37	0.30	0.10	0.18	0.20	0.09	0.11	0.08	0.14	0.11	0.126
	5	0.26	0.18	0.12	0.08	0.06	0.05	0.06	......	......	......	0.074
	7	0.15	0.21	0.17	0.13	0.05	0.05	0.05	0.05	0.05	0.03	0.064
	10	0.13	0.16	0.13	0.08	0.07	0.09	0.08	0.12	0.08	......	0.093
	12	0.35	0.29	0.11	0.07	0.04	0.01	0.04	0.02	0.00	......	0.039
Sulphur Pick-up (Per Cent Sulphur)	3	0.067	0.070	0.070	0.070	0.078	0.070	0.055	0.066	0.056	0.065	0.066
	5	0.046	0.054	0.058	0.055	0.052	0.054	0.054	......	......	......	0.055
	7	0.040	0.049	0.054	0.057	0.057	0.058	0.049	0.052	0.048	0.049	0.052
	10	0.072	0.065	0.056	0.061	0.065	0.058	0.059	0.048	0.050	......	0.056
	12	0.034	0.044	0.048	0.044	0.048	0.038	0.042	0.051	0.050	......	0.044
Per Cent Loss of Manganese	3	32	34	14	17	27	14	10	17	12	5	15
	5	34	29	22	19	22	22	19	----	----	----	21
	7	35	26	22	16	13	14	19	17	16	14	16
	10	37	27	20	21	14	12	16	14	16	----	16
	12	27	24	20	18	16	20	18	13	20	----	17
Per Cent Loss of Silicon	3	0.6	2.7	6.7	5.3	8.0	6.7	8.0	9.5	7.8	7.8	7.1
	5	4.6	3.1	6.2	6.2	7.0	9.3	9.3	----	----	----	7.6
	7	3.8	3.8	6.5	6.5	8.0	8.0	5.8	11.0	10.4	9.9	8.5
	10	9.0	7.7	9.7	9.9	10.4	10.6	12.5	9.3	9.6	----	10.3
	12	1.3	2.7	6.1	6.1	9.5	10.3	11.5	13.5	16.2	----	10.5

that the difficulty in measuring the bed height correctly for each heat and also the difficulty of collecting the same amount of iron in the basin before tapping out would prevent true comparisons of these first ladles.

16.   Table 1 shows numerically the data for temperature as plotted in Fig. 1 as well as the data for the analysis changes studied. Fig. 1 is shown merely as an example of the variation between ladles and similar charts for analysis changes will not be given.

17.   Fig. 2 shows the average values for melting rate, temperature and analysis changes plotted against blast pressure. The blast pressure scale is laid out with the intervals proportional to the square roots of the pressures, which is equivalent to a uniform scale for volume. Melting rate is calculated from the total blowing time and includes the time of blowing in. Carbon and sulphur changes are reported as the actual amounts of pick-up during melting, as these are the terms in general use among foundrymen.
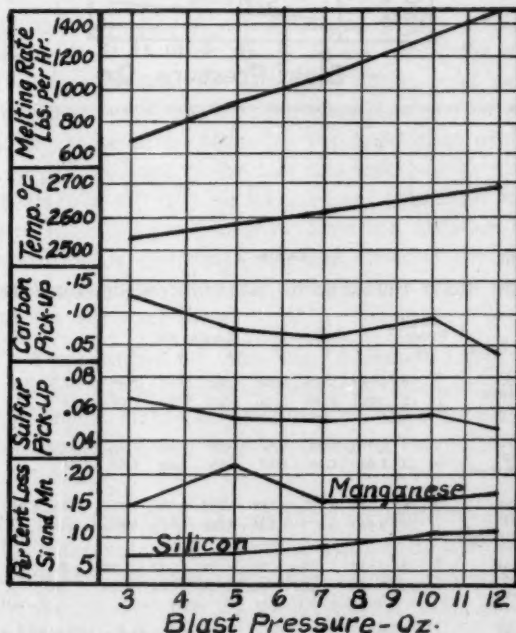
Fig. 2—Effect of Blast Pressure on Melting of Pipe Scrap.
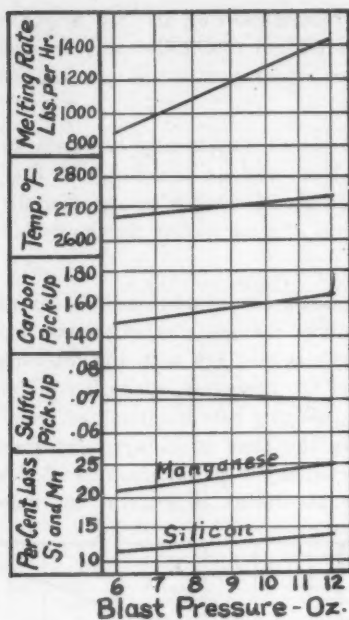
FIG. 3—EFFECT OF BLAST PRESSURE ON MELTING SILICON STEEL PIGS.

Table 2

EFFECT OF BLAST PRESSURE ON MELTING OF SILICON STEEL PIGS

	Blast Press. Oz.	1	2	3	Ladle No. 4	5	6	7	8	Average Omitting 1 and 2
Stream Temperature (°F.)	6	2600	2615	2660	2665	2680	2680	2695	2670	2675
	12	2690	2695	2720	2740	2730	2730	2720	....	2732
Carbon Pick-up (Per Cent Carbon)	6	1.80	1.65	1.42	1.40	1.44	1.34	1.40	1.64	1.44
	12	2.05	1.70	1.87	1.58	1.40	1.65	1.68	......	1.68
Sulphur Pick-up (Per Cent Sulphur)	6	0.114	0.101	0.082	0.082	0.082	0.070	0.074	0.060	0.078
	12	0.088	0.077	0.084	0.068	0.062	0.074	0.065	......	0.070
Per Cent Loss of Manganese	6	22	19	20	20	20	21	21	22	21
	12	26	23	24	24	24	25	30	....	25
Per Cent Loss of Silicon	6	6.4	9.0	11.5	13.0	14.0	11.5	10.0	10.0	11.7
	12	1.7	9.0	13.0	13.0	16.0	14.0	15.5	......	14.3

Silicon and manganese changes are expressed as percentage loss during melting for the same reason.

18. Table 2 and Fig. 3 give similar data for the two heats where the silicon steel pigs were melted.

19. To evaluate the effect of blast pressure on physical properties, the average values for each heat of the breaking load of the bars, the deflection at 1900 lb. (inversely proportional to stiffness $\dfrac{3,280,000}{\text{Def. (at 1900 lb.)}} = E$ at 1900 lb.), and the Brinell hardness (determined from sections of the broken bars) were plotted against the blast pressures. Since there was some variation in the analysis of the metal in the different heats, carbon equivalent (carbon plus 0.3 of the silicon and phosphorus) was plotted on the same chart, Fig. 4. It is to be noted that the scales for breaking load and hardness are reversed to bring out the relation between the curves.

Discussion of Results

20. Melting rate, of course, increased with increasing blast, the relation being approximately linear when the square root scale for blast pressure is used. The silicon steel heats showed somewhat slower melting for the same blast pressures than did the higher carbon scrap heats for the same blast pressures. This is due, in part at least, to the fact that iron to coke ratios for the scrap heats were 10 to 1 or 9.1 per cent coke, and for the steel heats 8 to 1, or 11.1 per cent coke. If carbon absorbed by the iron during melting is deducted, the coke effective for melting and superheating is about 9 per cent coke for the scrap heats and 9.8 per cent for the steel heats.

21. Temperature of the iron increased with increasing blast pressure in the ranges investigated. The silicon steel heats showed distinctly higher temperatures for the same blast pressures than did the scrap heats, largely because of the difference in coke ratios as noted.

22. The data from the scrap heats showed carbon pick-up generally decreased with increasing blast pressures, while in the heats using the low carbon steel the carbon pick-up was greater with the higher pressure. It would appear that, with the steel charges, melting temperature was the predominant factor which

more than nullified the effect of increased oxidation and more rapid melting caused by the higher blast pressure.

23.   Sulphur pick-up was practically unaffected by changing blast pressure.

24.   The percentage loss of manganese and silicon showed generally slight increase with increasing blast pressure.

25.   The rather remarkable correlation of properties with analyses, as shown in Fig. 4, indicate definitely that blast pressure *per se*, within the ranges investigated, has no effect upon the physical properties tested.

26.   Although these results strictly apply only to the particular conditions of the tests, they indicate that in ordinary cupola operation blast pressure may be varied within a wide range to vary the melting rate, and that, except for some slight changes in analysis, the melted iron will show no undue variations from normal.
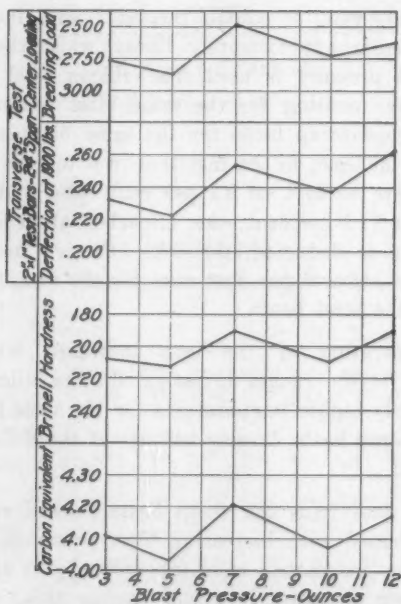
FIG. 4—RELATION OF CARBON EQUIVALENT AND PHYSICAL PROPERTIES OF PIPE SCRAP MELTED UNDER DIFFERENT BLAST PRESSURES.

DISCUSSION

Presiding: G. P. PHILLIPS, Tractor Works, International Harvester Co., Chicago.

CHAIRMAN PHILLIPS: Mr. Donoho, was there any difficulty in holding the blast pressure constant during your test, and just how was that accomplished?

MR. DONOHO: In these tests there wasn't any particular variation. We put all the charges in before turning on the blast. After putting in the last charge the stack was kept full of coke in order to prevent a change in the pressure/volume relationship. The Manometer was watched all the time so that if there was any change we corrected it, but there was very little necessity for doing so.

MEMBER: Where was the blast pressure taken and what size tuyeres were used?

MR. DONOHO: The blast pressure was measured by a manometer placed just before entering the tuyeres. There were two 4-in. round tuyeres on the cupola.

MEMBER: Why was not a higher range of pressure used, which would be more indicative of the operating range most of us are familiar with?

MR. DONOHO: It is true that on larger cupolas the blast pressure generally used are much higher, but on the smaller cupolas one has difficulty using as high blast pressure as on the larger cupolas. I know this is true of the attack on the lining. We have found it difficult for various reasons to operate a small cupola at a much higher blast pressure than 12 ounces.

MEMBER: Is it possible, by adjusting any of the factors of cupola operation, to obtain as high a temperature iron with 3 ounces of pressure as compared with the normal operating pressure of 16 ounces?

MR. DONOHO: I do not know. There are other factors to be considered. The variable of coke size has some effect on temperature. It might be that with a larger coke and a lower blast pressure you could obtain the same temperature you could with higher blast pressures.

CHAIRMAN PHILLIPS: It would probably be dependent upon the burning rate of the coke. It would be possible to take a very hard, dense coke with a slow burning rate and make a hot iron. The statement that the melting rate is probably dependent upon the amount of oxygen going through is probably true for coke that will respond to the velocity of that oxygen in combination in the particular cupola in question. However, there might be times when you had a coke that would not take a high velocity of oxygen. If you try to give it more air you are going to reduce the temperature rather than increase it, and also reduce the melting rate.

MEMBER: Was there any determination made on the weight of air or were the conditions such that the pressure was proportional to the weight of air introduced into the cupola.

MR. DONOHO: We did not have instruments available to measure the weight and volume of air. We had a volume meter which we thought would serve and found it would not record the low volumes on the small cupola. We tried to do the best we could by varying only blast pressure and controlling the other conditions as closely as possible.

MEMRER: In Figure 3, as the blast pressure increased, the carbon pick-up increased. Is that due to an increase in the temperature of the metal in that particular series?

MR. DONOHO: It is stated in the paper, "It would appear that, with the steel charges, melting temperature was the predominant factor which more than nullified the effect of increased oxidation and more rapid melting caused by the higher blast pressure." Just why we had an increased carbon pick-up with an increase in blast pressure in the steel heats and a decrease in carbon pick-up with an increase in blast pressure in the scrap heats, I am unable to say. There are several factors that may take precedence one over the other under varying conditions.

MEMBER: If I understood the author correctly, he came to the conclusion despite the varietion in temperature, that the difference in physical properties can be all accounted for by variations in composition, is that correct?

MR. DONOHO: Yes, within the ranges mentioned in the paper. The pouring temperature of these bars was approximately 2450°F.

MEMBER: The author claims he does not get any variation in physical properties due to temperature, while Messrs. Young, Crosby and Herzig claim [1] they get a wide variation in physical properties due to temperature.

MR. DONOHO: We did not have in any case enough super-heat to have a marked effect on the physical properties. Probably, if we had gone on to 100°F. higher we would have found the effect Messrs. Young, Crosby and Herzig found.

CHAIRMAN PHILLIPS: What would happen if the pressure was increased very much? Would free oxygen be forced up through the metal and create oxidation? The maximum pressure used in this particular case was 12 ounces.

MR. DONOHO: That remains to be proved. Our conclusions can only apply to the ranges investigated. However, it is a question as to whether increased oxidation of the metal would change the physical properties in a manner that would not show up in the analysis. As far as we went, there was no effect on the physical properties that was not shown by analysis.

MR. MACKENZIE: While these tests were run on a 21-in. cupola, we have in practice in Birmingham a 72-in. by 102-in. cupola running

[1] *Physical Properties of Cast Iron in Heavy Sections.*
Paper presented before 1938 A. F. A. Convention.

from 8 to 35 tons an hour. One of the reasons we ran this test in the little furnace was that it could be carefully controlled. We do not have to change our mixture when we make those tremendous changes in blast pressure to get the same analysis as in the shop. In tonnages varying from 12 to 40 tons an hour with the same metal charge, we get approximately the same analysis. In regard to increasing the blast pressures on the small cupolas, the coke would simply blow out. But on large cupolas, pressures up to 40 ounces would not be unreasonable at all, especially since large coke is used.

CHAIRMAN PHILLIPS: Mr. MacKenzie, do you believe that your carbon pick-up would not change regardless of whether you were melting at a low rate or an extremely high rate?

MR. MACKENZIE: Yes, however, a higher temperature is gotten with the higher blasts and a higher coke bed. Naturally, there will be an increase in carbon pick-up with a large amount of steel. Carbon pick-up does not show particularly on steel of about 25 per cent. If a large amount of fine, light scrap is used, there might be trouble with oxidation. The oxygen content of the blast will go very much higher in the stack in melting thin stock than heavy stock and thus cause more loss. However, the carbon absorption on what does melt will not be very different and a little in favor of the higher temperature.

L. L. ANTHES: [2] A great deal of discussion during various sessions last year and this year state than 16 ounces is a normal pressure on cupolas. Yet, in listening to the argument this afternoon, there seems to be a wide range of blast pressures. A certain cupola of a certain diameter is supposed to melt so many tons an hour. Presumably it requires 30,000 cubic feet of air to melt each ton. I would like to know just on what basis that 16 ounces of pressure is taken as normal.

MR. MACKENZIE: That pressure was true of the old behive cokes. You could not blow them very much above 16 ounces. The modern by-product cokes will stand almost an unlimited amount of pressure. I have used pressures up to 35 ounces and have not gotten the trouble from oxidation I had with the old beehive cokes at 16 to 20 ounces.

MR. ANTHES: For the 16-ounce normal blast pressure, Mr. Mac-Kenzie refers to beehive coke, but there is very little of it used in the north today. I read in a recent issue of The Foundry, an article on cupola practice where the reference again was made to 16 ounces as normal pressure. Evidently, however, from what we have heard this afternoon, that pressure varies with the different diameters of cupolas.

CHAIRMAN PHILLIPS: Is it not true that pressure will vary with the different types of charge? If a type of material that would tend to close up in the cupola is used, our resistance would go up and our pressure against which the fan is operating would increase.

MEMBER: In the small cupola, the coke would be blown out with the high pressure, and I think it is true that there is a normal pressure

² Anthes Foundry, Ltd., Toronto, Ont., Canada.

for each size cupola. Now, take a large cupola like Mr. MacKenzie has at Birmingham and use an enormous pressure. The velocity of that air up through the stack would not be equivalent (in that size cupola) to what it would be in a 56-in. cupola with the same pressure and volume of air going through. I believe when we talk about a normal pressure, we mean in each case for a certain size cupola.

MR. DONOHO: We have a 42-in. cupola in which the pressure can be about 12 ounces as a maximum without undue attack of the lining. Whereas, with the large cupola (72 inches by 102) that Mr. MacKenzie mentioned, we can run up to 35 ounces with no particularly severe attack on the lining.

MEMBER: If an operator runs a cupola with a normal pressure or a normal wind volume and wants to cut his volume down in order to melt slower, is it necessary for him to add to his bed in order to keep the same temperature?

MR. DONOHO: I do not know about keeping the same temperature. However, it can be done very successfully, as Mr. MacKenzie mentioned, without undue change in analysis or physical properties. There may be some slight loss in temperature if the blast pressure is decreased severely. We have a rule which states that we cannot decrease or increase the blast pressure more than two ounces in 15 minutes. If the change is taken slowly, we do not find any bad effects on the physical properties.

MR. MACKENZIE: Blast pressure change depends upon whether you are talking about changing it during a heat or getting set for the heat. If you know you are going to run a low blast pressure tomorrow, it is economical and perfectly feasible to cut your bed accordingly. But if you run into a breakdown, you cannot go into the cupola and bail out any coke. You do run a little soft iron usually when you have to cut the wind suddenly. If you raise the wind suddenly you get into all kinds of trouble.

Another point which Mr. Anthes brought up is the normal pressure of 16 ounces. Because of the square root law, you would have to go to 64 ounces to put twice as much wind in the cupola as you would at 16 ounces. Therefore, in going from 32 ounces to 64 ounces you only get 40 per cent more wind, and it is not economical to pump air at two pounds pressure if you can pump it at one pound pressure. If I was designing a cupola that was not to run at excessively high temperatures. I would design one to run at a normal pressure of something like 12 to 16 ounces.

Effect of Cupola Practice on Linings

By C. K. Donoho,* Birmingham, Ala., and
Chas. F. Greene,** Macon, Ga.

Abstract

This paper reports the findings of an investigation on the effect of blast pressure, nature of charge, length of heat, and diameter of cupola on consumption of refractories in the melting zone. The details of the method of investigation are described and the findings are discussed as to the effect of blast pressure, silicon content and length of heat. The results indicate that high blast pressures, low silicon irons, long heats and small diameter cupolas are factors which give rise to excessive attack on the lining of the melting zone.

Introduction

1. Cupola linings are subject to attack by physical, chemical, and mechanical actions.[1] The physical action to which linings are subjected are the high temperatures, 3,000°F. and over, and the temperature changes, often sharp enough to cause spalling. The chemical action of slag and of the hot blast against the lining is a major factor. Linings are subject to mechanical attack by the rubbing action of the descending charges and by impingement of small particles carried upward by the blast.

2. It is only the melting zone area which is subjected to all three types of attack. The hearth, below the tuyeres, must withstand temperature and some chemical action, the stack above the melting zone must withstand severe abrasion and some temperature changes, but in the melting zone area all three types of attack act together.

3. In most foundries, the cupola lining as a whole lasts indefinitely, but the melting zone area must be relined partially or patched after each heat.

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* Metallurgist, American Cast Iron Pipe Co.
** Superintendent, Atlantic Refractories Co.
[1]MacKenzie, J. T., *"Effect of Variations in Cupola Practice on the Life of the Refractory Blocks,"* Journal, American Ceramic Society, 1925, vol. 8, no. 11 p. 720.

Table 1

PROPERTIES OF TEST BRICK

Sample (Designation)	A & B	C	D
SiO_2 (Per Cent)	64.0	66.0	14.0
Al_2O_3 (Per Cent)	30.0	27.0	84.0
Fe_2O_3 (Per Cent)	3.0	3.0	0.5
Fusion Point (PCE)	30-31	29	37
Absorption (Per Cent)	5.8	2.9	16.5
App. Porosity	12.7	5.6	30.5

4. A suitable refractory for melting zone service should have, in addition to ability to withstand high temperatures, density, toughness, and spalling resistance. Density increases resistance to chemical action but usually increases tendency to spalling. Toughness is essential for resisting mechanical attack.

5. Fig. 1 shows four brick each of which was subjected to chemical action of molten iron and slag for an 8 hr. period. The brick were placed just over the breast of a 72x102-in., continuous-flow, front-slagging cupola used in production. This cupola was melting at a fairly constant rate of about 25 tons per hr. so that each brick was subjected to about the same action.

6. Comparison of the condition of the brick of Fig. 1 with the properties of these test brick as listed in Table 1, reveals how absorption and apparent porosity correlate with the chemical action of the slag. Examples A and B of Fig. 1 are two tests of brick of the same composition showing reproducibility of results.
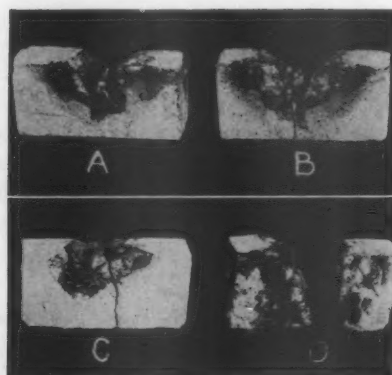
FIG. 1—TEST BRICK SUBJECTED TO ACTION OF IRON AND SLAG FOR 8 HR.

STANDARD LINING PRACTICE

7. At the plant of the American Cast Iron Pipe Company, whence the subsequent data were obtained, standard lining practice is as follows: On the day following a heat, the frozen slag and iron is chipped out in the melting zone area from the tuyeres upward 3 or 4 ft. until a fresh firebrick surface is exposed. Nine-inch straights and splits are laid-in as required to build the lining back to the original dimensions. A thin slurry of firebrick grog and plastic fireclay is used to cement the joints.

THE TEST BRICK

8. The brick, used during the period in which these data were collected, were of the composition of brick *A* and *B* in Table 1. The number of straights and splits required to patch the melting zone for each heat were recorded and are here expressed as 9-in. equivalents which is the number of the 9-in. straights used plus one-half the number of 9-in. splits used. This value is an expression of the amount of refractories consumed during the heat.

9. The flux used was dolomite which is composed of approximately 54 per cent $CaCO_3$ and 44 per cent $MgCO_3$, which was added with the coke to the amount of about 3 per cent of the metal charge. Iron to coke ratios averaged about 10 to 1.

10. No data were used in which the practice varied essentially from the standard.

EFFECT OF BLAST PRESSURE

11. Table 2 and Fig. 2 show results obtained from data on 72x102-in. link-shaped cupolas running 8-hr. heats in which the average melting rate was varied from about 15 to 35 tons per hr. These cupolas were melting a very constant mix of about 40 per

Table 2

EFFECT OF BLAST PRESSURE ON REFRACTORIES CONSUMPTION

72x102-IN. CUPOLAS

Av. Blast Pressure (Oz.)		11-12	13-14	15-16	17-18	19-20	21-22	23-24
Heats — (Number)		4	17	19	12	12	5	4
Refractories	Av.	168	253	269	327	341	357	443
Consumed	Max.	200	380	360	390	430	440	560
9-in. Equivalents	Min.	130	150	210	260	220	290	340
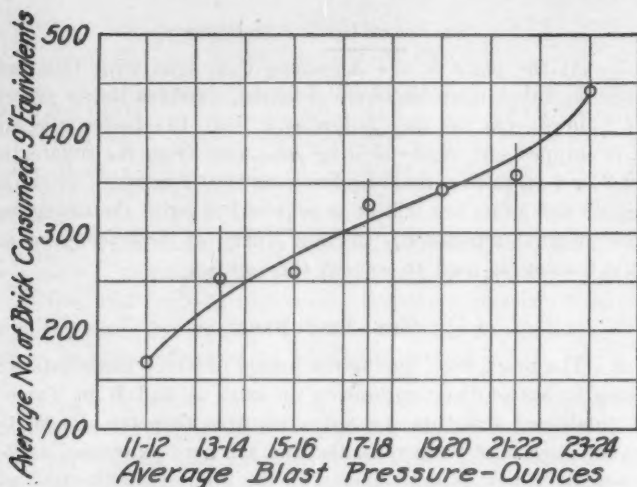
FIG. 2—EFFECT OF BLAST PRESSURE ON REFRACTORIES CONSUMED—72x102-IN. CUPOLAS.

cent pig iron, 20 per cent return scrap, and 20 per cent steel scrap so that the melted iron analyzed; carbon, 3.25 to 3.45 per cent; silicon, 1.25 to 1.45 per cent; sulphur, 0.06 to 0.07 per cent: manganese, 0.45 to 0.55 per cent and phosphorus, 0.50 to 0.60 per cent.

12. Flux and coke were approximately constant so that the variation in refractory consumption safely can be assumed to be due to changes in blast pressure.

13. Table 3 and Fig. 3 show similar correlation of data from 42-in. cupolas running 8-hr. heats. In these cupolas, the mix was varied more widely, as will be shown in the following section.

EFFECT OF SILICON CONTENT

14. An attempt was made to correlate refractories consumption with the percentage of steel used in the charges. It was found,

Table 3

EFFECT OF BLAST PRESSURE ON REFRACTORIES CONSUMPTION

42-IN. CUPOLAS

Av. Blast Pressure (Oz.)		3	4	5	6	7	8	9
Heats (Number)		4	23	24	13	15	5	1
Refractories	Av.	98	105	107	120	133	147	190
Consumed	Max.	110	135	150	155	160	230	190
9-in. Equivalents	Min.	80	80	75	80	110	115	190
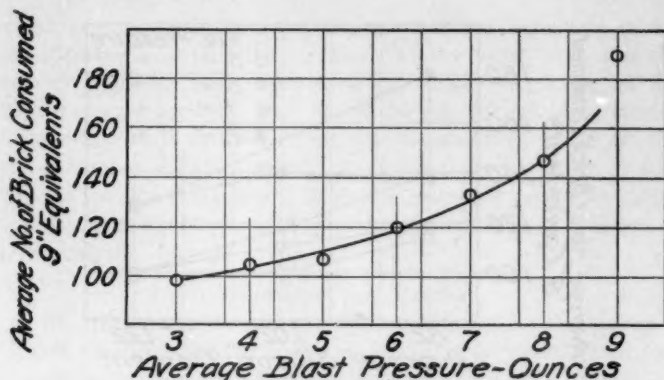
FIG. 3—EFFECT OF BLAST PRESSURE ON REFRACTORIES CONSUMED—42-IN. CUPOLAS.

however, that better correlation could be obtained with silicon content of the resulting iron melted. In several cases where high percentages of steel were used with high silicon pig iron, lining attack was not excessive; however, when low silicon pig iron was used with the same amount of steel, the attack was more severe.

15. Examination of the chemical analyses of the heats recorded in Table 3 showed that the silicon varied widely from day to day as the mix was changed to meet various shop conditions. Table 4 shows the refractory consumption of these heats when separated into three approximately equal ranges of silicon contents. Fig. 4 is a graphical representation of these data.

Table 4
EFFECT OF BLAST PRESSURE AND SILICON CONTENT OF IRON UPON REFRACTORIES CONSUMPTION — 42-IN. CUPOLAS

Av. Blast Pressure (Oz.)		3	4	5	6	7	8	9
		0.70 TO 0.94 PER CENT SILICON						
Tests (Number)		2	8	4	4	3	2	1
Refractories Consumed 9-in. Equivalents	Av.	95	114	119	139	138	180	190
	Max.	110	135	140	155	140	230	190
	Min.	80	100	90	125	135	130	..190
		0.95 TO 1.09 PER CENT SILICON						
Tests (Number)		2	10	8	6	4	......	......
Refractories Consumed 9-in. Equivalents	Av.	100	100	111	117	140	......	......
	Max.	100	125	140	140	160	......	......
	Min.	100	80	85	90	125	......	......
		1.10 TO 1.35 PER CENT SILICON						
Tests (Number)		......	5	12	3	8	3	......
Refractories Consumed 9-in. Equivalents	Av.	......	99	100	100	127	125	......
	Max.	......	105	150	110	150	130	......
	Min.	......	85	75	80	110	115	......
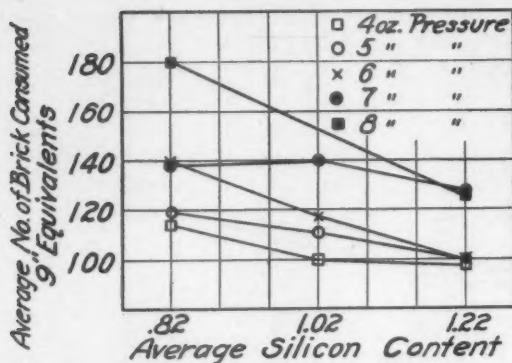
FIG. 4—EFFECT OF SILICON IN IRON ON REFRACTORIES CONSUMED—42-IN CUPOLAS.

EFFECT OF LENGTH OF HEAT

16. Some data were available on refractory consumption of 16-hr. heats compared to 8-hr. heats when other conditions were approximately the same. Table 5 lists these comparisons.

EFFECT OF CUPOLA DIAMETER

17. Because of difference in iron mixes, blast pressures, and tuyere design, there are no directly comparable data to determine the effect of cupola diameter upon lining attack. Observations of

Table 5

Length of Heat	Size Cupola, Inches	Av. Blast Press., Ounces	Silicon Per Cent	No. of Heats	Refractories Consumed 9-in. Equivalents Average	Ratio 16 hr. heat 8 hr. heat
16	72x102	13-14	1.25-1.45	3	487 ⎱	
8	72x102	13-14	1.25-1.45	17	253 ⎰	1.9
16	72x102	15-16	1.25-1.45	1	460 ⎱	
8	72x102	15-16	1.25-1.45	19	269 ⎰	1.7
16	72x102	17-18	1.25-1.45	3	363 ⎱	
8	72x102	17-18	1.25-1.45	12	327 ⎰	1.1
16	72x102	19-20	1.25-1.45	3	471 ⎱	
8	72x102	19-20	1.25-1.45	12	341 ⎰	1.4
16	72x102	21-22	1.25-1.45	1	560 ⎱	
8	72x102	21-22	1.25-1.45	5	357 ⎰	1.5
16	60 dia.	15-18	1.20-1.40	14	480 ⎱	
8	60 dia.	15-18	1.20-1.40	7	437 ⎰	1.1
16	60 dia.	7-10	0.95-1.15	6	488 ⎱	
8	60 dia.	7-10	0.95-1.15	12	411 ⎰	1.2

depth of burning in, however, indicate that the larger cupolas may
be operated economically on much higher blast pressures than
smaller cupolas. For example, 42-in. cupolas operating at 8 or
9 oz. pressure for 8 hr. have shown linings burned in as much as
6 in. deep, while the 72x102-in. cupolas operating at 24 oz. show
somewhat less depth of attack. In Table 5 are examples of 60-in.
cupolas requiring as many brick for patching as 72 by 102-in.
cupolas operating under nearly the same conditions of pressure and
iron analysis.

DISCUSSION OF RESULTS

18.   Numerous investigators[2] have referred to the violent at-
tack of fireclay refractories by iron oxide. Any factor which tends
to increase the amount of iron oxide in the slag or in contact with
the lining will tend to increase the degree of attack of the lining.
It is for this reason that lining attack is augmented by increasing
blast pressure as, obviously, higher blast pressure causes greater
degree of oxidation.

19.   Silicon in the charge is oxidized by the blast to an extent
of, roughly, 10 per cent of the total content. Since silicon is more
readily oxidized than iron, high percentages of silicon will tend
to protect the iron from oxidation and prevent the formation of
the corrosive iron oxide. Therefore, is is not surprising to find that
with equal blast pressures, the melting of lower silicon irons causes
greater attack of the lining.

20.   Rate of lining attack is greatest at the beginning of the
heat, as the burning of the lining increases the effective diameter
of the cupola, causing the air blast to be "softer" in the melting
zone next to the lining and the rate of lining attack to be progres-
sively less. The data given shows, in every case, greater attack
of the lining in the 16-hr. heats than in the 8-hr. heats, but in no
case is the ratio as great as two to one.

21.   As cupola diameter decreases, obviously the impact of
the jets of air from the tuyeres on the lining will be greater. This
causes greater severity of lining attack in a small cupola operating
under the same blast pressure as a larger cupola.

22.   Variations in practice markedly affect attack on cupola
linings. The evidence given indicates that high blast pressures, low
silicon irons, long heats, and small diameter cupolas are factors
which give rise to excessive attack of the lining of the melting zone.

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[2]Kuechler, A. H., "Influence of Fe_2O_3 and TiO_2 on Pure Clays," JOURNAL AMERICAN
CERAMIC SOCIETY, 1926, vol. 9, no. 2, pp. 105-106.

DISCUSSION

Presiding: J. A. BOWERS, melting superintendent, American Cast Iron Pipe Co., Birmingham, Ala.

MEMBER: I would like to ask Mr. Donoho about Brands C and D In Table 1. He found a difference in the apparent life of the brands, the silica ratio, and the porosity. The question arises as to whether the silica ratio or porosity is better.

MR. DONOHO: Mr. Greene, who is co-author of this paper and a ceramic engineer, supplied the brick and it was his opinion that porosity was the main property that affected the amount of slag attack. The high alumina brick had a much higher fusion point than either of the other two brick, but was dissolved much more by the slag. I believe the results of the test indicate density is the main property to consider when selecting brick to resist slag attack.

P. H. STUFF[1]: Evidently, in this particular case, the installation of the lining had a lot to do with the job. It is known that as good a lining cannot be secured in a small cupola as in a big one, also the cupola was running 16 instead of 8 hours.

MR. DONOHO: I think Mr. Stuff is right about variables in installation. There is no use comparing a few heats with single heats because there are so many variables, such as, the method of putting in the lining or how well the lining job was done. That is the reason, in these first figures, we found it necessary to average from 5 to 20 heats to get any direct correlation. As there are so many variables involved, the subject has to be studied statistically to correlate result.

MEMBER: Foundrymen have known for a long time that, with increased blast pressure, a refractory lining would cut out more rapidly and this is the first paper I have ever seen that had actual figures in it to show what the pressure and the degree of erosion was. Would Mr. Donoho tell us what the effect of this increased blast pressure is and whether it is a development of higher temperature in the melting zone or whether it is increased oxidation of the melting iron.

MR. DONOHO: Your guess is as good as mine. There probably would be some increased temperature with higher blast pressure up to a certain limit, but we thought it was more probably due to increased oxidation with higher blast pressure. That was the best reason we could give for this difference.

MEMBER: I notice you use a 20 per cent steel; suppose you used a 60 or 80 per cent steel, what would be the result?

MR. DONOHO: We found the percentage of steel did not make much difference if the percentage of silicon was the same. It probably would make some difference but we could not find a better correlation with the silicon content. If we used a high percentage of steel in the charge and a low silicon content, we had a severe attack on the lining. Where we

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[1] Metallurgist, Ross-Meehan Foundries, Chattanooga, Tenn.

had a high silicon content with the steel, it seemed to protect the lining somewhat.

MEMBER: What is generally considered to be the temperature in your melting zone with about 20 per cent steel in the charge?

MR. DONOHO: Several efforts have been made to measure the melting zone temperature of the cupola. It has been variously estimated as slightly over 3,000 to 3,200°F. The latter temperature probably is correct. I have heard of temperatures as high as 3,200°F. Perhaps someone else has a better figure.

MEMBER: On the general subject of the corrosion and erosion of linings with blast pressure, it would seem that there would be no question about the corrosion of the lining with the velocity. The velocity has a lot to do with all corrosion and erosion processes. When the blast pressure is increased, corrosion conditions naturally are accentuated. All fine material, such as liquid slag and drops of iron, is picked up by a high blast and thrown against the wall. A drop of iron that will fall through a soft blast will be picked up by a hard blast and thrown back up against the wall. Therefore, I think it would be very peculiar indeed if the action on the lining did not increase with the higher blast pressure.

MR. STUFF: Not long ago when we had a lot of difficulty getting the right size coke, we went to a smaller size, and did not change the blast pressure or anything else but the coke size. However, when the coke size was smaller, our erosion increased. When we went from a 42-in. stack to a 54-in. stack, our erosion increased so badly that the lining would not stay in for more than 2 days. Now it will stay in for more than 30 days with the correct size of the same coke.

MEMBER: Composition of the slag necessarily should be considered in this connection if a true picture of the cupola refractories is to be obtained. Compositions essentially will form a basic slag, yet silica brick is acid in nature. Have you done any work along this line that would indicate that corrosion of the brick is more rapid due to a basic than to an acid slag?

MR. DONOHO: By changing slags, we know we do change the degree of attack on the lining. The following data were presented in a paper* before the Fortieth Annual Meeting of the American Ceramic Society. There are a few slides showing the effect of varying slags on linings. None of these data were used in the paper just presented. In the paper referred to was another effect brought about by a different flux which was used, while the flux used in the paper just presented was dolomite, which contains a large percentage of magnesium carbonate. Our supply of dolomite gave out and we had to use a straight limestone flux, which is practically all calcium carbonate. We had some rather interesting results in its attack on the lining and had some very bad trouble with it. Table 6 shows data on the 72 by 102-in. cupola. The total number of heats in each case was 20, and they used from 150 to 500, nine-in.

* "Effect of Fluxes on Cupola Linings," Donoho, C. K. and MacKenzie, J. T. Bulletin, American Ceramic Society, vol. 17, no. 8, August, 1938, pp. 317-21.

equivalents. Where limestone was used, most of the heats required a higher number of 9-in. equivalents than when dolomite was used. However, there were some variations in blast pressures but the averages were close enough together so that the comparison was due to the flux used more than anything else.

On the bottom line, the average number of 9-in. equivalents is given. The dolomite flux seemed to give a more fluid slag, and the increased attack with limestone flux was probably because the slag was more viscous, there was more bridging in the cupola, and consequently, more oxidation. It all seems to point back to the iron oxides of the slag. We analyzed the slags from the two fluxes and found about 10 per cent iron oxide in the limestone slag and about 5 per cent in the dolomite slag, giving a 2 to 1 difference.

Table 6

EFFECT OF FLUX USED ON REFRACTORIES CONSUMED
72-BY 102-INCH CUPOLAS

		Flux	
		Limestone	Dolomite
Total number of heats		20	20
No. of heats requiring	150—200 9-in. equivalents		2
	200—250 " "		10
	250—300 " "	5	6
	300—350 " "	7	2
	350—400 " "	4	
	400—450 " "	1	
	450—500 " "	2	
	Over 500	1	
Av. number 9-in. equivalents required per heat		345	220

Table 7

EFFECT OF FLUX USED ON REFRACTORIES CONSUMED
60-INCH CUPOLAS

		Flux		
		Limestone	Limestone and 15% CaF_2	Dolomite
Total number of heats		14	12	8
No. of heats requiring	400—500 9-in. equivalents	9	8	8
	500—600 " "	2	4	
	600—700 " "	1		
	Over 700	2		
Av. number of 9-in. equivalents required per heat		522	483	442

Table 8

EFFECT OF FLUX USED ON REFRACTORIES CONSUMED

42-INCH CUPOLAS

Heat No.	Flux used		Refractories consumed (9-in. equivalents)	
1	Dolomite		188	
2	"		168 } av.	180
3	"		185	
4	Dolomite + 5% Purite		120	
5	"	"	125 } av.	121
6	"	"	118	
7	Dolomite + 10% Purite		178	178

Table 7 shows similar data on the 60-in. cupolas. We used lime-stone for 14 heats and had severe attack on the refractories. In some cases, we had hot spots in the cupola before we could finish the heats, which were of 16-hr. duration. In attempting to correct the attacks on the linings, we added some calcium fluoride, 15 per cent, to make a more fluid slag. Although fluorspar itself has a rapid attack on fireclay refractories, it did decrease the attack on the lining. When we finally got back to dolomite, we had 8 heats which were comparable with the others and all 8 of those required the smallest number of brick.

Table 8 shows data on 42-in. cupolas which were operated at rather high blast pressures and long heats and in which we had some trouble with the lining. Remembering our experience on the fluidity of the slag, we tried adding a little sodium carbonate or purites, which helped. There were 7 consecutive heats. In the first three heats, dolomite was used as a flux, with a very consistent average. When we added 7 per cent purite, the slag was more fluid and the number of brick consumed in the next three heats a great deal lower. We tried to get it lower by adding twice the amount of purite, but we had a severe attack on the lining and decided not to try that again.

This more or less agrees with the findings of other investigators, that a small percentage of purite does decrease the attack on the cupola lining.

MEMBER: We use purite in our slagging process and find it some-times very difficult to keep the iron clear when we are pouring our castings. Would the difference between that which is saved on the lining and the loss on the castings, offset the loss on the castings?

MR. DONOHO: I cannot answer that question because our castings are centrifugally cast and we actually add sodium carbonate at the pour-ing ladle. Therefore, a little more sodium carbonate in the slag certainly cannot hurt.

MEMBER: I would like to suggest as an answer to this question if the amount of purite is cut in half, it might not give that continued

action. A company in Chicago had that trouble and cut the purite in half and then did not have the attacks on the lining.

A. H. DIERKER[2]: There is often a misconception in some people's minds concerning the effect of slag on the cupola lining. It does little good to analyze, for this purpose, the slag that comes from the cupola slag hole or skimming spout for that slag is more or less in equilibrium with the lining and will not corrode it to any extent. As has been well shown in this paper, it is the more active fluxes like concentrated iron oxide, free lime, fluorspar, etc., which, before they have time to combine with the materials they should join up with, as silica sand, coke, ash etc., are thrown against the refractory walls in the melting zone and naturally attack them rapidly. I do not know exactly how it might be done but it would seem worth while to consider forming the slag earlier in the cupola. In this way, the active fluxes would be combined with the slag before they had time to attack the lining.

DR. MACKENZIE[3]: Both soda ash and the fluorite do that very thing. In our big cupolas, which hold about 1½-hr. melt, it takes just about 1½-hr. to get the effect of extra limestone. But if we put in, say, 40 to 50-lb. purite, the effect is obtained in 30 min. or less. This shows that purite is melting up in the stack and running down ahead of the charge. Fluorite does the same thing, but not so quickly.

Cupola Operation[1]

By Marcel Dudouet[2], Calvados, France

Translation by W. H. Spencer[3], Muskegon, Mich.

Abstract

This paper discusses the operation and functioning of a cupola from both the theoretical and practice standpoints. The author divides his presentation in four main parts. Part 1, theoretical functioning, Part 2, practical considerations, Part 3, construction data and Part 4, cupola practice. Under part 1, he discusses the mechanism of melting, influence of the cupola on the molten products, and reactions in the basin. Under practical considerations in the second part, he discusses height and position of tuyeres and cross sectional area and gives several operating rules. In the third part on construction data, he presents some formulae for stack heights, basin capacity, and tuyere wind box details. In the fourth part under cupola practice, the make up of charges is discussed and a list of operating rules plus items in the control of operation is given. He concludes this section by citing causes of troubles and presents remedies.

Introduction

In spite of the invention and improvement of newer equipment, the cupola remains by far the most used melting unit in gray cast iron foundries. It is probable that this extensive use is due to the extreme simplicity of construction, installation and operation of the cupola and also to its use of coke, an economical and easily handled source of heat.

It must be recognized that cupola operation has been improved constantly to meet the increasing demands of a growing industry. Frequently abandoned, somewhat hastily, for certain more delicate types of melting units, it has seen itself reinstalled later by those

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who condemned it. Capable of fulfilling its function in the great majority of cases, it has a very distinct advantage over most of the more perfected equipment opposed to it.

Although foundries in general retain the cupola as a melting unit for economic reasons, yet many hesitate to attribute to it the possibility of use for other than ordinary castings. This well-known reticence has its origin principally in the irregularity of results obtained in different cupolas, or even in the same cupola, from time to time. The variations obtained are found in the following characteristics: Chemical composition, structure and casting qualities.

Such disappointments, more or less proved, spring principally from two technological causes; (1) construction rules for the cupola badly adapted, (2) badly defined operation rules for the equipment.

The determination of these rules necessitates first an examination, without preconceived ideas, of the phenomena of cupola operation. The author, in his observations which follow, does not pretend to present completely the mechanism of such phenomena, since they are extremely complex and it is a gross error to try to extend the evident simplicity of construction to the operation of the equipment. These are two points which, from this point of view, have no connection.

Part I—Theoretical Functioning of the Cupola

The phenomenon of combustion in the cupola controls the temperature problem and also produces a condition reacting more Therefore, it is necessary to examine this action in detail.
or less strongly on the intrinsic worth of the molten products.

A. MECHANISM OF MELTING—PHYSICAL ASPECT OF THE PROBLEM

(1) *Basic Principle*

The degree of temperature necessary to cause the melting of the metallic charge is obtained by combustion of a fuel, generally coke, by means of a supporter of combustion, the oxygen contained in air, introduced at the bottom of the equipment through tuyeres.

Supposing the operation under way, the reactions of combustion are (neglecting the effect of nitrogen):

(Reaction 1). $C + O_2 = CO_2 + 97$ Calories—Complete combustion

(Reaction 2). $C + O = CO + 29$ Calories—Incomplete combustion
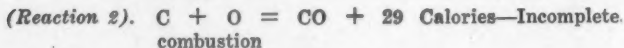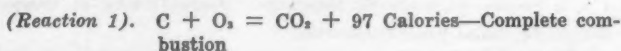
Whatever may be the nature of the reaction considered, the heat given off by combustion raises the temperature of the surrounding gas and the metallic charge by convection and radiation. The hot gases rising in the stack give part of their sensible heat up in preheating the charges (coke, flux, metal) contained in the cupola.

Hence, there is direct contact between:

(1)   The metallic charge (prior to its fusion).

(2)   The combustible material.

(3)   The products of combustion.

(2)   *Combustion of Coke in the Cupola*

To the two preceding reactions 1 and 2 there should be added that which controls the system of concentrations CO/CO_2 according to the temperature. In effect, CO_2 formed at high temperature reacts directly with the coke according to the following formula:

(*Reaction 3*). $CO_2 + C = 2\ CO - 39$ Calories
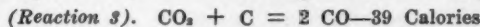
The intensity of the attack of CO_2 on C is greater to the same extent as the reacting materials are at a higher temperature. Boudouart's curve (Fig. 1) summarizes this property which leads to instability—the instability of CO at low temperatures and of CO_2 at high temperatures. Hence, it is necessary, in the presence of the incandescent coke filling the base of the cupola, to consider a noticeable attack on the C by CO_2 with the formation of CO in front of the tuyere entrance. This fact has been verified experimentally and the *average* results of numerous tests are represented by the graph in Fig. 2. Proceeding away from the tuyeres, three successive zones are found, each characterized by a higher concentration; first in O_2, second in CO_2 and third in CO.

Taking account of the experimental results and the general direction of the gaseous streams from the tuyeres, it must be conceded that the distribution of the base above the plane of the blast is represented graphically by Fig. 3. It should be noted likewise, that the position and the relative importance of the 3 zones in which O_2, CO_2 and CO predominate is influenced by:

(1)   The importance of reaction 3 as a function of reaction 1.

(2)   The velocity of the gaseous currents at the origin (the tuyere entrance).

(3)   The rate of blast (rate of flow by weight).

(4)   The nature and size of the coke.
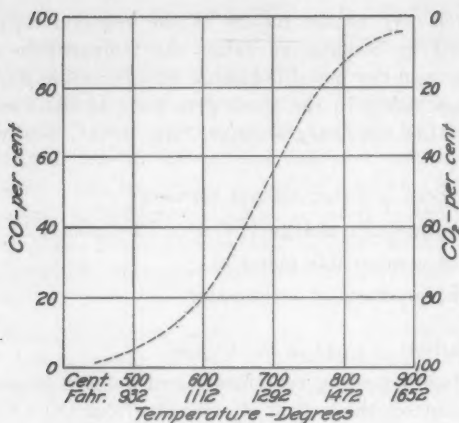
FIG. 1—BOUDOUART CURVE—PER CENT OF CO.
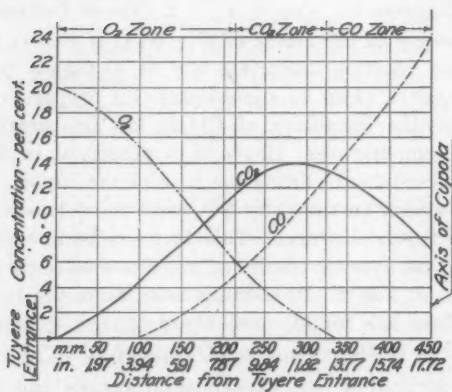
FIG. 2—AVERAGE COMPOSITION OF O_2, CO AND CO_2 IN THE PLANE OF THE TUYERES.
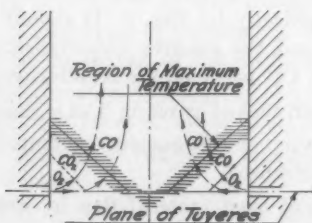
FIG. 3—POSITION OF THE ZONE OF HIGH TEMPERATURE (ACTIVE MELTING).

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(3) *Relation Between Combustion-Coke and Fusion-Metal*

According to the thermal balance of the reactions 1, 2 and 3, a liberation of heat results that is more important in the region where CO_2 predominates, which corresponds to the zone of higher temperatures. The temperature is moreover higher as the formation of CO_2 is greater and the attack on C by CO_2 is as weak as possible. It is tending toward a double objective to use a non-reactive coke and to charge it in layers having a well determined maximum thickness. It seems that the size of the lumps of coke has a real influence on the ratio CO_2/CO. In effect, all other factors being constant, the surface presented by the small coke is relatively greater and, in the presence of a current of O_2 sharply directed, increases rapidly the importance of the attack on C by CO_2 and diminishes the ratio CO_2/CO. The results arising from this statement are in accord with the conclusions of a paper* by J. T. MacKenzie and J. A. Bowers.

The temperature increases according to and in proportion to the descent, the material heats and, *under normal conditions,* melting is achieved, at the latest, at the level of the high temperature zone. It is evident that fusion is affected over a certain height but becomes constantly more and more active as the elements of the charge approach the high temperature zone. This region of active melting is commonly called the "melting zone" and is situated in the cupola in the neighborhood of the zone rich in CO_2 (Fig. 3).

(4) *Relation Between Combustion Rate and Melting Rate*

In order to represent clearly the melting process in the cupola, it is necessary to bring in the idea of "melting rate," defined as the time necessary to entirely melt a given proportion of the charge.

The material in the cupola is always present in the form of pieces of a certain thickness; each piece on descending heats up according to its approach to the tuyere zone. The temperature is raised progressively but always more rapidly on the surface than in the center, so it follows that melting commences on the surface and gradually reaches the center of the piece. This progress of the melting is regulated in velocity by the following factors: (1) the initial temperature of the metal; (2) the surrounding temperature; (3) the conductivity of the metal; (4) the melting point of the metal.

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* *"The Influence of Coke Size in the Cupola,"* TRANS. A.F.A., vol. 45 (1937) pp. 293-319.

If, under favorable conditions, these four factors are considered as constant, it follows that the time required for the melting of a piece of metal is proportional to the thickness or section of the piece. During this time, T, the piece considered descends in the stack a distance regulated by the combustion rate of the coke (a function of the kind and amount of coke and the blast supply).

If the amount of this descent is greater than the extent of the zone of temperature high enough to produce melting, it is evident that the piece of metal in question will pass through the normal melting zone and enter a lower temperature region rich in oxygen. The piece will undergo a profound and dangerous alteration before passing into the liquid state. There is an important relation between the course of combustion and the physical aspect of the charge which must be recognized in order to establish rules for efficient cupola operation.

B. INFLUENCE OF THE CUPOLA ON THE MOLTEN PRODUCTS

The existence of several widely different atmospheres around the melting zone implies a more or less profound modification in the properties of the molten products. In order to obtain a desired result with certainty, it is necessary to know the importance and mechanism of this change and to adopt corrective measures in advance.

I. ACTION ON THE CHEMICAL COMPOSITION

(A) *Before Melting*

(1) *Influence of Ascending Gas:* On leaving the melting zone, the gas is composed essentially of CO_2, CO, N, H (moisture in air), SO_2 (sulphur in coke), and sometimes a small amount of O_2 (excess blast). After the reactions in the stack, the gas composition may vary greatly at the charging door.

a. *Action on Low Carbon Iron:* Examination of the phenomena indicate that there is simultaneous oxidation by CO_2 and O_2 and reduction by CO. Under normal conditions, it is the action of CO which dominates and brings about, by raising the carbon content of the surface, a lowering of the melting temperature of the superficial metal which acquires a carbon content around 1.6 to 2 per cent. The interdependence of the phenomena of carbonization and melting implies the idea of the relation "rate of carbonization—rate of combustion." The thickness of the pieces of metal

likewise plays an important part when the unmelted metal reaches
the zone rich in O_2 where the heat present will not melt them, for
their enrichment in carbon has become impossible. They undergo
a very important change and pass into the liquid state only by
solution in the liquid bath, lowering the carbon content (without
heat addition) and increasing the amount of impurities.

b. *Action on Higher Carbon Irons:* (Usual case for cast
iron.) There are three classes, according to carbon content of each
element of the charge:

(1) Carbon less than 3 per cent. Slight tendency to carboni-
zation which decreases as the carbon content approaches
3 per cent.

(2) Carbon between 3 and 3.6 per cent. This interval corre-
sponds to what may be termed the equilibrium zone,
represented by the irons having the lowest melting
points (supposing the Si. content about 1 to 2 per cent).
For these metals there is little change in the carbon
content.

(3) Carbon greater than 3.6 per cent. These irons have
rapidly rising melting temperatures, melting is more
difficult to obtain and tends to occur near the zone rich
in oxygen. The oxidizing action tends to diminish the
carbon and lower the melting temperature as the com-
position approaches that of the equilibrium zone. The
loss of carbon seems more rapid in the case of graphitic
irons, probably because of the presence of large, easily
oxidized flakes. The carbon in Fe_3C is more difficult to
attack.

c. *Action on the Silicon:* The carbonizing action of CO is
practically non-existent, because the alloys of silicon used in the
cupola are saturated with carbon. On the other hand, due to the
affinity of silicon for oxygen, the action of CO_2 and O_2 takes place
very rapidly. As the silicon is present in the charges, either in
low percentage in the pig iron or in more concentrated (25 per
cent) alloys having a melting point approximately the same as
that of the pig iron, melting takes place under normal melting
conditions before the danger zone is reached and the changes under-
gone by silicon are reduced in importance.

d. *Action on the Manganese:* As in the case of silicon, only

the action of O_2 and CO_2 is to be feared. By avoiding the alloys rich in manganese (less than 40 per cent) having high melting points (confer Tamman diagram), the danger of having the melting take place in the zone where O_2 predominates with the consequent loss of manganese avoided.

e. *Action on the Sulphur and Phosphorus:* These two metalloids do not seem to suffer from the influence of the gas in the cupola 'and in practice do not undergo any alteration.

f. *Particular Action of SO_2:* The SO_2 (coming from the combustion of the coke) mixed in the gas tends to diffuse into the iron at high temperature (phenomenon of cementation) and raises the final sulphur content of the molten metal. This reaction is the principal cause of sulphur pick-up in cupola melting.

(2) *Mutual Reaction of Products Charged:* At the points of contact between iron and coke in the charge, there may be a chemical change (carbonization), favored by high temperature and low carbon content in the metal (low carbon iron). However, the contact surfaces being small and not changed much, the direct exchange of carbon may be considered as negligible.

(B) *Melting*

Normal melting occurs above the zone rich in O_2 and CO_2. But the drops of molten metal traverse this zone and run over the pieces of glowing coke to reach the basin or fore-hearth. During this trip, there occurs simultaneously:

(1) Oxidizing action by the gas.

(2) Carbonizing reduction by direct contact of the metal and coke.

In practice, from these two opposed reactions there results an oxidizing action principally on the iron, silicon and manganese. According to the law of mass action, the iron receives the largest amount of O_2 in absolute value. In relative value, it is the manganese and silicon especially which are most changed, with all the more intensity as the charges are composed of richer alloys. In the particular case, where the drops of metal are poor in carbon (2 per cent), there occurs during the trickling down an increase in carbon, bringing the final content between 2.3 and 3 per cent. The amount of carbonization, all other factors remaining constant, is proportional to the length of the path of the molten drops and is greater as the initial carbon content is lower.

C. *In the Basin*

The drops of molten material, more or less changed, collect in the basin where the slag separates from the liquid metal by gravity. The liquid surrounds the pieces of incandescent coke which sustains the charges in the stack.

(1) *Reaction Between Molten Metallic Products:* The liquid state permits the homogenization of the chemical composition. Moreover, the contact of iron oxide with the reducing substances such as silicon and manganese permits a displacement of the oxygen according to the reaction:

$$2 \ FeO + Si = Fe + SiO_2$$
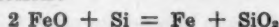
Normally, because of its density, the silica formed will float into the slag but more slowly as the bath is less fluid. In certain cases, this limits the reaction and leads to solidification with the imprisonment of impurities. However this may be, there results a loss* of silicon and manganese which is greater as the iron is more oxidized.

(2) *Reaction Between Metal and Slag:* The contact of the two materials may be made either at the surface of separation while at rest or as the drops of metal pass through the slag layer. In spite of this, there does not seem to be an active reaction because of the lack of temperature and the chemical composition of the slag. Certain foundrymen claim that the lime of the slag brings about desulphurization according to the following equation:

$$CaO + FeS + C = CaS + Fe + CO$$

In practice, it seems that this reaction is of appreciable importance only in the case of high sulphur contents (.0150 per cent) and acts only as a limit for enrichment.

(3) *Reaction Between Molten Products and Coke:* The recarbonization of the cast iron by contact with the coke in the basin is very limited for two reasons:

(a) The iron arrives, with a carbon content above 2.5 per cent, near the equilibrium point where the speed of reaction is greatly reduced.

(b) The contact between the coke and iron, while over an extended surface, is not renewed and the reaction is limited by the speed of diffusion of the carbon in the metal at repose, a speed relatively low.

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* The analysis does not always give the true silicon loss because, if the silica remains in the metal, it is determined the same as the active silicon. The use of chill tests gives a more judicious determination.

Consequently, the slight enrichment in carbon content, by contact in the basin, depends on the initial carbon content of the metal as it arrives in the basin and on the time it remains in the basin. Normally, for final carbon content in the neighborhood of 3 per cent, the recarbonization is of the order of 0.2 per cent.

Contrary to what has been frequently stated, the sojourn of the iron in contact with the coke has a very small influence of the final sulphur content. The sulphur, being uniformly distributed in the coke, for a carbon gain of 0.2 per cent in the basin with a fuel having 80 per cent carbon and 1 per cent sulphur, the final sulphur content of the iron should increase 0.0025 per cent. This is very low in relation to the usual content.

(4) *Reaction Between Cast Iron and Lining:* The chemical action is very slight due to the relatively low temperature. However, if, at the beginning of a heat, the lining of the basin is impregnated with products from a former heat, these may dissolve in the first metal and sensibly modify the desired composition.

II.   ACTION ON THE PHYSICO-CHEMICAL PROPERTIES

There remains the serious consideration that the change in the metal, under the form of loss in silicon accompanied by inclusion of entrapped silica, leads to a sensible lowering of graphitizing power on freezing. This applies especially to irons of low fluidity, such as high tensile irons low in carbon and phosphorus. This not only renders the iron less machineable, but also decreases its casting properties, such as fluidity and resistance to cracking and shrinking, without which there would be no possibility of producing high strength castings. Moreover, it should be noted that this action, having a persistent character, is checked with difficulty by non-preventive remedies.

C.   CONSIDERATIONS

(1) *Air Supply*

The rate of blowing controls the quantity of carbon burned in a unit of time and is one of the factors in the combustion rate. Any increase of air supply diminishes proportionally the loss by radiation (a function of time) and should improve the thermal efficiency. This consideration is limited by:

a.   The extension of the zones rich in O_2 and CO_2 increases the changes in the metal.

b. The melting speed*, which is for the most part independent of the air supply, ought to be in proportion to the combustion rate.

It should be noted that the two conditions tend to aggravate each other mutually.

When it is the custom of a foundry to blow at constant pressure, it may be stated that the air supply, under normal conditions and for a given coke, is a function of:

a. The reduced section of the stack (varying with the size and arrangement of the pieces and the height of the stack).

b. The density of the gases in the stack (varying with temperature and composition of the gases, CO_2/CO).

c. The pressure in front of the tuyeres (motive energy).

d. The loss of pressure due to the tuyeres.

Supposing the conditions a, c and d constant, it is seen that the air supply is a function of the course of the combustion (CO_2/CO). It does not seem to be of interest to maintain the air supply rigorously constant when conditions of the charge necessarily bring variations in the regular operation of combustion. The best value for the blowing rate probably should vary according to the operation of the cupola. It always happens that the variations in operation tend to increase the equilibrium variations. A patented French device seeks to correct this condition by assuring an instantaneous control of the air as a function of the operation of the cupola.

(2) *Static Pressure Before the Tuyeres*

Representing the motive energy of the air which should circulate in the cupola, the manometric pressure in the wind-box is a factor in the blowing rate. This is generally the reference point in foundries for controlling the blast supply, but this knowledge is not sufficient. The operation of the cupola, as well as the blast supply, may vary considerably without the pressure figure changing.

(3) *Wind Speed at Mouth of Tuyeres*

This factor, wrongfully neglected, controls the combustion activity in the region of the tuyeres. Moreover, a sufficient speed

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* TRANSLATORS NOTE: (By melting speed the author apparently means the time necessary for a given piece of the metal charge to entirely melt.

is necessary to permit the air reaching the center of the cupola and to combat the tendency of the gas to travel along the walls, cooling the lining immediately above the tuyeres and forming slag bridges. The increase of the velocity at the tuyeres without increase of volume should be greater as the cupola is larger in diameter.

Part 2—Practical Considerations

From an examination of the theoretical operation as given above, the following practical conclusions may be drawn:

A. CONSTRUCTION

(1) *Height of Basin (Vertical Distance Between Bottom and Tuyeres)*

When producing low carbon cast irons, it is necessary to reduce this distance (the path of the iron drops trickling over the coke), to a minimum and in extreme cases to utilize a fore-hearth.

(2) *Total Cross-Sectional Area of Tuyeres*

To obtain a high blast velocity at the tuyere opening, it is necessary to have a relatively small cross-sectional area for the tuyeres, comprising between 1/12 and 1/16 of the cross-sectional area of the cupola at the level of the tuyeres. Larger cupolas require more reduction of the tuyere opening. To avoid the formation of slag bridges, round tuyeres are preferred to flat tuyeres.

(3) *Position of Tuyeres*

Except for very large cupolas, there is no need for placing the tuyeres in two or three different planes (extension of the zone rich in O_2 and CO_2). In any case, the distance between two tuyeres should not exceed one fourth of the inside diameter of the cupola.

B. OPERATION

The Relation: Combustion rate — melting rate controls the stability of the melting zone in respect to the zone rich in O_2 and CO_2. It is indispensable, for a given coke, to adopt a blowing operation that takes into account the size of the pieces and the melting point of each material in the charge.

The charging order in the cupola ought to take account of this

double consideration. It is necessary to charge first the largest pieces of the material, which are most difficult to melt.

To diminish the importance of the attack of CO_2 on carbon in the high temperature region, the fuel should be charged in layers having a maximum thickness of 10-in. no matter how large the cupola may be. When the ratio of metal to coke must be changed, the metallic charge should be changed and the coke charge left alone.

Apart from the preceding considerations, it is not necessary to use steel scrap to obtain low carbon cast iron. It is sufficient to have recourse to metals containing around 2.5 per cent of carbon and to limit the carbonization after melting (using a fore-hearth). This practice permits greater regularity of operation.

The necessary silicon and manganese additions should always be made in the form of alloys of low silicon and manganese content having low melting points. For silicon, use an alloy of 25 per cent and for manganese a spiegel of 10/12 or 18/20 per cent at the most. This rule applies to all the readily oxidizable elements.

The change in chemical composition of the portions of the charge being practically inevitable, it is necessary to provide a remedy to seek a regular melting method. Only permanent control, revealing the incidents of the operation, permits attainment of this result. (See elements of control later.)

Part 3—Construction Data

Beyond the limits given in the preceding part A, other characteristics may vary within wide limits. For clearness, it seems preferable to give the construction data, which is independent of the type of work, in a table. (Fig. 4.)

Normally, the cupola is cylindrical throughout its height. The hourly production or melting rate is given by the formula:

$$\text{Production in tons per hour} = 5.5 \ D^2$$
$$\text{Where } D = \text{Diameter in yards}$$

Stack Height

Vertical distance (H) from tuyeres to charging door is given by the empirical formula:

$$H \text{ (in yards)} = 1.1 + 3 \ D \text{ (in yards)}$$

taking into account the necessity of assuring heat exchange between the gas and the solid products and avoidance of hang-ups.

Basin

The vertical distance between the bottom and the slag hole depends on the maximum amount of metal it is desired to accumulate. To mix the liquid metal, the minimum capacity should be 30 to 40 per cent of the hourly melting rate of the cupola. The minimum height of the basin for any size of cupola is 18 to 20 in.

Distance From Slag Hole to Tuyeres

For construction reasons and likewise to avoid frequent cleaning, this distance is taken as 8 to 12 in.

Number of Tuyeres

The total cross sectional area having been given above, the number is determined from the cross sectional area of a tuyere. From all view points, the best size has a diameter between 4 and 5.5-in.
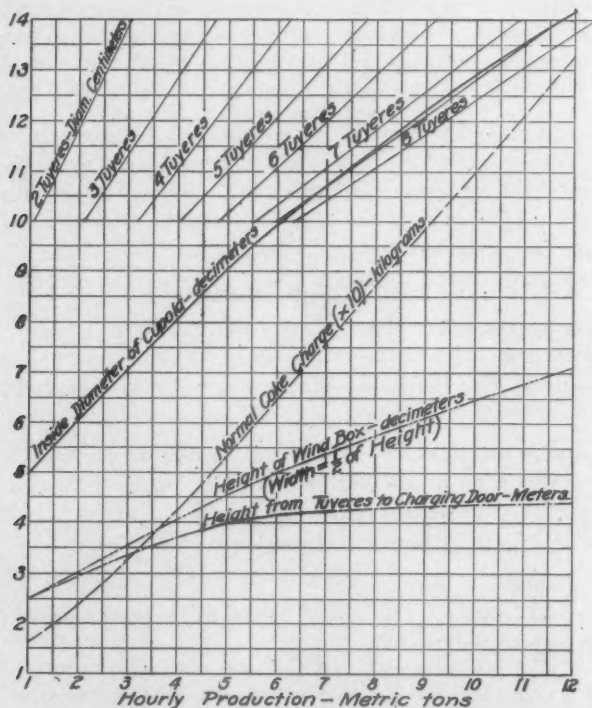
FIG. 4—VARIOUS FACTORS RELATING TO HOURLY PRODUCTION.
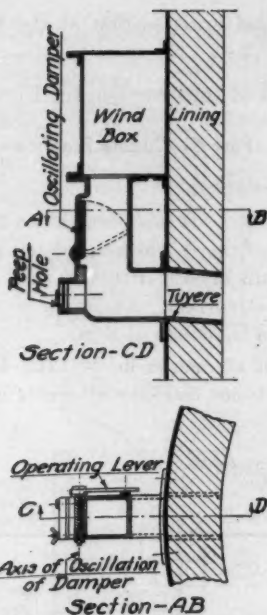
FIG. 5—AN ADJUSTABLE TUYERE ARRANGEMENT (SECTION C-D).

In case it is desired to regulate the tuyere opening, possibly closing each tuyere altogether, it is well to adopt one of the arrangements shown in Figs. 5 and 6. The latter is preferable.

Wind Box

To assure a good distribution of the air to the tuyeres, the box should be large. It is necessary to calculate a cross section
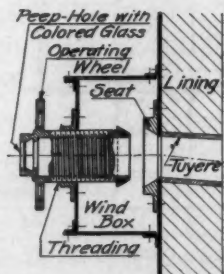
FIG. 6—A SECOND TYPE OF ADJUSTABLE TUYERE ARRANGEMENT.

about that of the total cross section of the tuyeres. (Height =
Twice the Width.)

For other details of construction, see Figs. 7 and 8.

Part 4—Cupola Practice

A. *Make-up of the Charge*

(1). *Coke Charge:* It has been shown that, to avoid a rapid
attack of the coke by CO_2, it was helpful to charge the coke in the
form of relatively thin layers. Practice indicates that a thickness
of from 8 to 10-in. is correct. Assuming a weight of 707 lb. of
coke per cu. yd., for a thickness of 8 in.

$$\text{Weight of coke in lb.} = 121.4 \ D^2$$

$D =$ inside diameter of cupola in yds.
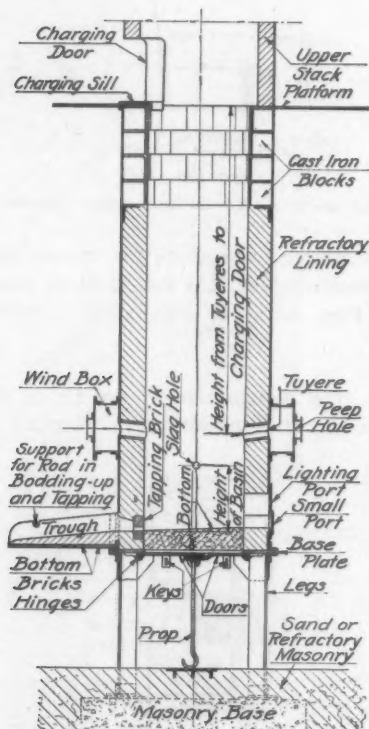
FIG. 7—A CUPOLA PLAN.
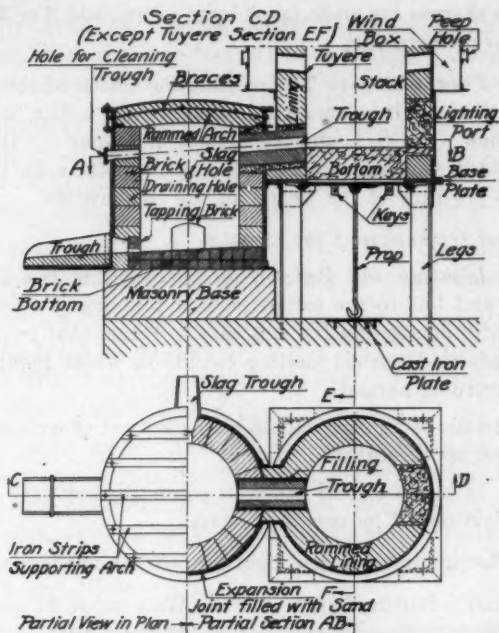
Fig. 8—Cupola with Fore-hearth Details.

(2). *Metallic Charge:* There remains to be determined the weight of metal to be charged and the melting ratio. As a general rule, the percentage of coke to iron varies from 7 to 15 per cent according to:

a.   The chemical composition of the charge.

b.   The physical make-up of the charge (size of pieces).

For pieces of metal of ordinary size, between 0.2 and 2-in. in thickness, the coke required will be:

8 per cent for high phosphorus (Phosphorus greater than 1.3 per cent) irons of normal carbon content.

10 per cent for hematite iron with normal carbon (Carbon greater than 3.2 per cent).

12 per cent for low carbon (Carbon around 2.5 per cent) irons.

If the charges are made up of large pieces, add 2 or 3 per cent to the above percentages.

(3). *Flux Charge:* To facilitate the fusion of the coke ash, generally silicious, it is common practice to add a flux in the form of limestone ($CaCO_3$) and sometimes fluorspar (CaF_4). The amount added, varying with the ash content of the coke, is between 15 and 40 per cent of the weight of the combustible.

B. *Normal Operation of the Cupola*

(1). *Lighting the Bed:* This operation is too frequently neglected and left to the initiative of a foundry laborer. It is a very harmful negligence, for it is easy to forget that poor lighting always leads to disturbed melting conditions which require a long time to return to normal.

The details of lighting should be the subject of written instructions drawn up carefully.

Here, in chronological order, is given the sequence of operations in lighting off by means of wood:

a. The lining being dry, light a wood fire on the bottom sand.

b. Add a fourth charge of coke. Wait about 30 min. so that the coke may be well kindled by the natural draft of the stack. Fill up with coke to 40-in. above the level of the tuyeres, adding the normal charge of flux composed of limestone and fluorspar in equal parts.

c. When the coke is well kindled above the tuyeres, close the large light-off opening leaving the small hole in the large door open. (Figs. 7, 8.) Close the charging door and blow at a low pressure (1 oz. per sq. in.). This blowing should continue for 5 to 8 min., heating the basin, the bottom, and the tap hole, the latter being left open.

d. Close the small light-off hole immediately and prepare the tap hole. To avoid freezing of the iron in the hole at the beginning of the heat, it is necessary to proceed as follows:

Introduce a handful of dry sand mixed with an equal part of powdered wood charcoal into the tap hole to form a protective plug behind the tapping brick. This assures a definite stoppage while drying the molding sand in the niche before the tapping brick. In this way, the first tap is made without trouble even with an iron of low fluidity.

e. By means of a gauge, verify the height of the bed coke,
which ought to be from 31 to 39 in. above the level of the
tuyeres. Be sure that the coke has not arched or bridged.

f. Begin charging by taking the precaution to form slag by
means of equal parts of limestone and fluorspar, to assure
fluidity of the first slag in spite of a lower temperature
than later in the heat. Do this for three or four charges.

g. When there are four or five charges in the stack, turn on
the blast, close the tuyeres and continue charging.

Under such lighting conditions, the first drops of liquid iron
should appear at the tuyeres five or six minutes after the blast is
turned on with charges having normal sized pieces, and in 10 to 15
min. for larger pieces of iron.

(2). *Blast Rate:* The rate of flow value has only a limited
interest since most foundrymen measure only the pressure in the
wind-box and consequently can not use any other measurement.
The pressure, as previously stated, is partially a function of the
dimensions of the cupola. In practice, the pressure used varies
between 4 and 12 oz. per sq. in.

It should be noted that high pressures and consequent high
volumes only give an impression of a high working temperature,
because of the fact that a part of the silicon is burned by extend-
ing the zone rich in oxygen. This oxidation of silicon gives off a
large amount of heat. The slower working is recommended, giving
high temperature due to furnishing the optimum amount of air.

(3). *Control of Operation:* The control should permit the
quick observation of variations which may occur in the normal
operation. It is necessary to mention:

a. *Periodic Controls That Do Not Give Immediate Results:*

(a) control of the raw materials (kind, analysis, structure).

(b) control of the combustible (sulphur content, moisture,
density).

(c) chemical analysis of the product obtained.

b. *Methods of Control That Are Quick and Give Results
Expressible in Numbers:*

(a) chill test (giving the graphitizing ability)

(b) fluidity test (function of the working temperature and of changes in the molten metal), the Cury method is used in France.

(c) the temperature of the molten metal at the tap hole (by optical pyrometer).

c. *Methods of Control That Are Quick but Expressed by Estimates:*

(a) descent of the charges

(b) inspection of the tuyeres

(c) appearance of the slag (viscosity hot, color when cold)

(d) appearance of the molten metal.

(4). *End of Heat:* The level of the charge descends and the resistance in the stack is lowered. This condition immediately affects the volume of air if the pressure is the only blast control used. So it is proper to lower the pressure in proportion to the lowering of the level of the charges in the stack.

If a volume meter is used, it is sufficient to simply keep the volume constant.

C. *Operation Troubles*

It is frequently stated that a well operated cupola does not have operation troubles, but it is better to consider that a knowledge of the possible troubles, their causes, symptoms, and remedies give a guarantee for their avoidance.

(1). *Hang-up*

Symptoms — A prolonged arrest in the descent of the charges, molten iron does not appear at some or all of the tuyeres.

Causes —

a. Pieces of too great a length arch themselves against the wall;

b. general welding of a charge composed of very small pieces due to an accidental mounting of the flame in the stack;

c. general welding of the material at the level of the melting zone because of a prolonged shut down (especially dangerous at the beginning of the heat).

Remedies — Cut the blast and poke the charge at the door. If the charge falls more than 20 in. when the hang-up is loosened, it is necessary to add 2 or 3 extra coke charges with the flux added as fluorspar. A cold metal with diminished graphitizing power is to be expected.

Precautions — The only preventive measure consists in distributing the charge well at the door and avoiding the use of too large pieces (the rule is: maximum length less than 0.5 the inside diameter of the cupola.)

(2). *Lowering of Melting Zone* —

Symptoms — Unmelted pieces of iron appear before one or more of the tuyeres at regular intervals.

Causes — Lowering of the melting zone which originates:

a. in irregular charging, if the unmelted pieces always appear before the same tuyeres;

b. too low a coke ratio or too high a combustion rate if the pieces appear before all the tuyeres.

Remedies — Besides checking the charging methods and correcting any errors which are observed, it is necessary:

a. to increase the percentage of combustible (reducing the metallic charge);

b. to add extra coke charges at regular intervals.

The latter practice gives best results if it is impossible to determine the cause.

(3). *Penetration of Lining*

Symptom — Redness of the shell at a point level with the melting zone, sometimes proceeding to melting of the shell with the escape of gas and material.

Causes — Rapid wear of the lining or very long heats with charges difficult to melt.

Remedies — Direct a jet of compressed air or water on the hot spot. End the heat as soon as possible.

Precautions — This trouble is often an indication of a poor patching job done on a lining poorly cleaned of the slag from a previous heat. Otherwise it may indicate, if the patching was well

done, that the patching material used was not sufficiently refractory

(4). *Freezing of Tap Hole*

Symptom — Inability to tap out.

Cause — Frozen metal in the canal leading from the basin to the outer channel.

Remedy — Burning with oxyhydrogen torch.

Precautions — The tap hole should be from $1\frac{1}{4}$ to $2\frac{1}{2}$ in. long, according to the size of the cupola. If it is too long, the metal remaining in contact with the plug cools and tends to freeze. The plug, after stopping up, should completely fill the tap hole; its composition, equal parts of clayey sand and pulverized burnt coke mixed dry, ought to resist strong pressure and yet break easily when tapped.

Remarks — With the use of a fore-hearth, freezing up is more frequent for the metal is generally colder. On this subject, it should be noted that the heating of the fore-hearth is easily accomplished in the following manner: leave open the slag hole of the fore-hearth and bring about a circulation of hot gas coming from the melting zone while bringing into the fore-hearth a current of air from the wind-box, thus burning the gas and assuring a very efficient heating of the fore-hearth. In this case, it is necessary to provide a lining which is very refractory to resist the temperature of the flame.

(5). *Flame at Charging Door*

This is not properly speaking a trouble, but does indicate bad practice.

Causes —

a. Charge very thin, flame mounts in the charges.

b. Charge very permeable (case of using small thin pieces such as turnings, stampings, etc.); melting is more rapid than the combustion.

c. Charging door too low; the heat exchange is insufficient to cool the gases.

d. Existence of a hang-up of the charges.

Remedies — Except in the case of too low a charging door, it will help to modify the charge in favor of a higher coke bed.

Precautions — Check the dimensions of the cupola. In case turnings or other permeable constituents are used, it is better to compress them before charging to increase their apparent density. If this is impossible, it is necessary to operate at a slow rate with a reduced coke ratio and large charges.

(6). *Accidental Stoppages of Melting*

During normal melting, it is possible to stop blowing the cupola for a certain time without much trouble. In that case, as soon as the blast is cut-off, open the peep-holes of the tuyeres, so that the natural draft (which may be increased by closing the charging door), permits the maintenance inside the cupola of a temperature sufficient to avoid the accumulation of CO in the windbox and the danger of an explosion when the blast is put on again.

CONCLUSION

Given working temperatures too low, the cupola can not be a metallurgical furnace employing slags of a corrective or purifying nature. Its role remains, as it was originally, that of putting metal into a liquid form, permitting the economical pouring of any castings. Certainly, advantage has always been taken of this melting to effect changes in the form of additions of corrective elements. These are added to the charge and melt at the same time, diffusing in the molten metal. Likewise, there has been a tendency for some years to introduce certain corrective elements directly into the liquid metal at the spout. This practice has been adopted because of influences, poorly explained but favorable, for obtaining products having excellent qualities both in respect to physical properties and casting properties. This method of addition (studies by Portevin and Lemoine) as one of the principal manufacturing methods frequently includes the use of an alloy very rich in silicon alloyed with calcium, titanium, aluminum, etc. Such a method, which may be greatly improved, has contributed effectively to the extension of the field of action of the cupola and it is hoped that it will be used by the majority of foundrymen.

To end this modest contribution to the study of the cupola, it seems necessary to repeat that the production of quality cast iron takes into account:

(1). The chemical analysis desired in the finished product, as everyone knows.

(2). The physico-chemical nature of the raw materials used, which many wish to ignore.

SUMMARY

A badly operated cupola produces poor cast iron even with good raw material.

A well operated cupola, using good raw material, gives excellent results.

A well operated cupola, using raw material of poor quality, often produces cast iron unfit for use.

This summary is necessary for in the presence of a question as complex as that of the production of cupola cast iron, it is requisite to seek the solution of each of the factors composing it. Even perfectly solving only one of the factors does not give a complete solution of the given problem.

DISCUSSION

W. H. SPENCER[1] *(Written Discussion):* Small lumps of coke frequently have a different structure from large lumps and are small because of the structural weakness. The CO_2/CO ratio from such small lumps will be due not only to the surface presented but also to the nature of this surface. M. Dudouet's remark, "all other factors being constant", is of great importance. Commercially, smaller coke usually does not have all other factors constant.

The relation between the course of combustion and the physical aspect of the charge is, as the author states, of great importance. For a given melting condition there is an optimum $\dfrac{\text{surface}}{\text{volume}}$ ratio for the pieces of metal in the charge.

The gas composition at the charging door is not the same as it was on leaving the melting zone. Gas analyses of samples taken at the charging door would not represent directly conditions in the melting zone.

In discussing the action on high carbon irons before melting, the lower loss of carbon in irons having their carbon in the form of Fe_3C is mentioned. The statement is made that the carbon in Fe_3C is more difficult to attack. Is it not possible also that the irons having their

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[1] Metallurgist, Sealed Power Corp., Muskegon, Mich.

carbon in the form of Fe₃C instead of graphite have lower melting points? In this case, melting would take place above the oxidizing zone giving the result noted, less carbon loss when carbon was in the combined form.

The diffusion of SO₂ into the unmelted iron in the stack very likely explains why cupolas having different lengths from tuyere level to charging door give different sulphur pick-up under similar operating conditions. The amount of sulphur gained by the metal in this way can be appreciated if the metal charge is whitewashed. The lime in the whitewash should be balanced by a removal of limestone from the flux charge in order to maintain the constant slag composition.

The enrichment in carbon content of the molten metal by contact with the coke in the basin is stated to depend on the carbon content of the metal as it arrives in the basin and the time it remains in the basin. It would seem that the reactivity of the coke and the temperature of the metal would also affect the carbon enrichment of the metal.

While the reaction between the iron and the cupola lining is negligible, the reaction between the flux and lining is very appreciable. The difference in composition of slags due to different lining materials will have more or less effect on the molten metal.

In America a large portion of the foundries are equipped with means of measuring either volume or weight of air supplied to the cupola. The chill test as a control method is also quite widely used, and the importance of the fluidity test is becoming more widely recognized.

G. P. PHILLIPS[2] (*Written Discussion*): Under the heading "Action On Higher Carbon Irons" the statement is made that there is little change in carbon content of iron charges with carbon contents of from 3.0 to 3.6 per cent. We find appreciable increase in carbon during melting with charges containing initially about 3.00 per cent carbon. This increase ranges from 0.30 to 0.45 per cent resulting in final carbons of about 3.30 to 3.45 per cent. Our experience also indicates that it would be necessary to go well above 3.60 per cent carbon to have a reduction in carbon content during melting.

Under the heading "Reaction Between Metal And Slag" the statement is made that the desulphurizing action of slag is of importance only in the case of high sulphur content (0.15 per cent) and acts only as a limit of enrichment. We find that slightly less than half the sulphur in the coke is removed by the slag. Our metal charges enter the cupola with a sulphur content of from 0.055 to 0.065 per cent. Melting at an iron-to-coke ratio of 8 to 1 we have sufficient sulphur in the coke to raise the sulphur content of the metal by 0.085 per cent. Actually our sulphur increase is from 0.035 to 0.050 and the slag will contain up to about 3.00 per cent sulphur.

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[2] Foundry Met., International Harvester Co., Chicago, Ill.

J. A. BOWERS[s] (*Written Discussion*): We would like to call attention to the recommended ratio of tuyere area to cupola area. The author states that the cross-sectional area of the tuyeres should be between 1/12 and 1/16 of the cross-sectional area of the cupola. This is quite different from the usual practice in any of the foundries with which I am familiar, and particularly in the plant of the American Cast Iron Pipe Co. In this plant the ratio is in the following order: Approximately 1/9 on a link-shaped cupola 72 x 102 in.; 1/5 on a 42 in. diameter cupola. This ratio usually varies from 1/5 to 1/10, depending on the diameter of the cupola. Mr. Dudouet further recommends the use of round tuyeres which is also contrary to our practice, except on the extremely small diameter experimental cupola (21 in.). All of our production cupolas have rectangular tuyeres and are flared toward the inner diameter to avoid any nozzle effect to the air entering the cupola. The tuyeres on the link-shaped cupola measure 4-1/2 in. high and 16 in. wide at the air inlet and are flared to 24 in. wide at the inner diameter. The long sides of the rectangle area are, of course, in the horizontal position.

In section 4, under "Blast Rate", the author states that high blast pressure only gives the impression of high working temperature. We have not found that true except in those few cases where there was not sufficient coke which was very soon apparent by the dark brown smoke issuing from the stack. The author's statement is not in accord with the conclusions of a paper[**] by J. T. MacKenzie and C. K. Donoho, nor several notable instances in American practice.

It is interesting to note that the author says that although considerable flame at the charging door is not necessarily a trouble, it is an indication of bad practice. He gives several good reasons, but leaves out the very one that causes considerable flame at our charging door, namely—blast pressure. The cupolas in our plant are blown from 30 to 35 oz. pressure, and at the higher pressure we always have more or less flame at the charging door. This is probably due to the high blast carrying sparks up the stack in sufficient quantity to keep the flame burning. However, this particular phenomenon has not been an indication of trouble whatsoever. Furthermore, cupola operators here prefer a slight excess of carbon monoxide at the charging door level, whether it is burned or not, as they consider this is a very good indication that there is sufficient coke in the cupola for the amount of air, and, therefore, no danger of excessive oxidation. It is particularly important in our operation to carry a slight excess coke in the cupola because the tonnage may vary from 20 tons per hour to 35 tons per hour.

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[s] American Cast Iron Pipe Co., Birmingham, Ala.
[**] Effect of Blast Pressure in Cupola Melting, presented at 1938 Annual Convention of American Foundrymen's Association.

Austenite Transformation in Gray Iron*

BY D. W. MURPHY[1], W. P. WOOD[2], AND C. D. D'AMICO[3],
ANN ARBOR, MICH.

1. Variations in the properties and behavior of cast irons of like composition and method of manufacture have stimulated much investigation leading to a better understanding of the material and of methods for improving its properties. The amount, type, and distribution of graphite in the matrix of cast irons have long been recognized as contributing greatly to the properties of castings and much of the recent progress in gray irons is a direct result of intensive studies of the methods of control and the effects of these factors. Composition, methods of processing, and section size, for example, are all known to have an important influence on the graphite phase in gray irons. The major concern of the present investigation has been with the matrix of the iron rather than with the graphite characteristics, although it is entirely possible that subsequent research will demonstrate a considerable degree of inter-relationship between graphite and matrix characteristics.

2. Certain of the effects of changes in the matrix structure are well established. An iron having a structure made up of ferrite and pearlite is usually found to be rather soft and weak, while an iron containing pearlite and free carbide would be much harder and stronger, except perhaps in shock resistance, by comparison. Composition, methods of manufacture and section size are again of much importance in controlling these aspects of structure. The well-known Maurer diagrams relating silicon and carbon content to structure for various section sizes summarize much of our knowledge of the effects of composition and section size on structure.

3. In addition to these variations in structure and their effects on properties it has been observed that in the fully pearlitic irons a fairly wide range of properties is frequently encountered regardless of a close similarity in composition and other factors which are commonly associated with the determination of structure and properties in gray irons. The first[4] of a series of papers dealing with such variations described the properties of piston ring irons of nearly the same analysis as obtained during regular daily production, all irons being melted in a cupola under conditions as nearly identical as possible. The irons discussed in this paper, all fully pearlitic, showed a variation in tensile strength from 37,500 to 53,000 lb. per sq. in. and in hardness from Rockwell C24 to C30. The composition of this series was within the following range: 3.73 to 3.75 per cent total carbon, 0.54 to 0.57 per cent manganese, 2.86 to 2.93 per cent silicon, 0.60 to 0.68 per cent combined carbon, 0.40 to 0.43 per cent phosphorus and approximately 0.065 per cent sulphur. It was observed that the hardness, strength, and tendency to distortion increased as the size of a network pattern increased. At that time it was suggested that this network pattern outlined an austenitic grain size which was perhaps determined during or very shortly after solidification but which affected the final

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[4] Murphy, D. W., and Wood, W. P., Trans. Am. Soc. Metals. Vol. 24, p. 75 (1936).
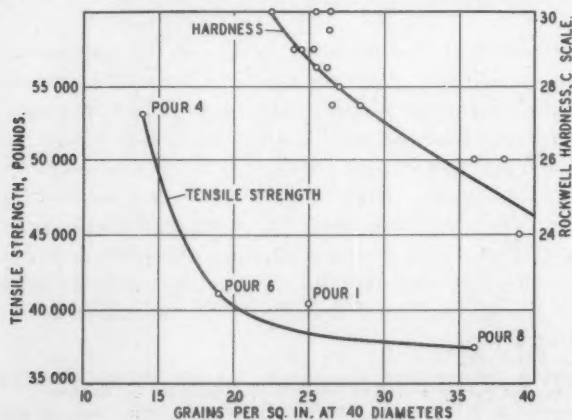
FIG. 1—VARIATIONS IN STRENGTH AND HARDNESS OF PISTON RING IRONS.

properties of the castings. The variations in properties of these cupola irons plotted against the observed grain size at 40 diameters is shown in Fig. 1 in which it will be observed that strength and hardness increased very materially with increasing grain size. A further indication from this work was that the ability of an iron to harden by quenching from above the critical point increased as the size of the network pattern increased. Comparing these variations with those observed in steel as resulting from grain size variations offered some justification for considering the cell pattern to be a trace of an austenitic grain size.

4. The second paper of the series[5] was devoted principally to a study of the hardenability characteristics and properties of irons cast as standard 1.2 in. bars from five electric furnace heats having the nominal composition of 3.0 per cent total carbon, 2.0 per cent silicon with 0.7 to 0.9 per cent combined carbon. In this investigation it was found that treating the molten iron in the ladle with about 0.02 to 0.03 per cent aluminum resulted in an iron of shallow hardening characteristics and fine grain size.

5. The results also indicated that the conditions of oxidation of the iron prior to the aluminum addition were also very important in securing the maximum effect of the aluminum addition on hardenability and grain size. In general, the deep hardening coarse-grained irons showed higher as-cast transverse and tensile strength, a somewhat higher as-cast hardness and a lower rate of deflection and impact strength compared to the shallow hardening fine-grained irons. Although the magnitude of effects of grain size and hardenability variations on as-cast properties was less than reported in the first paper, it seemed logical to attribute much of this difference to the great difference in the section sizes of the test bars used in the two investigations; in other words, it appeared as if the effects of grain size and hardenability variations on physical properties increased as the section size of the casting decreased, a conclusion which might not be altogether unexpected.

6. It was suggested at this time that the shallow hardening and deep hardening irons might be expected to show considerable differences in the overall rate at which austenite decomposed to form pearlite or any of the other transformation products, the

[5] Murphy, D. W., and Wood, W. P., Trans. American Foundrymen's Association, vol. 45, pp. 571-598, 1938.

previous results indicating that the shallow hardening irons should possess the higher rate. In a third paper[6] data on the austenite decomposition at 1000°F. were presented on two of the irons
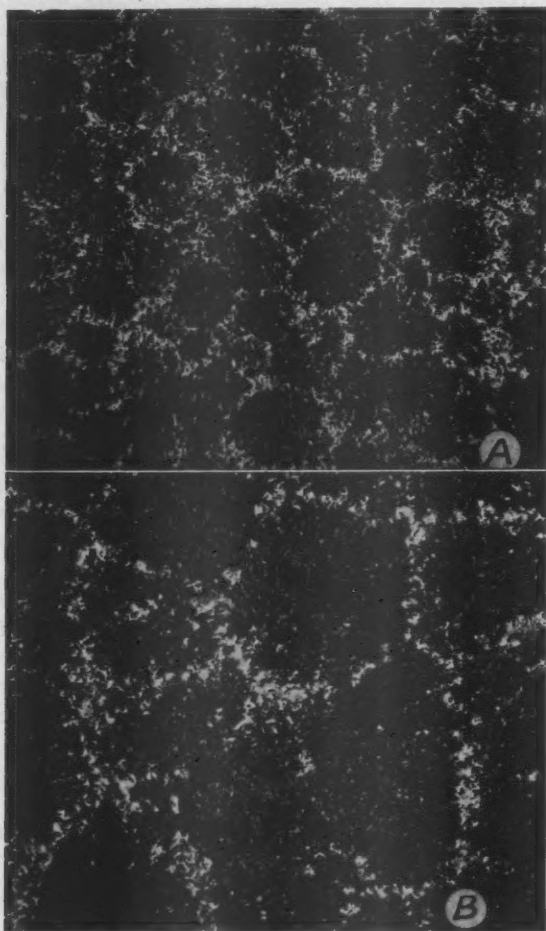
FIG. 2—MACRO GRAIN SIZE OF IRONS.
(A)   MACRO GRAIN SIZE OF IRON E-7; QUENCHED FROM 1600°F., INTO LEAD AT 1000°F. AND HELD FOR 75 SEC. ETCHED WITH 4 PER CENT NITAL—X-80.
(B)   MACRO GRAIN SIZE OF IRON E-4; QUENCHED FROM 1600°F. INTO LEAD AT 1000°F. AND HELD FOR 120 SEC. ETCHED WITH 4 PER CENT NITAL—X-80.
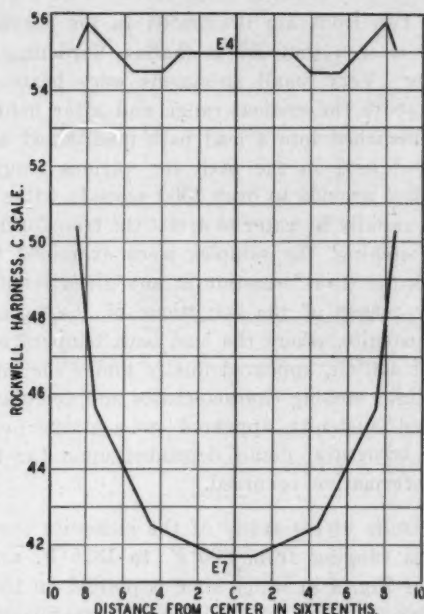
discussed in the second of the series. This investigation showed that the shallow hardening iron did possess a higher overall rate of transformation in comparison with the deep hardening iron. However, the higher overall rate of transformation of the fine-grained shallow hardening iron was principally due to the fact that a shorter time was required to start the austenite decomposition for this iron.

7. The logical extension of the work reported in that paper to other subcritical temperatures would lead to the formulation of typical S-curves, similar to those obtained by Davenport and Bain[7] and their co-workers for steels. In continuing this study the main portion of the work was done on irons E4 and E7 (3 per cent carbon, 2 per cent silicon) previously described in a paper[5] before the American Foundrymen's Association in 1937. The grain size characteristics are shown in Fig. 2, iron E7 being fined grained and E4 coarse. The hardenability character-

[7] Davenport, E. S., and Bain, E. C. Trans. A. I. M. E. Iron and Steel Division. Vol. 90. p. 117-154 (1930) E. C. Bain. Trans. A. I. M. E. Vol. 100, p. 13-46 (1932).

istics of these two irons are illustrated in the curves of Fig. 3 from which it is seen that E7 is shallow hardening and E4 is deep hardening. Very small specimens were heated to a temperature well above the critical range and after holding for the desired time, quenched into a lead bath maintained at a selected temperature and held in the bath for various lengths of time ranging from five seconds to over 2000 seconds, after which they were quenched rapidly in water to arrest the transformation. After polishing and etching, the samples were examined to note the extent to which the transformation at any given temperature had proceeded. By reason of the conditions of the experiment, untransformed austenite, where the lead bath temperatures were in excess of about 400°F., appeared finally under the microscope as martensite of light etching characteristics and acicular structure; the transformed austenite appeared as a darker etching constituent whose structural detail depended upon the temperature at which transformation occurred.

8. The results of the study of the austenite transformation at temperatures ranging from 300°F. to 1325°F. are expressed in the curves of Fig. 4 in which time is plotted on the horizontal logarithmic scale and the temperature of transformation on the vertical scale. The curves at the left (full for the fine-grained iron E7 and dashed for the coarse-grained iron E4) show the time required for the transformation to start at any given temperature while those at the right give the time required for virtual completion of the reaction. For example, with the quenching bath at 1100°F. samples of iron E7 previously heated to 1550°F. required approximately 7.5 seconds before the existing austenite showed any traces of transformation to pearlite and at 1100°F., some austenite remained untransformed until the periods of holding were 30 seconds or greater. Within the area to the left of the starting curves the austenite has not started to transform and the quenched structures in this field are fully martensitic. Between the two sets of curves austenite and its decomposition products co-exist so that the finally quenched structures are mixtures of martensite derived from the remaining austenite and the transformation product characteristic of the temperature of transformation. In the area to the right of the curves no austenite remains and the finally quenched structures are those characteristic of the temperature of the transformation. It will be observed
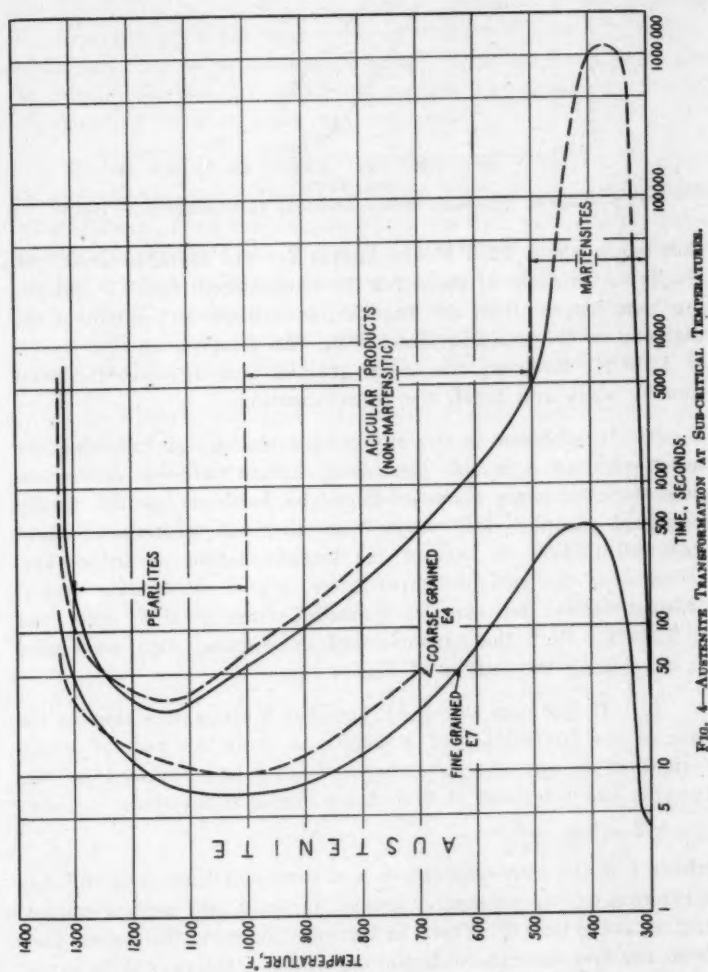
Fig. 4.—Austenite Transformation at Sub-critical Temperature.
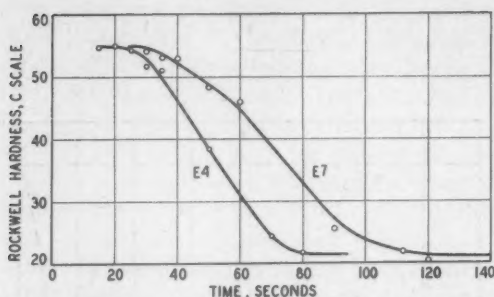
FIG. 5—HARDNESS VARIATIONS DURING AUSTENITE TRANSFORMATION AT 1285°F.

that below about 1250°F. the curves for the coarse-grained iron E4 lie to the right of those for the fine-grained iron E7, indicating that longer times are required to initiate and complete the reaction in the coarse-grained iron. At temperatures in excess of 1250°F., however, the coarse-grained iron requires the least time to start and finish the transformation.

9. In addition to the microscopic method of following the transformation it is also possible to utilize hardness determinations for the same purpose since the hardness of the finally quenched samples will vary from that of martensite, about Rockwell C54-56, to that of the transformation products characteristic of the various temperatures. Fig. 5 shows data typical of this method for samples quenched from 1550°F. into lead at 1285°F. Both the hardness and microscopic data were used in developing the curves of Fig. 4.

10. It has been shown by previous investigators that in the case of the breakdown of austenite in steel the rate of transformation at any given temperature can be expressed by the formula characteristic of first order chemical reactions, namely:

$$k \; \theta = \log \; \frac{a}{a-x}$$

where k is the rate constant, θ is a corrected time, a is the concentration of the substance present initially and $a-x$ its concentration at the time θ. (Theta, in this case, denoting the elapsed time from the first perceptible beginning of the reaction). As a consequence, if the value of $\log \frac{a}{a-x}$ be plotted against θ a straight line whose slope is equal to k should be the result, if the reaction is of the first order type.

11. Fig. 6 shows the data for iron E7 at three temperatures, 800°F., 900°F., and 1200°F. plotted in this manner. It is evident that in cast irons the same law will very satisfactorily express the data. The intercept of the linear expression of the data on the time axis gives the time required to start the transformation, while the time to attain approximately complete transformation is given by the point at which the straight line attains a value of 40, denoting about 97.5 per cent completion.

12. It should be pointed out here that the time required to start the transformation as given by this method may be somewhat different than the time at which the first products of transformation can be seen. In reality this time is an incubation period in which nuclei are forming and growing slightly and it is only after the completion of this period that the reaction may proceed at a constant rate. Ordinarily it is possible to detect some transformation before the end of the incubation period.

13. Photomicrographs of partially and completely transformed specimens obtained at several temperatures are shown in Figs. 7 to 13. For the most part these are quite similar to structures observed in eutectoid steels[7] with the possible exception of those obtained at about 800 to 900°F. (Fig. 10a and b) and in the region of 1300°F. (Fig. 7).

14. From about 900°F. to 800°F. the structures obtained
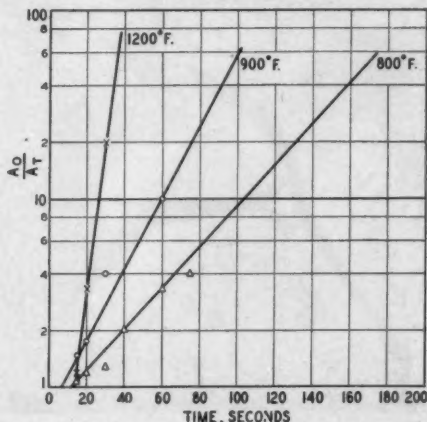
FIG. 6—DATA ON MICROSCOPIC OBSERVATION OF TRANSFORMATION.

are coarser in appearance than those obtained with steels in the
same temperature interval as shown in Fig. 10. As the tempera-
ture of quenching is lowered further, the constituent increases
in fineness of dispersion until the familiar fibrous structure
(Villela, Guellich and Bain's[8] "unnamed aggregate") is attained
at about 700°F. as shown in Fig. 11. The whole series of struc-
tures obtained in the temperature interval of 900°F. to about
500°F. is evidently continuous and varies chiefly in the degree
of fineness of the dispersion and not in the mode of transforma-
tion. The same characteristic variation in degree of dispersion is
of course found in the pearlite structures, the lamellae occurring
in finer form and more densely packed as the quenching bath
temperature is lowered from about 1300°F. to 1000°F.

15. The other structural peculiarity found in the samples
quenched into lead at temperatures ranging from 1285°F. to
1325°F. is that some austenite remains untransformed even after
long periods of holding at these temperatures. At 1285°F. the
amount of austenite remaining untransformed as evidenced by
the residual martensite is extremely small but at 1325°F. the
amount is considerably larger. This arises from the fact that in

[8] Trans. Am. Soc. Metals 24, p. 225, 1936.
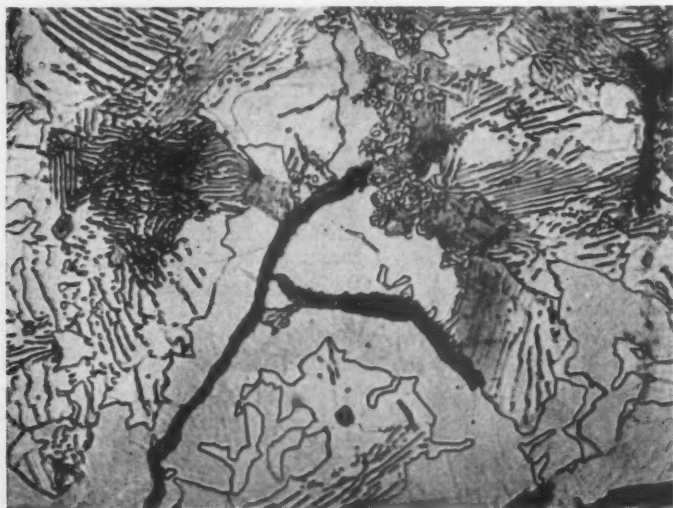
FIG. 7—PARTIALLY TRANSFORMED STRUCTURE OBTAINED BY QUENCHING FROM 1550°F. INTO
LEAD AT 1325°F. ETCHED WITH 4 PER CENT NITAL—X1000.
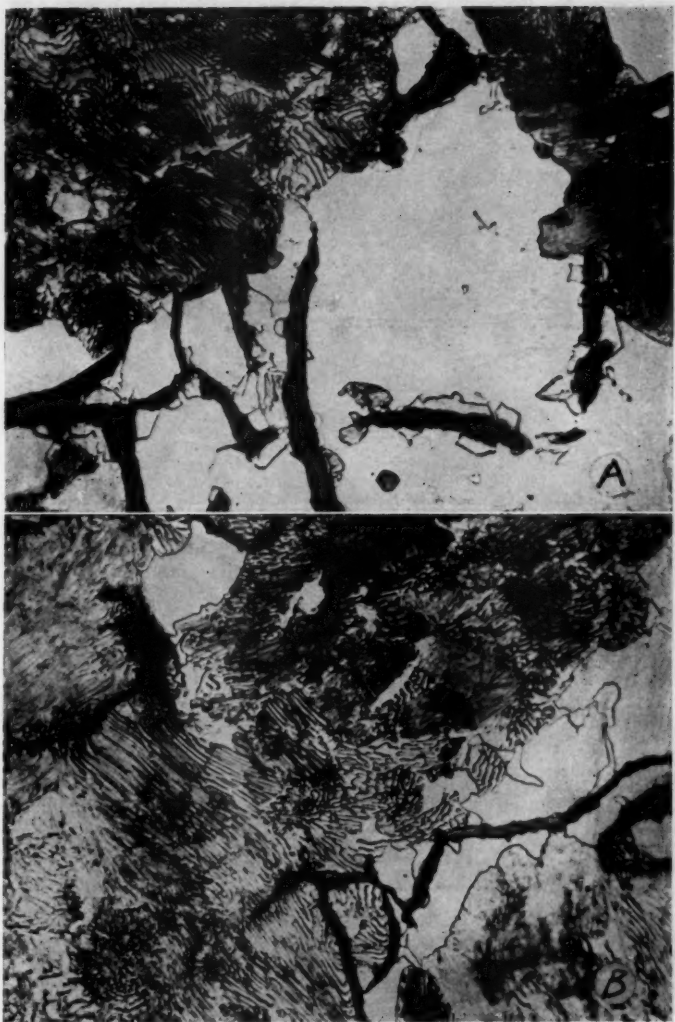
Fig. 8a—Partially Transformed Structure Obtained by Quenching from 1550°F. into Lead at 1285°F. Etched with 4 per cent nital—X1000.

Fig. 8b—Fully Transformed Structure Obtained by Quenching from 1550°F. into Lead at 1285°F. Etched with 4 per cent nital—X1000.
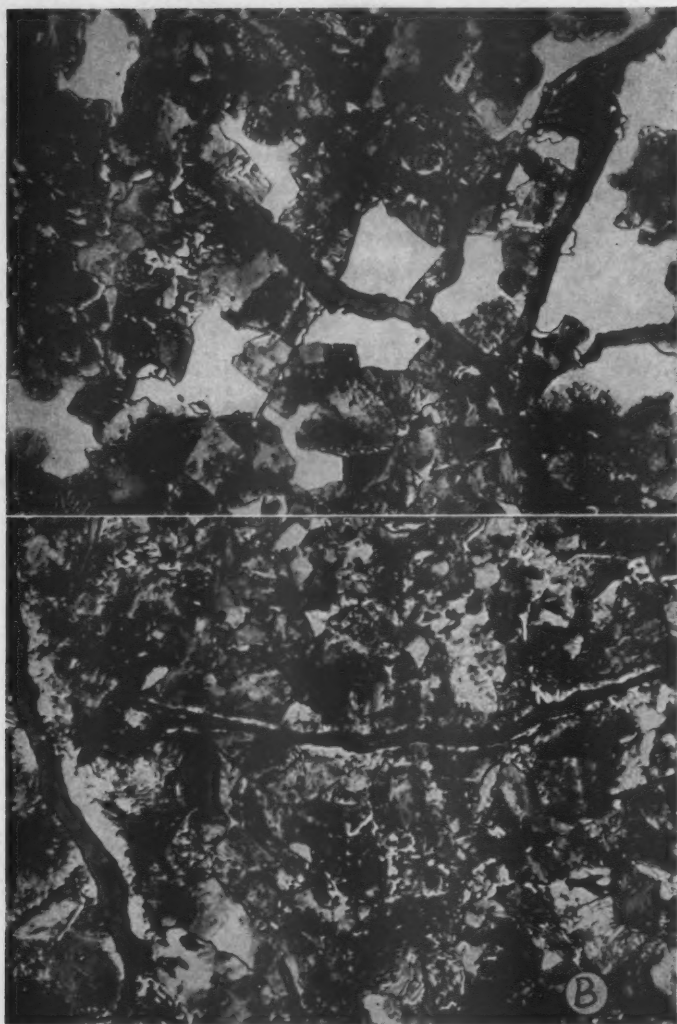
FIG. 9A—PARTIALLY TRANSFORMED STRUCTURE OBTAINED BY QUENCHING FROM 1550°F. INTO LEAD AT 1100°F. ETCHED WITH 4 PER CENT NITAL—X1000.

FIG. 9B—FULLY TRANSFORMED STRUCTURE OBTAINED BY QUENCHING FROM 1550°F. INTO LEAD AT 1100°F. ETCHED WITH 4 PER CENT NITAL—X1000.
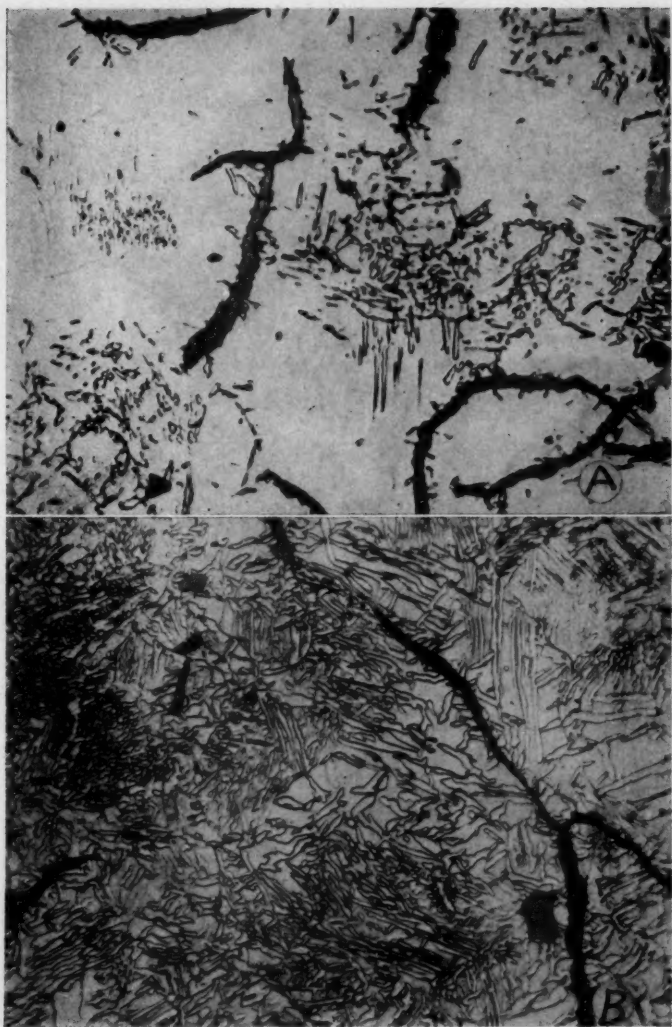
Fig. 10a—Partially Transformed Structure Obtained by Quenching from 1550°F. into Lead at 900°F. Etched with 4 per cent nital—X1000.

Fig. 10b—Fully Transformed Structure Obtained by Quenching from 1550°F. into Lead at 900°F. Etched with 4 per cent nital—X1000.
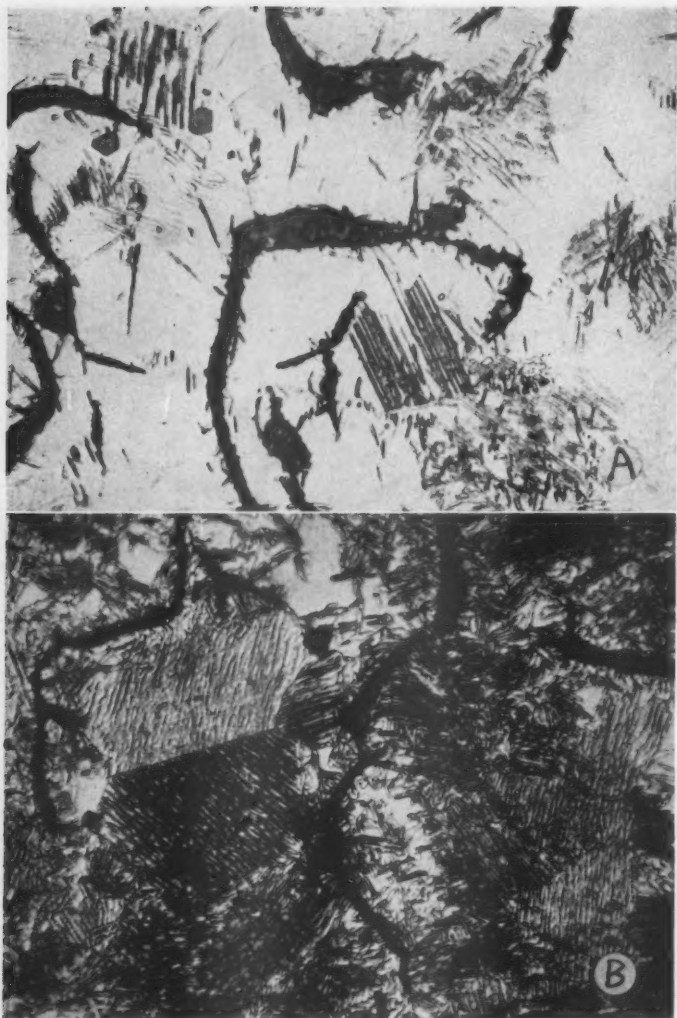
FIG. 11A—PARTIALLY TRANSFORMED STRUCTURE OBTAINED BY QUENCHING FROM 1550°F. INTO
LEAD AT 700°F. ETCHED WITH 4 PER CENT NITAL—X1000.

FIG. 11B—FULLY TRANSFORMED STRUCTURE OBTAINED BY QUENCHING FROM 1550°F INTO
LEAD AT 700°F. ETCHED WITH 4 PER CENT NITAL—X1000.
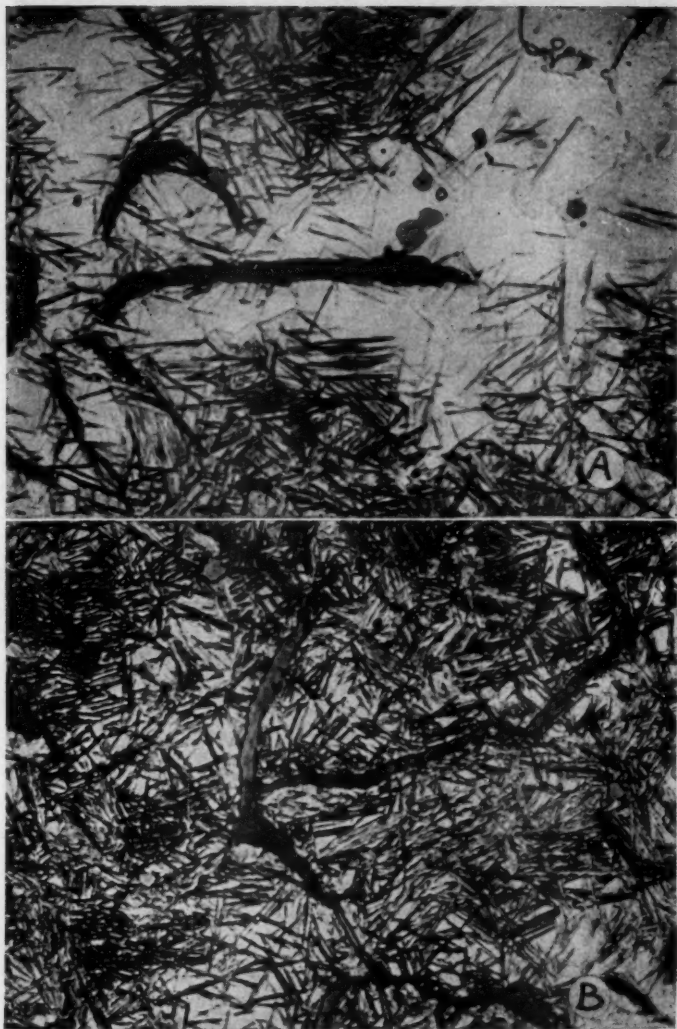
Fig. 12a—Partially Transformed Structure Obtained by Quenching from 1550°F. into Lead at 500°F. Etched with 4 per cent Nital—X1000.

Fig. 12b—Fully Transformed Structure Obtained by Quenching from 1550°F. into Lead at 500°F. Etched with 4 per cent Nital—X1000.

FIG. 18—FULLY TRANSFORMED STRUCTURE OBTAINED BY QUENCHING FROM 1550°F. INTO
METAL BATH AT 500°F. ETCHED WITH 4 PER CENT NITAL—X1000.

the ternary system Fe-Si-C, a region exists in which α Fe, austenite and carbides are in equilibrium, the relative amounts depending in part, upon the temerperature which is chosen. In the experiments as carried out the α Fe and carbides form the familiar pearlite structures and the austenite transforms to martensite on quenching into water.

16.   The critical points on cooling for these irons were observed at 1315°F. for E7 and 1325°F. for E4, which probably accounts, in part at least, for some differences in behavior between these two irons in this temperature range. It might be pointed out here that the curves showing the end of transformation in Fig. 4 should be asymptotic to the critical temperatures.

17.   Essentially the transformation of austenite in cast irons at subcritical temperatures is the same as for steels. From the critical range down to 1000°F. the transformation proceeds according to the nodular mode by which pearlite of varying degrees of dispersion is formed. Below 900°F. the transformation proceeds according to the acicular mode, in which two variations are discernible, depending upon the temperature of transforma-

tion. In the range from about 500°F. to 900°F. diffusion of carbon takes place with precipitation occurring on the austenite cleavage planes. Below about 400°F., however, the only process which may occur is the change in the iron lattice from the γ to the α spacing and precipitation of carbides is prevented by the slowness of diffusion. In this temperature region martensite is the product of the austenite transformation.

18. A further check upon the validity and consistency of the data from which Fig. 4 was drawn can be obtained from the fact that log k plotted against 1/T, where T is absolute temperature, has been found to be a linear function in first order reactions for limited temperature ranges. When the data are treated in this manner as shown in Fig. 14 it will be seen that in the temperature range from about 500°F. to 1100°F. the results are well represented by a straight line. Deviations from the linear relationship such as found below 500°F. and above 1100°F. are found in most first order reactions at the extremes of temperature.

19. The austenite transformation curve as shown in Fig. 4 is extremely useful in interpreting the heat treating process for gray irons and, as will be discussed later, it may also be advantageous to consider the relationship of this type of curve to such properties as hardness and strength in the as-cast condition. It is well known that with gray irons, comparatively large sections may be thoroughly hardened by quenching in oil even though no
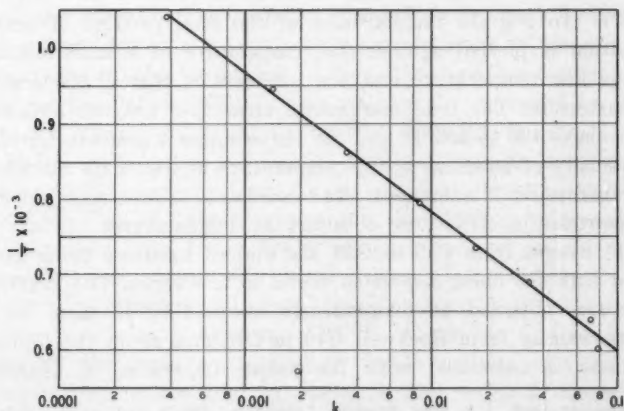
Fig. 14—Variation of Rate Constant of Austenite Transformation With Temperature.

additional alloys are used. In the case of steels, however, deep
hardening is very difficult to attain even with very drastic
quenching, except where alloys are present. As an example the
authors have hardened completely, by means of oil quenching,
irons having a section size of 1.2 in. while in the case of steel
rather considerable amounts of alloying elements must be present
to enable such complete hardening under similar conditions.

20. In order to secure a completely martensitic structure the
cooling rate in quenching either iron or steel must be fast enough
to prevent any transformation of the austenite at temperatures
in the vicinity of 1100°F. to 900°F. where in general the trans-
formation of austenite starts most rapidly. Reference to Bain's[9]
curves for a eutectoid steel will show that it is necessary to cool
the steel from 1100°F. to 900°F. in a period of about one second
to prevent any reaction in this temperature range. Such cooling
rates are capable of realization only when small sections and
drastic cooling is used. On the other hand the fine-grained iron
whose S-curve is given in Fig. 4 can be cooled from 1100°F. to
900°F. in seven seconds without allowing any transformation at
these temperatures. Consequently much larger sections are capable
of being hardened completely even with a comparatively mild
quench such as into oil. In other words, the critical cooling rate,
i.e., that rate which is just necessary to prevent the transforma-
tion in the range 1100°F. to 900°F. is much lower for cast irons
than for steels.

21. In Fig. 15 the hardness of the final product of trans-
formation is plotted against the temperature of transformation.
At the low temperature end the hardness is that characteristic
of martensite. The true martensitic structures are not obtained
much above 400 to 500°F. and the curve shows a gradual decrease
in intensity of hardness as the temperature at which the austenite
is transformed is increased. The hardness of the acicular but
non-martensitic structures obtained at temperatures of 500 to
900°F. ranges from C40 to C30, the highest hardness being asso-
ciated with the finest dispersion found in this series. The pearlitic
structures obtained at temperatures above 1000°F. have hard-
nesses ranging from Rockwell C40 to C13 and again the highest
hardness is associated with the finest dispersion, of lamellae

* Research Staff, U.S. Steel Research Laboratory, Metals and Alloys, Vol. 8,
p. 22-24. (1937).
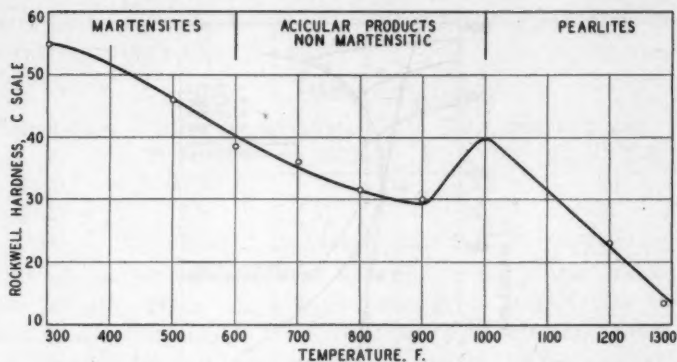
FIG. 15—VARIATION OF HARDNESS OF TRANSFORMATION PRODUCTS, WITH TEMPERATURE OF TRANSFORMATION.

obtained in the region of 1000°F. Although data on other physical properties are not available at present, it is logical to conclude that similar ranges exist in such properties as tensile and transverse strength depending entirely on the character of the pearlitic matrix, that is to say, on the temperature at which the transformation of austenite occurred. Subsequent study might well show that some classes of castings could be considerably improved in physical properties by definitely controlling the austenite transformation temperature in order to secure a given type of pearlite.

22. Although the discussion up to this point has been confined to illustrating the behavior of irons during heat treatment, it is indicated that in addition to their usefulness in interpreting the process of the hardening of iron, the S-curves particularly those portions above 1000°F. may well furnish a logical explanation for variations in hardness and strength in the as-cast condition, when other conditions, such as graphite distribution and chemical composition, are for all practical purposes, constant. Curve A, Fig. 16, might represent a slow cooling rate such as that obtained in a fairly heavy section. Such a cooling curve would produce a partially ferritic iron of low hardness and structurally rather weak, since transformation of the matrix on cooling would begin at about 1300°F. where the cooling curve intersects the S-curve. A slightly faster cooling rate, as, for example, that represented by curve B, and brought about possibly by a smaller section, would produce a pearlitic iron of
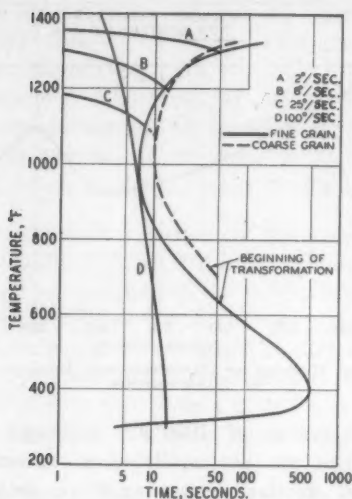
FIG. 16—INFLUENCE OF COOLING RATE ON TEMPERATURE OF BEGINNING OF TRANSFORMATION.

medium hardness and fair strength. The austenitic matrix in this case would not begin to transform until a temperature of about 1200°F. was reached at the intersection of the S-curve and the given cooling curve. A still faster cooling rate represented by curve C would produce a relatively harder and stronger casting again completely pearlitic in structure. In this case the matrix would not begin to break down to form pearlite until a temperature of about 1100°F. was attained, in which region the finest and hardest pearlite structures are formed. Curve D represents a condition which is theoretically possible, but actually difficult to attain in unalloyed castings. The matrix of austenite does not begin to transform on cooling until low temperatures are reached. In this case the cooling curve of the casting does not intersect the S-curve until a temperature of about 300°F. is reached in which region martensite is formed. The resulting casting would possess the maximum hardness and considerable, although not necessarily, great strength.

23. The position of the S-curve as shown for the fine-grained iron in Fig. 4 may be shifted to right or left by factors such as grain size and chemical composition. The influence of

grain size on the transformation at a few temperatures is shown
in the following table:

Table 1

Temperature	Time, seconds to start		Time, seconds to end	
°F.	F. G. E-7	C. G. E-4	F. G. E-7	C. G. E-4
1200	12.5	15	33	39
1100	7.5	11	30	35
900	8	13	102	129

Although the transformation of austenite in the course grained
iron at any given temperature requires a longer time to
start than in the case of the fine-grained iron, after once start-
ing, the time for required completion is nearly the same so that
the general shape of the curves is essentially unchanged. The
chief effect of the larger grain size is to shift the S-curve to the
right and somewhat upward as compared to the position for a
fine-grained iron as shown in Fig. 4.

24. Returning for a moment to a discussion of Fig. 16, it
will be apparent that a shift to the right of the S-curve to the
position of the dotted curve such as that caused by increasing
grain size would cause the given cooling curves to intersect the
new S-curve at lower temperatures. Consequently, castings hav-
ing the cooling rates illustrated by curves A, B, and C, begin to
transform at lower temperatures than in the case of the fine-
grained iron and their properties judging from the range of
hardness obtained in these experiments would be increased. In
other words, for the same cooling rate it would seem logical on
this basis to expect that a coarse-grained iron would show higher
strength and greater hardness as-cast than a fine-grained iron,
since the austenite in the coarse-grained iron would begin to
transform at a lower temperature under given cooling conditions
and provided of course that other factors of composition and
structure remained fairly constant.

25. Data contained in previous papers by the authors have
shown that coarse-grained irons have higher tensile and transverse
strength and hardness than do fine-grained irons. For example,
in very small sections[4] a coarse-grained iron was found to
have a strength of 53,000 lb. per sq. in. and a Rockwell hardness
of C30 compared to a tensile strength of 37,500 lb. per sq. in.
and a hardness of C24 for a fine-grained iron of similar composi-
tion. A smaller difference in properties was observed in larger

sections[5] but again the coarse-grained deep hardening irons
showed the higher properties.

26. The results just mentioned strongly support the ex-
planation previously advanced that the temperature at which the
austenite transforms as the casting cools down exerts a very
powerful influence on the properties of the casting and further-
more that a change in conditions which causes the S-curve to be
displaced toward the right, such as an increase in grain size,
will in general increase such properties as hardness and tensile
and transverse strength. In view of these results it appears as
though in addition to a more comprehensive knowledge of the
mechanism of graphite formation and the factors which influence
that formation it will be necessary to study much more exten-
sively the factors which influence the matrix behavior in order
to use the metal to the greatest advantage.

27. In considering the heat treatment of cast irons it is now
apparent that differences in grain size will result in large varia-
tions in response to hardening. The critical cooling rates of the
irons shown in Fig. 4 are such that the iron must pass through
the range from 1100°F. to 900°F. in from 7 to 10 seconds. These
curves show that a coarse-grained iron requires a much slower
critical rate of cooling to effect complete hardening, since a period
of about 10 seconds to cool from 1100°F. to 900°F. is sufficient
to prevent any transformation in this range. The fine-grained
iron, however must be cooled in 7 seconds through this range
to prevent transformation. In other words, a considerably larger
section size in the coarse-grained iron could be hardened com-
pletely than would be the case with the fine-grained type using
the same quenching treatment; or for a given section size and
quenching treatment the coarse-grained iron would tend to harden
more deeply than the fine-grained type. It will be recalled that
precisely this difference in hardening power between the coarse-
grained iron E4 and the fine-grained iron E7 is shown in Fig. 3.

28. It has been indicated that composition as well as grain
size affects the position of the S-curve. A very few experiments
with a higher silicon iron, for example, indicate that the increase
of silicon over two per cent causes a shift of the curve to the
right. On the basis of comparison with existing curves for steel
with about the same combined carbon content, it seems logical to
conclude that decreasing the silicon below two per cent would

cause a shift of the curve to the left, more closely approaching the position of the eutectoid steel curve[9].

29. In view of the similarity of the curves for steel and cast iron, it might be predicted qualitatively that alloying elements such as nickel, chromium, molybdenum, and manganese would also cause a shift of the S-curve to the right, and probably also cause a displacement on the temperature axis, although the degree of the shifts might be different than for steels. In general this view seems reasonable since it is well-known that the addition of sufficient nickel, nickel and molybdenum, nickel and chromium, or nickel and manganese can produce castings with a martensitic structure as-cast. In these cases the addition of these alloys has shifted the S-curve sufficiently far to the right so that the conditions illustrated by curve D, Fig. 16, are realized, namely, that the cooling rate of the casting is such that the beginning austenite transformation curve is not intersected on cooling until a temperature of about 300°F. has been attained.

30. The relationship of the use of alloys to facilitate the heat treatment of irons and their probable effects on the position of the S-curve is worthy of comment. Considering that most alloys would, if added to the iron described in this paper, probably result in a shift of the S-curve to the right it can be said that they lower the critical cooling rate which is necessary to secure hardening. In other words, larger sections in alloyed irons could be successfully hardened by a given quenching treatment than would be possible in the unalloyed irons because of the effect of the alloys on the critical cooling rate. The corollary to the above statement is that for a given size of section and degree of hardening, the addition of alloys to the iron makes possible the use of milder quenching than would be necessary with un-alloyed irons. Several types of alloyed irons which possess sufficiently low critical cooling rates so that they harden to a full martensitic structure by a simple air cool are in use today, particularly where the nature of the casting is such that a more rapid quench might produce undesirable distortion or even breakage.

31. It has been shown that increasing grain size causes the S-curve for a given iron to be displaced toward the right and that through this influence on the position of the curve, the as-cast properties such as strength and hardness, are increased since with given cooling conditions the austenite of the coarse-

grained iron will transform at lower temperatures than the fine-grained iron resulting in pearlite of higher hardness and strength. Composition variations will undoubtedly influence the position of the S-curve and consequently, other factors being equal, alloying a given iron would probably tend to increase the as-cast hardness and strength of the iron by reason of the effect of the alloy in shifting the S-curve to the right. There is an indication that a considerable part of the effect of alloys on properties is associated with their effects on the austenite transformation and that much useful information might be developed by studying alloying from this point of view. And in addition to the influences of grain size and composition which have been discussed, other factors such as section size and melting practice (including deoxidation) will probably be found to have an influence on the austenite transformation curves, and hence on the properties of the iron.

CONCLUSIONS

32. The following conclusions seem justified by the data presented in this investigation:

(1) That the austenitic matrix of a cast iron on cooling from a temperature above the critical will show the same type of behavior as illustrated by the S-curve for steel and, in fact, such curves can be drawn for cast irons.

(2) That it seems probable that the cooling rate in a given casting and the position of the S-curve for the given iron will determine to a considerable degree the matrix structure and the physical properties such as hardness and strength in the as-cast condition. It is recognized of course that the type and distribution of graphite is also a factor of great importance in determining the properties of an iron.

(3) That increasing grain size and probably the presence of alloys such as nickel, chromium, molybdenum and manganese will cause the S-curve for a given type of iron to be displaced to the right.

Heat Treating Zinc Bronze Pressure Castings to Close Leakage

By H. Fleck*† and T. C. Bunch*†, Honolulu, T. H.

Abstract

In producing castings of "gun metal," leakage under pressure frequently is encountered. The authors in this paper report their experiments in "closing up" this type of leakage by an annealing treatment. Results of previous work along these lines is discussed. The usual types of troubles encountered are listed, described and remedies proposed. Microscopic porosity is discussed in detail. Test bars from a series of experiments were made and tested, and the results reviewed. Best annealing results are reported when annealing between 1300° and 1400°F. Conclusions drawn from the author's experiences are given.

1.   Bronze, of the composition commonly known as "Gun Metal," finds wide application for castings subjected to various degrees of pressure, either hydraulic, steam or pneumatic. Frequently these castings are rejected at the test plant due to their inability to withstand the desired pressure. Considerable experimental work has been published concerning the heat treatment of gun metal both to insure pressure tightness and improve its physical properties. Jenkins[1] has presented an excellent correlation of this work.

2.   The purpose of this paper is to report our experiments in "closing-up" leakers, by annealing, and to stress the possibilities of this method as applied to shop practice. We also shall deal briefly with some of the more prevalent foundry troubles encountered when pressure tight castings are manufactured.

3.   During routine production of castings, the foundryman may meet with several types of trouble. Some of these can be

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checked easily in the foundry and always are obvious. Briefly, these are:

(A). "Copper Spots"

4. These spots form under risers on flanges, etc., and are caused by a drainage of metal away from the section in question. An average chemical analysis is copper, 90 per cent; tin, 5.1 per cent; and zinc, 3.3 per cent as compared with copper, 87.5 per cent; tin, 8.3 per cent; and zinc, 4.2 per cent in the body of the casting. This indicates that the later freezing material, higher in tin, drained away and fed down into the lower sections. The riser, obviously of improper dimensions, froze prematurely and did not continue to feed the casting. When "copper" spots occur at the junction of the gate and casting, it is necessary to either enlarge the gate or change its location in respect to the adjacent section thickness of the casting.

5. Pouring temperatures are extremely important in this particular case. To permit risers to feed satisfactorily, it is necessary that the casting be poured at an initial high temperature or that hot metal be poured down the risers after the metal has been brought up to a certain level in the mold. This last method is more desirable as it favors progressive solidification. At this foundry, the pouring range for gun metal lies between 2000 and 2200°F. Each type of casting is considered an individual problem and the exact pouring temperature for any particular casting is based upon the careful consideration of a number of factors.

(B). Blow Holes, Gas Pockets, Scabs, etc.

6. These flaws are caused by faulty molding technique such as: improper venting, damp molds, mold rammed either too hard or too soft, cores too hard or carelessly dried. Bubbling of metal in the risers and, in extreme cases, ejection of the metal from the mold is nearly always indicative of a casting that will leak under pressure. As in the preceding case, this type of trouble is primarily a molding problem.

(C). Gassy Metal

7. With proper melting conditions and the use of deoxidizers such as manganese and phosphor-copper, gassy metal is unlikely in the type of alloy under discussion. However, should it occur, it is easily detected. Risers will not feed properly and, in extreme cases, the released gases actually will cause the metal to "purge," or be forced back up the riser. This "purging" occurs while the

casting is freezing. If this condition occurs, it is a reliable indication that the casting will be porous.

(D). Microscopic Porosity.

8. Microscopic porosity in the castings is another source of trouble and is most difficult to detect before testing. In this case, an apparently perfect casting will "weep" under hydrostatic pressure. It is believed that any or all of the following factors may promote this condition:

9. *(a) Intercrystalline Hairline Cracks:* The formation of these cracks generally is attributed to restricted contraction in the mold. They usually occur at the base of deep flanges and at the intersection of right angled sections. Adequate fillets, soft cores and molds rammed with the necessary precautions to permit unrestricted contraction do much to alleviate this evil.

10. *(b) Porosity Caused by Metal Drainage to Lower Levels During Solidification of the Casting:* Gun metal particularly is susceptible to this type of porosity, since its freezing range covers approximately 350°F. The fact that the alloy freezes with a typical dendritic structure further complicates the problem of drainage and enhances the possibility of voids around the dendrites.

11. *(c) Cooling Condition:* The physical properties and soundness of gun metal depend upon the rate of solidification[2]. The pressure tightness is affected to a considerable extent by cooling differences such as would exist between chill castings, green and dry sand castings. The surface of the casting which has been chilled by the mold has a fine grained exterior, as compared with its interior, and is pressure tight[3]. In one sense, the depth of the chilled surface is a measure of the pressure resistance of the metal. It is necessary to form this chilled surface to such a depth that subsequent machining will not remove this fine grained structure entirely and expose the interior metal which is subject to intercrystalline porosity. Our recent experience with gun metal liners (10$\frac{7}{8}$ in. I.D. x 13$\frac{1}{4}$ in. O.D. x 6 ft. 6 in. long) that had machine allowances, of 7/16 in. on the outside and $\frac{1}{4}$ in. bore, on radius, convinced us that excessive machine allowance should be avoided. After machining, these liners leaked profusely at 40 lb. per sq. in. pressure. Annealing closed up the greater part of this leakage. By reducing the machine allowance to $\frac{1}{8}$ in. (outside and bore, on the radius) and eliminating the pattern taper on these liners, we prevented excessive machining and preserved the pressure tight surface so necessary in this type of casting[4].

12. *(d) Pouring Temperatures:* We have observed that castings poured below 2000°F. are more likely to leak than those poured between 2000 and 2200°F. This is due, we believe, to the relation between pouring temperature and "rate of solidification." This factor should be given serious consideration in establishing the exact casting temperature in any given case.

CLOSING LEAKS BY ANNEALING

13. Castings that leak because of microscopic porosity, as a general rule, can be closed-up completely by annealing within the temperature range 1300 to 1400°F. This range was selected as a result of microscopic examination of samples heat-treated to various temperatures (See photomicrographs). The procedure commonly followed is to machine castings with a "gasket finish" and apply test. Leakers are then annealed and retested before further machining.

14. During the past year, we have closed-up valve bodies, manifolds, fittings, etc., consistently using this method. Castings that leaked upon the application of 60 lb. per sq. in. pressure have retested after annealing to 400 lb. per sq. in. perfectly. Subsequent machining operations proved that the "closing up" was effected throughout and was not due to scale formation, etc.

15. It seems appropriate to point out that peening, soldering and brazing is resorted to in many instances in an effort to overcome this type of leakage. These methods of repair, however, merely close over the surface and provide little assurance that the casting will give satisfactory service. The better method, and the one that we believe fundamentally correct, is that of annealing.

CHECK TESTS ON ANNEALING EFFECTS

16. During our experiments, a series of tests was conducted which included microscopic and physical examination. Specimens for microscopic examination were obtained from two valves, of different chemical analysis, that leaked under hydraulic pressure. Photomicrographs were made (Figs. 1 and 2) of the metal in the "as cast" condition and after annealing to various degrees of temperature.

17. Annealing was accomplished in a laboratory size electric furnace, 2 hr. being allowed for each inch of section thickness, followed by furnace cooling. Temperatures were ascertained with a pyrometer calibrated by the Bureau of Standards. The photo-
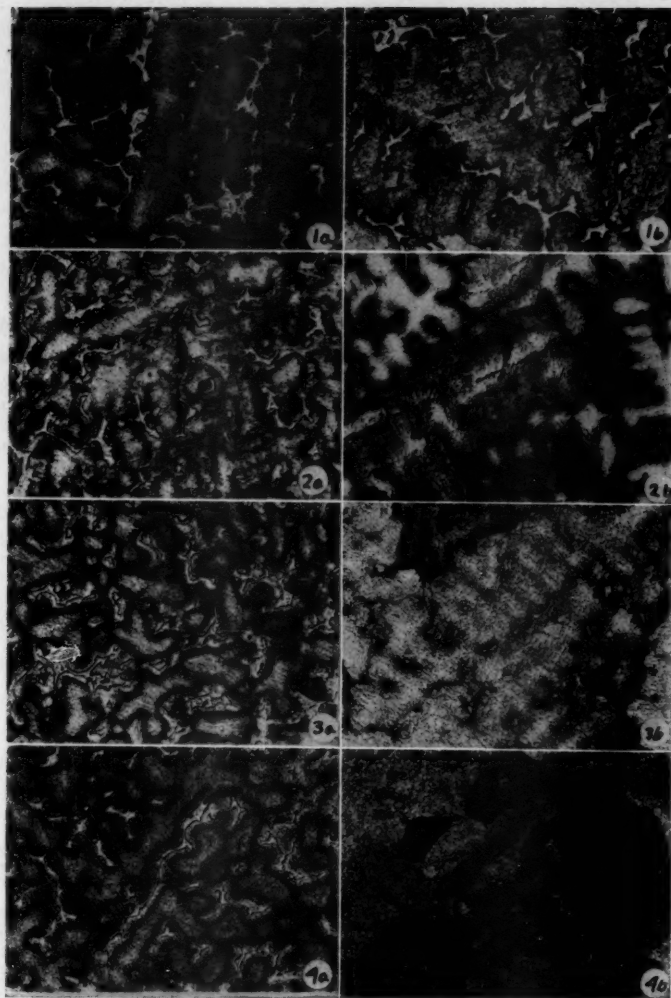
FIG. 1—PHOTOMICROGRAPHS SHOWING EFFECTS OF ANNEALING AT VARIOUS
TEMPERATURES ON TWO "GUN METAL" BRONZES.

	Composition A	Composition B
Cu., per cent	86.4	85.2
Sn., per cent	10.5	10.6
Zn., per cent	2.39	3.4
Ni., per cent	0.59	0.68
Pb., per cent	0.07	0.08

(1) A AND B—AS CAST, DELTA IN ALPHA.—(2) ANNEALED AT 700°F.—SAME AS IN 1A AND
1B.—(3) ANNEALED AT 900°F.—(a) SAME AS 1A AND 2A.—(b) SLIGHT ABSORPTION OF
DELTA.—(4) ANNEALED AT 1100°F.—(a) SLIGHT ABSORPTION OF DELTA.—(b) DELTA NEARLY
ABSORBED—DENDRITES DISAPPEARING.
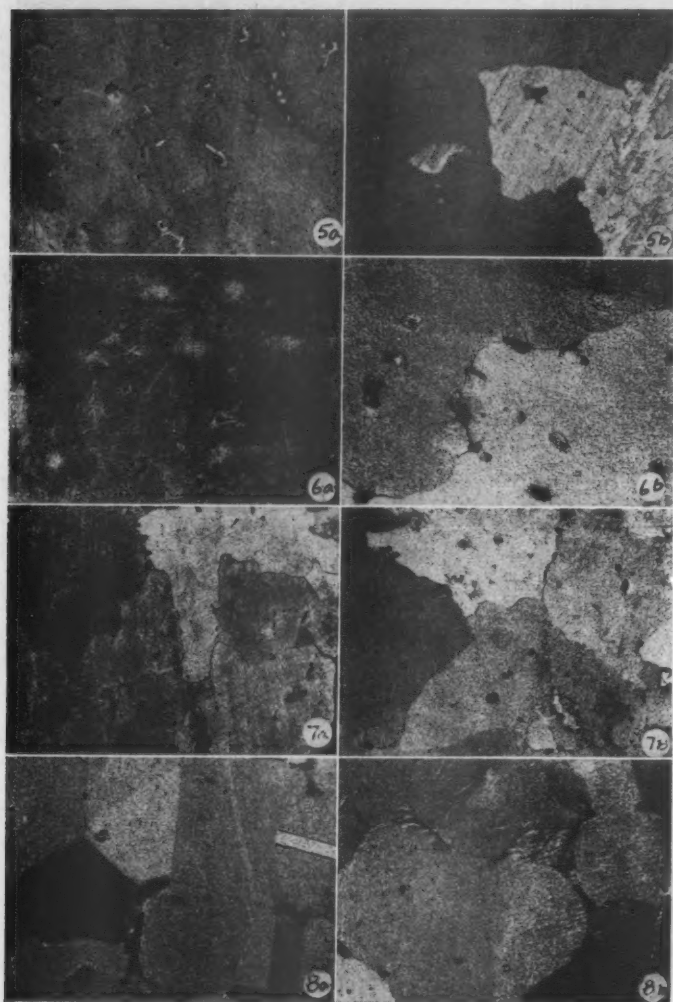
Fig. 2—Continuation of Annealing Series Shown in Fig. 1.
(5) Annealed at 1300°F.—A and B—Dendrites Eliminated and Delta Absorbed.—
Homogeneous Alpha Constituent.—(6) Annealed at 1400°F.—(a) Same as 5A.—(b)
Slight Enlargement of Grain Boundaries as Compared to 5B.—(7) Annealed at
1500°F.—A and B—Grain Boundaries Enlarged as Compared to 6A and B.—(8)
Annealed at 1600°F.—A and B—Further Enlargement of Grain Boundaries as
Compared to 7A and B.

micrographs, as shown in Figs. 1 and 2, were made a magnification
of 100 diameter (reduced 1/3 in the reproduction). The specimens
were etched with ferric chloride.

18. As shown in the photomicrographs, both valves, as cast,
exhibit a typical dendritic structure. Subjected to annealing tem-
peratures up to 900°F., very little change is noted in the micro-
structure of either alloy. The samples annealed at 1100°F. show
a slight to moderate absorption of the delta constituent and
evidence of the gradual disappearance of the dendrites. Annealing
in the range 1300 to 1400°F. results in a complete absorption of the
delta and elimination of the dendritic crystallization. Portevin[5]
has stated that when the alloy reaches a condition of physico-
chemical equilibrium, the delta or tin rich phase will disappear
by being absorbed by the alpha solution; the heterogenity of the
alpha solution will disappear by diffusion.

19. While there is no evidence of grain growth within the
temperature range 1400 to 1600°F., it appears that the grain
boundaries have been somewhat enlarged.

20. Practically, annealing within the temperature range
1300 to 1400°F. results in a closing of voids within the metal and
lessens the chance of hydraulic failure. Several arguments have
been advanced as to the actual mechanism by which the alloy is
made pressure tight.

21. Jenkins[1] states "an opinion has been advanced that
the voids might be closed through an increase in volume of the
matrix taking place concurrently with the absorption of the den-
dritic structure. That is, the homogenized structure may occupy
greater volume than the cast structure."

22. Rawdon[6], working with the alloy (88 per cent copper,
10 per cent tin, 2 per cent zinc), found that crystals, character-
istic of the annealed state, appeared long before the dendritic
structure had disappeared. He further says, "This implies that if
the metal is finally perfectly homogeneous—i. e., adjoining crystals
are of the same composition and structure—there must be a mutual
transfer of material across the crystal boundaries during diffusion."

PHYSICAL RESULTS

23. A series of test bars were poured from two alloys.
Physical results were obtained in the "as cast" condition and
after annealing for one hour at 1325°F. Pouring temperatures

Table 1

DATA FROM TEST BARS OF TWO ALLOYS

Composition = Cu. — 88.69; Sn. — 8.08; Zn. — 2.53; Ni. — 0.70; P. — Trace

Pouring Temperature — 2075° — 2100°F. Accurate ± 20°F.

	Tensile strength lb. per sq. in.		Yield Point lb. per sq. in.		Reduction in Area per cent		Elongation in 2-in. per cent	
	As Cast	Annealed	As Cast	Annealed	As Cast	Annealed	As Cast	Annealed
High	41,850	40,650	19,650	21,000	....	....	48.0	40.0
Average	38,216	39,066	19,183	20,500	40.09	42.00	40.3	36.5
Low	35,650	37,300	18,900	19,900	....	....	36.0	32.5

Composition = Cu. — 86.41; Sn. — 8.02; Zn. — 4.84; Ni. — 0.73; P. — Trace

Pouring Temperature — 2050° — 2075°F. — Accurate ± 20°F.

	Tensile strength lb. per sq. in.		Yield Point lb. per sq. in.		Reduction in Area per cent		Elongation in 2-in. per cent	
	As Cast	Annealed	As Cast	Annealed	As Cast	Annealed	As Cast	Annealed
High	44,025	46,475	19,100	19,600	....	....	66.5	65.0
Average	43,766	46,175	18,516	19,366	55.00	51.00	62.0	61.6
Low	43,500	45,850	18,000	19,100	....	....	60.1	58.0

were ascertained with a Thwing immersion type pyrometer. The test bar pattern* used is illustrated in Fig. 3.

24.   It is apparent that, as shown in Table 1, heat-treatment has improved the physical properties (with the exception of one case noted under Brinell hardness) as well as the pressure tightness. Due to space involved in presenting results on individual bars, it has been deemed advisable to present high, low and average figures.

BRINELL HARDNESS

25.   Some interesting data were brought to light when Brinell hardness tests were conducted on the samples that were used for microscopic examination. It appears that the low zinc alloy (2.39 per cent zinc), while possessing higher hardness as cast, softens rapidly as the annealing temperature is raised. Whereas, the higher zinc alloy (3.39 per cent zinc), with a lower hardness as cast, softened only 12 points throughout the range 700-1600°F. More experimentation is contemplated in regard to ''controlled hardness,'' since this is an important item. Average hardness values are tabulated in Table 2.

SUMMARY

26.   A critical study was made of the possibility of closing up porous castings by heat treatment. The procedure followed was to determine by microscopic and physical examination the annealing temperature that would produce maximum chemical and physical

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*This test bar is the same as Fig. 10-A Federal Stock Catalogue—QQ-M-151-21—Section LV, Part 5.

GATED HERE AND TWO BARS PER MOLD POURED FROM COMMON RUNNER
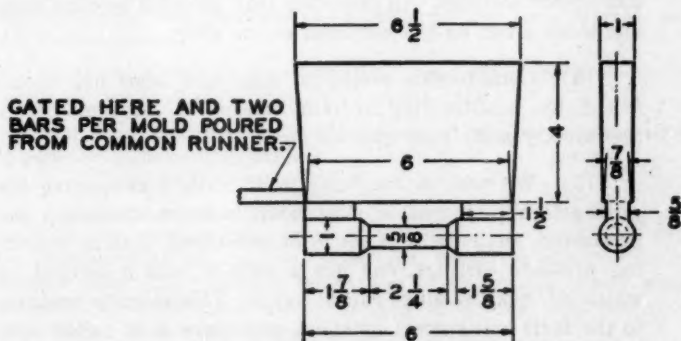
FIG. 3—TEST BAR PATTERN.

Table 2

BRINELL HARDNESS DATA ON SAMPLES USED FOR MICROSCOPIC EXAMINATION

Chemical Analysis Cu.—86.4; Sn.—10.5; Zn.—2.39 Ni.—0.59; Pb.—0.07		B.H.N.*	Chemical Analysis Cu.—85.2; Sn.—10.6; Zn.—3.89 Ni.—0.68; Pb.—.08		B.H.N.*
As cast		96.3	As cast		76.3
Annealed	700°F.	92.6	Annealed	700°F.	74.1
Annealed	900°F.	85.7	Annealed	900°F.	74.1
Annealed	1100°F.	85.7	Annealed	1100°F.	74.1
Annealed	1300°F.	76.8	Annealed	1300°F.	71.5
Annealed	1400°F.	74.1	Annealed	1400°F.	71.5
Annealed	1500°F.	60.5	Annealed	1500°F.	64.6
Annealed	1600°F.	44.9	Annealed	1600°F.	64.6

*Using 10 mm. ball, 500 kg. load, 30 seconds.

homogeneity within the metal. These experiments were followed by the application of this method (for one year) to routine shop production.

CONCLUSIONS

27. The following conclusions were drawn:

(1) Zinc bronze castings that leak because of microscopic porosity, as a general rule, can be closed by suitable heat treatment.

(2) The temperature range for this treatment lies between 1300 and 1400°F. Our standard practice is to anneal at 1325°F. Castings are held at this temperature 2 hr. for each inch of thickness, and furnace cooled.

(3) Heat treatment improved the physical properties of zinc-bronze coatings. An exception may be noted in some cases due to its affect on the hardness of the alloy.

(4) Considerable saving in time and labor can be effected by heat-treating certain classes of castings before preliminary machining and testing.

(5) *We wish to emphasize* that while heat-treating has proven highly efficient, in most cases, under routine shop and production methods, it is not to be construed, that by annealing pressure castings, one has a cure-all and a method by which all leaky castings can be saved. This is quite contrary to the facts and success with this procedure must follow only the most exacting foundry practice.

Bibliography

1. Jenkins, Francis G., *"Zinc Bronze—Composition 'G'*—Treating It for Pressure Tightness," METALS AND ALLOYS, vol. 7, November, 1936.

2. Primrose, H. S., *"Metallography as an Aid to the Brassfounder,"* JOURNAL, INSTITUTE OF METALS, vol. 4, 1910, p. 248.

3. Fawcett, L. H., *"Properties of Bronze,"* (88-10-2 and 88-8-4), METALS HANDBOOK, Cleveland, 1936, p. 1089.

4. Private Communication, L. H. Fawcett, U. S. Navy Yard, Wash., D. C., Nov. 30, 1937.

5. Portevin, Rev. de Me'tal, vol. 6, p. 677, 1913.

6. Rawdon, H. S., *"Microstructural Changes Accompanying the Annealing of Cast Bronze* (88-10-2)," U. S. Bureau of Standards, Technical Paper No. 60, 1916.

DISCUSSION

Presiding: HAROLD J. ROAST, Canadian Bronze Co., Montreal, P.Q., Canada.

NOTE: The paper, in the absence of the authors, was presented in abstract form by T. C. Watts, Falcon Bronze Co., Youngstown, Ohio.

J. J. CURRAN[1] *(Written discussion):* This contribution by Messrs. Fleck and Bunch is a praiseworthy effort to add to the sum total of human knowledge. As such it is to be commended. At the same time, the work they have done and the conclusions they have reached, call for some comment.

Composition "B," on page 591 of the preprint, is not "Gun Metal" within the scope of the current Navy Specification.

The Navy Specification for this material requires, among other things, that "The castings shall be......free from......porosity...... etc.......", and that "Castings shall not be repaired......"

The present paper is concerned with the procedure used by a Navy Yard to repair porous valve castings, presumably for Navy use. This is reminiscent of an instance, years ago, when the metallurgist of another Navy Yard extolled the virtues of manganese bronze made in that Yard from scrap metals, in spite of the fact that the Navy Specification of that day forbade the use of scrap.

[1] Research Metallurgist, Walworth Co., Greensburg, Penna.

The annealing method of closing porosity in pressure parts of Gun Metal has been reported in the past as a means of overcoming porosity. Additions of lead in an amount somewhat in excess of Navy Specification limits have also been found to reduce difficulty from leakage, and the preponderance of evidence would indicate that such an addition is desirable from other standpoints.

The writer feels, however, that pressure-tight castings can be made of material conforming to the Navy "Gun Metal" specification without repairs, providing proper foundry practice is followed. The fact that slight deviations from specified composition would minimize difficulties in production is pertinent only insofar as it may indicate need for revision of the specification.

In their paper, the authors do not mention any details of the foundry practice which produced the leaky castings. They do state, however, that the skin of a casting is pressure-tight, and that precautions must be taken to preserve this skin if the casting, as a whole, is to be pressure-tight.

They also indicate their familiarity with "gas" troubles, which, incidentally, are probably the cause of the "microscopic porosity" which they cure by repairs. They apparently do not realize, however, that a casting which does not "purge" may still be sufficiently porous to leak under pressure when the skin has been machined off. A substantially gas-free casting does not recognize the need for a pressure-tight skin.

Consistent production of gas-free bronze is attended with difficulties and requires constant vigilance and good foundry practice. Such production is not uncommon in the valve and fitting industry—without, we believe, resort to repair by annealing.

T. C. Watts[2] *(Written discussion):* The report of this experiment in heat-treating castings, designed for pressure work, has its place in the literature of the A.F.A. and the authors should be commended for their work and efforts in preparing it.

The paper brings to mind the great difference between a jobbing foundry and a production foundry. When a plant goes through the same procedure day after day, using the same patterns and equipment, there should be no necessity to go to the expense of heat treating a casting to withstand ordinary pressures. A jobbing foundry, to be successful, must survey a pattern from all angles in order to make an acceptable casting on the first try. Experience is the most valuable of all factors which govern the melting, alloying, gating and pouring of a casting. Needless to say, if an order of one or two castings is made they have to withstand pressure and if leakers are encountered, the ability to heat treat such castings is of importance to the jobber.

Emphasis of the author's conclusion 5 is advisable. (Read Section 5.)

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[2] Metallurgist, Falcon Bronze Co., Youngstown, Ohio.

I believe with a proper melting condition and the intelligent use of test bars made before the heat, then pulled and broken for visual examination, no gassy or oxidized metal will result. High pressure bronze can be made with the alloys mentioned in the paper without heat treating. However, if occasional tests are made where porosity, even though microscopic, is evident, then it would be cheaper to heat treat the castings than to make them over.

A. J. SMITH[3], S. A. WEIGAND[3], AND J. W. BOLTON[3] *(Written discussion):* Any safe method of eliminating porosity in bronze castings is of decided interest to the manufacturer and user of these products. Many bronze castings are used at elevated temperatures and under considerable internal pressures. Safety in service must be considered. Possibility of explosion, scalding workmen with steam, freeing of inflammable vapors or liquids, or leakage of corrosive fluids always must be considered in the selection of metals, wrought or cast, for any severe engineering application or applications.

The usual approach of the foundryman is to so regulate his melting and other foundry practices that material of best density—hence greatest degree of freedom from porosity—internal or otherwise—will be attained. To that end a number of firms have invested in research and have made real and sound progress.

Messrs. Fleck & Bunch propose the salvaging of porous castings—castings that are unsound, leakers, whose metal structure initially is weak and not tough and dense. If internal welding without the setting up of embrittlement was or could be shown, a constructive accomplishment would ensue, although even then the writers prefer getting sound metal in the first place. It is true that Messrs. Fleck and Bunch emphasize the danger of indiscriminate attempts to apply their methods, rightly recognizing that improper application of any method may react to the consumers detriment.

Briefly, Messrs. Fleck and Bunch propose a heat treatment at 1300°-1400°F. for closing up leaky 88-10-2 castings. Aside from the fact that they closed up the voids to fluid penetration, the evidence that they give as to the actual effect on the strength of the metal, the structural changes, and the behavior of such treated castings under severe engineering service is at best vague. No evidence has been presented that the test bars on which physical property tests were run were porous as the photomicrographs given are all of valve body material and from a pattern without doubt considerably more difficult to cast soundly.

The results of the physical property tests are a little difficult to interpret. Annealing at the recommended temperatures would bring about homogenization of the material with production of a more uniform

[3] Lunkenheimer Co., Cincinnati, Ohio.

alpha phase. Such homogenization would lead one to expect a certain softening and increased ductility. The softening takes place as expected yet the elongation also shows a drop with increasing temperatures of anneal. Evidently some reaction besides homogenization comes into play. In some work we have performed on valve bronze conforming to A.S.T.M.-B-61 specification or "Navy M" metal no improvement was found in physical properties by annealing at any temperature up to 1600°F. At 1600°F. the metal was burned and the physical properties were very poor. Quenching from above 1000°F. completely ruined the more useful properties and a slow air cool or even furnace cool was necessary to retain them.

The mechanism involved in the closing of leakers by the method of Messrs. Fleck and Bunch is obscure. Any leaky casting submitted to a hydrostatic pressure test would take up and hold a good deal of water and unless indeed extremely porous on annealing such a casting the oxidation product would fill the pores sufficiently to make the casting tight. A steel casting that is a slight leaker will usually shut as a consequence of such a test. It is unfortunate that the photomicrographs are not of sufficient clarity or magnification to tell what is happening precisely in the porous regions. It would appear from the photomicrographs that approximately 5 per cent of the as cast volume is delta constituent. The homogenization of such an alloy would bring about a volume expansion of some 0.2 per cent as calculated from data in the International Critical Table for a 10 per cent Sn Cu alloy. Such an expansion should effectively shut a leaker where this amount of delta is present. Whether a mechanically strong joint would result is questionable. In bronzes such as the A.S.T.M.-B-61 referred to above no delta should be found in the as cast material and a closure from expansion could not take place.

A question should be directed to Fig. 2. Photomicrograph number 8 A. This photomicrograph apparently shows signs of mechanical working of the sample as evidenced by the type of grain configuration and the annealing twins at the right hand side of the picture. Needless to say, nothing can be remarked of the closing of voids through plastic deformation.

It has been shown by one of the writers (Proc. A.S.T.M. — 1935, Part 2) that exposure of the 88-10-2 alloy for long periods to temperatures under 600°F. results in definite and serious embrittlement. Unless fully understood, and carefully tested out, some suspicion might be in order as to the effects of annealing treatments at 1300°-1400°F. on the usefulness of the metal thereafter for severe engineering service.

The method proposed by Messrs. Fleck and Bunch is not new. Geo. K. Elliott (Proc. A.S.T.M. — 1917, Part 2, pp. 234-235) speaking of leaky castings of difficult design said "The final of many remedies tried was that of annealing for a short time at 1400°F. Upon retesting the annealed castings it was found that more than 75 per cent of those which previously had leaked were now tight under the hydrostatic test.

Microscopic examination of several of the annealed castings revealed that they had a very homogeneous structure—the effect of the annealing was to dissolve the eutectic, leaving a uniformly unbroken homogeneous structure." He then goes on to suggest that the shrinkage of the eutectic accounted for the leakage and that bronze porous from other causes could not be corrected by any method of heat treatment. Elliott further stated that the annealing "perceptibly increased the elongation of the metal while the ultimate strength remained unchanged." Possibility of porosity due to shrinkage of an eutectic also has been mentioned by Cowan (Jour. Inst. of Metals (1927), Vol. 38, p. 4).

The late Geo. K. Elliott, a careful and thorough workman, undoubtedly followed up the heat treating method further, and it is very significant that this practice was never adopted commercially by his firm. Records are not available now, but doubtless the eutectic contraction, which might be safely cured by heat treatment, and intercrystalline porosity, incipient shrinkage, were and are not distinguishable without destructive testing. Experience teaches that most bronze porosity is in fact general incipient shrinkage which is not closed up by fusion, but only by "stuffing" the interstices with oxides or other substance. Any practice which does not produce actual internal welding is of doubtful value.

Without meaning to imply that there is any questioning of the results of the author's work, that is, pressure tightness was obtained, it seems that we may well consider the philosophy of this matter. Is this good foundry practice? Does it advance the foundry art itself? It is conceivable that there are cases where it is permissible to close up castings, but certainly such a permission should not be extended to castings which will in the normal course of events be subjected to severe engineering service. And it would be most highly desirable to know just what the strength and ductility of the metal in the actually porous sections would be both before and after treatment.

Concluding, the writers believe that the sound approach is to get good castings, and avoid dangers inherent in attempts to salvage those with internal porosity or incipient shrinkage. The best place for "leakers" is the scrap heap.

W. M. BALL:[4] Many castings have to be furnished machined before testing. Should these machined castings be found porous and then subjected to a heat treatment much discoloration and distortion would take place that would be highly undesirable and possibly require some remachining.

W. B. GEORGE:[5] These porous castings have a more or less true tensile drop of from 40 down to 10 or 15 thousand. I believe all will agree on that. And, I cannot see where the heat treating will bring the tensile strength of the gaseous or porous section back.

[4] General Supt., Edna Brass Mfg. Co., Cincinnati, O.
[5] Metallurgist, R. Lavin & Sons, Inc., Chicago, Ill.

MR. BOLTON: I might add here that if there was evidence shown there was intercrystalline welding in the casting, a very useful purpose has been accomplished. No such mechanism has been shown.

H. FLECK and T. C. BUNCH *(Authors' written closure)*: The authors especially wish to thank Mr. Watts for his so able presentation of our paper, and for his written discussion. It is clearly evident from his discussion that he accepted the paper in precisely the manner in which it was intended. Further, we are very grateful to those who submitted written comments.

We have read Mr. Curran's comments with considerable interest and are in complete accord with him on several points. We have found that the addition of lead up to approximately one per cent materially reduces difficulty from leakage. It is also our experience that pressure-tight castings can be made, without repair, of materials conforming to Navy specifications. In fact, over 95 per cent of our castings are so made. Heat treating is resorted to only on a very small percentage of castings where urgency and limited quantity forbid experimentation. At this point we would call Mr. Curran's attention to the great difference, in type of work and its associated problems, that exists between a production foundry and a jobbing foundry.

We believe that our foundry molding and melting practices are correct in every detail. We purposely covered this phase of the paper very briefly in order that we might devote more space to microscopic porosity and the heat treating process. The authors have long realized that molten metal may be substantially gas-free or that it may contain gases in sufficient quantity to cause "purging." We do not believe, in our case, that gassy metal is the direct cause of microscopic porosity—although we can readily see where a small quantity of gas would definitely have this effect.

We are in serious disagreement with Mr. Curran's statement, that "a substantially gas-free casting does not recognize the need for a pressure tight skin." We believe that his line of thought follows the assumption that small quantities of gas are responsible for the porosity. Our experience, however, is that the need for a pressure tight skin is due entirely to freezing conditions that exist while a casting is cooling in the mold. It is well known that zinc-bronze assumes a fine crystalline structure when chilled. The surfaces of a casting in contact with the mold are fine-grained. Castings that have section thicknesses up to one-half inch may show this fine grained structure throughout. However, on heavy castings the grain increases in size towards the center and it is possible by machining excessively on both inner and outer surfaces, as on liners and sleeves, to expose this large grain material. In most cases, weeping or sweating will result. We realize that the proper procedure in this case is to cut down machine allowances and we have followed this plan with successful results.

Messrs. Smith, Wiegand and Bolton have shown considerable interest in our paper and we wish to express our appreciation for their

constructive criticism. We would, however, like to reply to some of
their comments.

First, the castings that we heat treat are represented by test bars
that meet Navy strength specifications. These test bars are either at-
tached to the castings or poured separately from the same melt. If we
accept the test bar results as representing the strength of the castings
—then we must necessarily admit that the castings are sufficiently
strong. When we encounter a casting that weeps or sweats upon the
application of pressure, we examine the casting macroscopically and if
the metal appears sound, we believe that it is good practice to resort
to heat treament in order to close this microscopic porosity. Castings
that we have so heat treated, have been placed in service under very
severe engineering conditions, and in two years no failure of this
material has been reported. In view of these facts, we feel it both good
engineering as well as economical practice to anneal castings that fall
in the scope covered by this paper.

Secondly, Messrs Smith, Wiegand and Bolton point out that the
elongation of the heat treated test bars, of one alloy, as listed in Table 1,
does not conform with the result that would generally be expected from
a heat-treating operation that produces a softer and more homogeneous
material. As a matter of fact, the elongation shows an average drop
of approximately four per cent. The lowest percentage elongation shown
by a heat treated test bar was 32.5 per cent. Navy specifications for
gun metal require a minimum of 20 per cent elongation. It appears that
the lowest elongation encountered is still considerably above the minimum
required. Since all strength calculations must necessarily be based on
minimum specifications, we are of the opinion that it is better to sacrifice
a small percentage of the elongation factor, and salvage the casting by
heat treating, than to relegate the casting to the scrap heap. It will be
noted that alloy No. 2, Table 1, shows very little variance in elongation
in the "as cast" and heat treated test bars, whereas the average tensile
strength and yield of both alloys has been increased.

In the interest of clarity of metallurgical terms, we suggest that
the word "eutectic" as applied to the delta constituent be changed to
eutectoid.

In reply to Mr. Ball, we wish to state that in an ordinary annealing
furnace the presence of oxygen cannot be avoided. This naturally causes
oxidation and discoloration. We believe, however, that with a furnace
of the "controlled atmosphere type" it would be possible to bright anneal
castings. Distortion of castings during heat treatment within the tem-
perature range 1300° to 1400°F., is very slight. We had occasion to
measure this on a sleeve casting of the approximate dimensions 12 in.
inside diameter, 13 in. outside diameter and 6 in. long that had been
machined to finished dimensions. The greatest distortion after heat
treating and sand blasting was 0.005 in. This difference in measure-
ment we attributed to scale formation rather than to distortion. We
wish to point out that the majority of castings that we have heat treated

were so treated in the rough machined state and it follows that any distortion would be removed by further machining.

Mr. George has brought up an interesting point regarding the poor tensile strengths exhibited by porous castings. We agree heartily with his figures, with the stipulation that, the drop in tensile strength is as a general rule proportional to the degree of porosity. Porosity as delt with in this paper is that which is discovered only by microscopic examination or that which becomes evident when a casting weeps or sweats slightly under pressure. May we point out that test bars poured along with the castings, and out of the same melt met and in most cases exceeded all strength specifications. Microscopic porosity in our experience is generally localized, and in most cases leakage develops on surfaces that have suffered from excessive machining. This type of porosity we believe is not due to gassy metal. Castings poured from gassy metal will leak profusely throughout and it follows that all castings poured from a heat of gassy metal would leak. We have experienced no such difficulty at Pearl Harbor. Metal that appears porous on macroscopic examination generally has poor physical characteristics. Test bars will not meet specifications and the castings should be condemned for obvious reasons.

Steel Casting Design for the Engineer and Foundryman†

By C. W. Briggs*, R. A. Gezelius***, and A. R. Donaldson**, Washington, D. C.

Abstract

Steel casting design has been considered from two viewpoints, namely, that of the designing engineer and that of the foundryman. A code or the nucleus of a set of rules is presented for the engineers' guidance and experimental results on padding, external and internal chills, and controlled directional solidification are reported as of possible interest to foundrymen.

Part I—The Designing Engineer and Steel Casting Design

Introduction

1. It is the plan, in this section, to introduce to the designing engineer the concepts of steel casting design in a logical manner, simplified as much as possible.

2. It is felt that a discussion of design features at this time is appropriate, as the requirements for new equipment operating under higher pressures and at higher temperatures are calling for new designs. Also, requests for reductions in weight have introduced casting problems which require more tolerant design. These new demands have resulted in reductions in wall thicknesses which have, in some cases, produced wall sections that are but small fractions of those formerly employed.

3. The reduction of wall thicknesses, to effect weight savings, is causing considerable thought and drawing much attention (among consumers and foundrymen) to the internal condition of castings. Defects that were once hidden by very heavy sections are

now brought to light. Also, weight savings have reduced the factor of safety and more concern is being given to the integrity of the castings. Castings that operate under high pressures and temperatures are now required to be absolutely homogeneous, and nondestructive testing methods, especially X-ray and gamma ray inspections, are being frequently used to check on the internal conditions. Thus many things have combined to make the producer and the consumer more conscious of the internal integrity of the casting.

4. In general, it may be said that most of the defective castings are due to poor design, but as there are no assembled data on design, one can hardly blame the designing engineers for these undesirable conditions.

DIFFICULTIES DUE TO POOR DESIGN

5. It might be most appropriate to point out to the designer those defects that occur in steel castings because of poor design. Both hot tears and shrinkage cavities are prevalent in poorly designed castings. Hot tears form due to large temperature differences in castings where excessive internal stresses cause members to separate or crack at temperatures slightly below the steels' solidification temperature. Cracks occur at abrupt changes in section and at sharp angles.

6. Shrinkage cavities are due to insufficient metal to care for metal contraction at the time of casting solidification. They are found in sections that must be fed through smaller sections.

7. Even though the above discussion may mean little or nothing to the designer, it is sufficient to point out, regardless of the terms employed, that internal stresses and contraction phenomena can not be disregarded in the planning of a design from which may be produced perfect castings. It also should be added that the mechanical properties of the casting will vary with size of section. This point will be discussed later at greater length.

DESIGNING TO PREVENT HOT TEARS

8. There are two ways in which the designer may prevent the possibility of his embryo casting from cracking or being subject to hot tear formation. These are:

(1) Eliminating hot spots that are under stresses.

(2)   Eliminating the stresses that would act on the hot
spots or other points of stress centralization.

9.   In principle, the above points could be stated more easily
by merely emphasizing that all stress should be eliminated, since
the high stresses are primarily the cause of hot tear formation.
Unfortunately, casting stresses cannot be eliminated, but under
certain circumstances they can be controlled. It is not necessary
that the details of stress formation and action be presented here,
as the mechanics have been fully described in other publications
by the authors[1, 2, 3*]. It may be pointed out, however, that because
of the temperature-contraction relationship, temperature changes
are causes of stress formation. Thus, if hot spots are eliminated
the excessive temperature changes are eliminated, stresses are thus
reduced, and hot tear formation prevented.

10.   Under certain conditions the design cannot be changed
to remove the hot spot. In this case, the stresses must be relieved
in some other manner. That can best be discussed by reverting to
illustrations. But first, how is a designer to know what hot spots
are and where to look for them?

Formation of Hot Spots

11.   A member of a steel casting solidifies from the sand faces
toward the center of the mold cavity. The rate of solidification for
practical application is about the same whether the section be one
inch or four inches thick. Thus, if the rate of skin formation is
about the same, it naturally follows that it will take the heavier
section longer to completely solidify. If it is considered that the
thinner section is integrally connected to the heavier section in the
casting, it may be pointed out that the thinner section is completely
solidified and is cooling, and therefore contracting, while the heavier
section is still solidifying. The temperature gradients so estab-
lished will result in different rates of contraction and hence the
formation of stress. Cracking will occur at the weakest section,
which, of course, will be the hottest section.

12.   Thus, reverting to the question "what are hot spots and
where are they found?" it may be said that hot spots refer to
sections of extra mass and they are found at positions of joining
sections. One or two examples may be sufficient to crystallize the
statement. In a cast wheel, the solid hub, which is much heavier in

* Superior numbers refer to bibliography item numbers at end of this paper.

section than the arms or the rim, is a section of extra mass and is located at a position of joining sections. Thus the hub may unquestionably be spoken of as a hot spot. In the typically designed valve, the flanges or the seat are sections of heavy mass joined to much thinner wall sections and therefore may be termed hot spots.

13. The above discussion leads to the statement of the first rule of the code:

> **Rule:** *An attempt should be made to design all sections in a casting with a uniform thickness.*

This rule, if followed, reduces the presence of hot spots in a casting to a minimum. These hot spots are responsible for the setting up of temperature gradients, contraction variations, stress formations and their resultant — the deplorable occurrence of hot tears.

Elimination of Casting Stresses

14. It has been suggested previously that the designer will find that under certain conditions the design can not be changed to remove the hot spot. Such a condition necessitates the application of the second plan for the prevention of hot tears by design; namely, eliminating the stresses that would act on the hot spots or other points of stress centralization.

15. In the first place, what may be considered as points of stress centralization? Stresses acting upon a casting usually con-
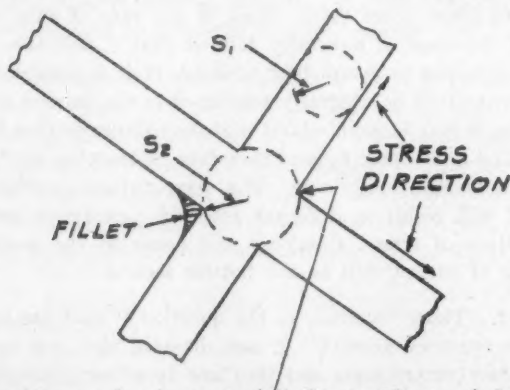
Concentration of stress and formation of hot tear.

FIG. 1—CONCENTRATION OF STRESS.

centrate at abrupt changes in section or at sharp corners. These positions are structurally weak and, coupled with the fact that they are usually hot spots with low mechanical properties, it is easily seen that they are potential positions of hot tear formation.

16. For example, two sections, which are an integral part of a casting, are joined together in the shape of an "X" (Fig. 1). By inscribed circles, a method proposed by Heuvers[7], the effective mass can be ascertained. If, however, liberal fillets replace the sharp corner junctions the stresses will be more evenly distributed and hot tear formation will be less prone to take place.

17. In the discussion of fillets for the prevention of hot tears, it should be pointed out that the elimination of hot tears must not be the sole consideration since an increase in the radius of the fillet also increases the size of the possible shrinkage cavity. Therefore, judgment as to what constitutes an adequate fillet should be reserved until the section on internal unsoundness due to shrinkage has been thoroughly studied.

18. Abrupt changes in section may be considered analogous to the case described above in that sharp corners, temperature gradients, and stress concentration are similar. Fig. 2 illustrates how an abrupt change of section can be altered from a poor design to designs that will reduce the possibilities of hot tear formation to a minimum.

19. The above explanatory notes lead to the statement of further rules to the code:

Rule: *It is not desirable to design cast steel structures with abrupt changes in section.*

Rule: *Sharp corners at adjoining sections are sources of defects, and, if possible, should be eliminated.*

Elimination of Major Stresses

20. The points of stress centralization have been discussed as to their nature and disposition. There is, however, still remaining for discussion that phase pertaining to the elimination of major stresses, which, acting on hot spots, results in the formation of hot tears.

21. It has already been pointed out that if sections throughout the casting are uniform, major stresses will not arise from

excessive temperature gradients and, of course, prominent hot spots will not be present within the casting. Under certain conditions, it is impossible to design a casting so that prominent hot spots are eliminated, and, since excessive temperature gradients will thus be present, it becomes necessary to relieve the major stresses resulting therefrom by some other means. The foundryman has at his command a method to reduce these stresses by employing mold relieving principles. The designer also can aid toward producing a more stress-free situation by designing an intricate casting in two, three, or more parts, which are finally assembled into one complete casting by joining the various cast parts by welding or bolting.

22. Castings of the weld-assembly design are becoming increasingly popular with design engineers and casting purchasers and are being looked upon quite favorably in many foundries.

Weld Assemblies

23. The principles involved in the weld-assembly design are discernible readily to the designing engineer, but for the sake of clarity a simple example will be illustrated. Fig. 3 is a cross sec-

(1)—Poor Design.

(2)—Not Recommended.

(3)—Fair.

(4)—Good.

(5)—Best, and in Some Cases Better than 6.

(6a)—No Change of Section. Recommended Design.

(6b)—No Change of Section. Recommended Design.
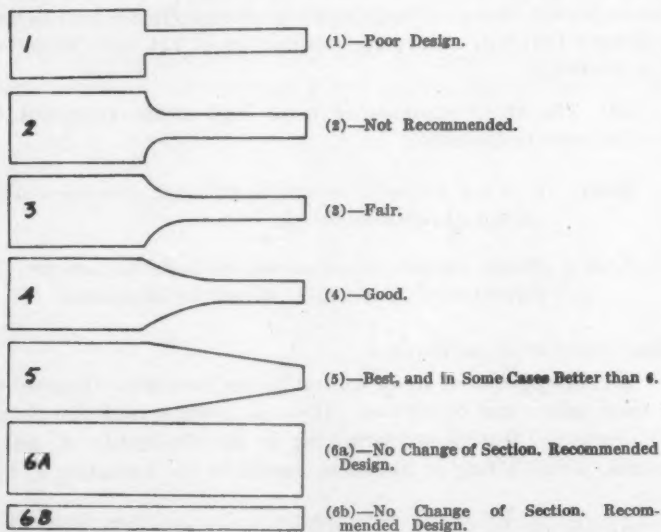
FIG. 2—ALTERATION OF DESIGN TO REDUCE POSSIBILITIES OF HOT TEAR FORMATION TO A MINIMUM

tion involving two features of poor design; namely, the hot spot and a continuous wall section which would be responsible for stresses mainly in the direction of the arrows, due to hindered contraction. A casting so produced would very likely show hot tears in the central section, if the foundryman did not take precautionary steps. If, however, the designer eliminated the cross end pieces, A and B, the castings would not be under the major stresses as shown and
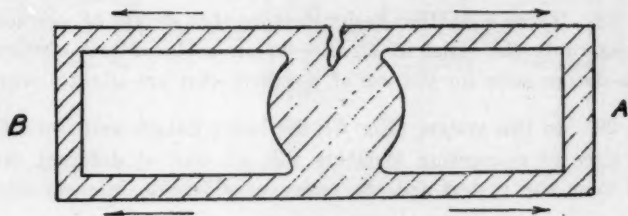
FIG. 3—HINDERED CONTRACTION STRESSES.

failure would not be a possibility. The casting could be assembled in the final form by welding previously cast or fabricated sections, A and B, into position.

24. An actual commercial application of these principles is illustrated in the manufacture of a throttle valve casting. Fig. 4 portrays a sketch of the first design of this casting. Considerable trouble was experienced in production and a number of attempts were made before the casting was finally produced without major defects. In Fig. 5, the design was altered to the cast-weld-assembly job. All extraneous parts were removed from the base casting, cast separately, and then these parts were welded together to complete the assembly. This casting was produced without defects on the first attempt. It is needless to add that the saving in time and cost of production was considerable.

25. This principle was extended to the turbine casting as illustrated in Figs. 32 to 35 inclusive.

26. It is felt that designers should not be alarmed by this type of construction for excellent properties and good homogeneity can be obtained, as has been demonstrated by White, Clark, and Crocker[5].

27.   The above discussion leads to the statement of another rule:

Rule: *When the design of a cast steel structure becomes very complicated or intricate, it is suggested that it be broken up into parts so that they may be cast separately and then assembled by welding or bolting.*

DESIGNING WITH A SINE WAVE CONSTRUCTION

28.   There is another design feature that should be mentioned in regard to the relief of stresses in an enclosed cast structure. This design calls for the use of members that are slightly waved.

29.   In this system (Fig. 6), the heavy flanges and two different sizes of connecting members will all cool at different rates, and thus the system will be under considerable internal stress.
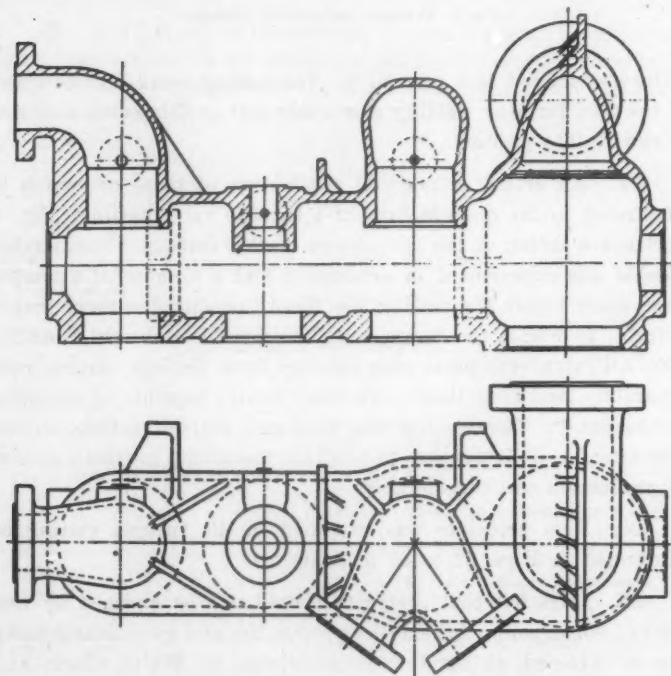
FIG. 4—THROTTLE VALVE CASTING (ORIGINAL DESIGN).
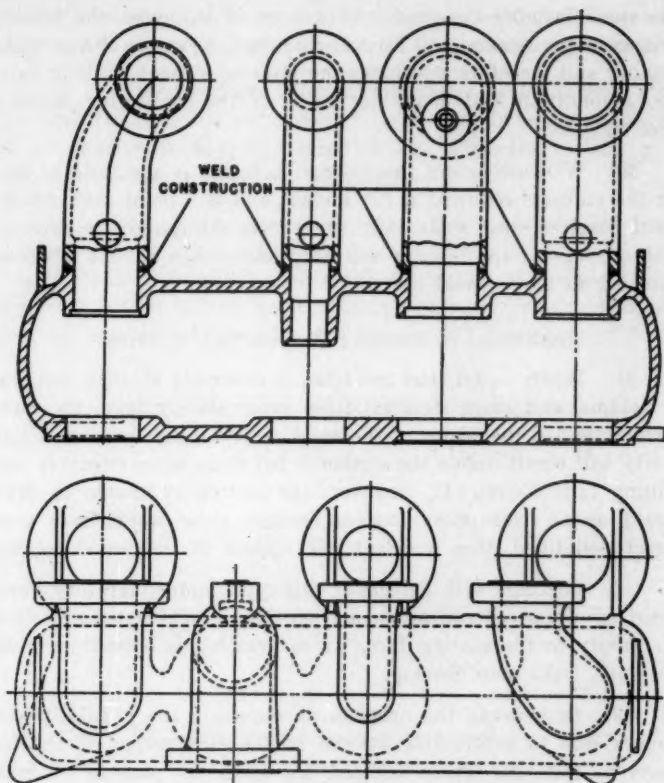
FIG. 5—THROTTLE VALVE CASTING (CAST WELD CONSTRUCTION).
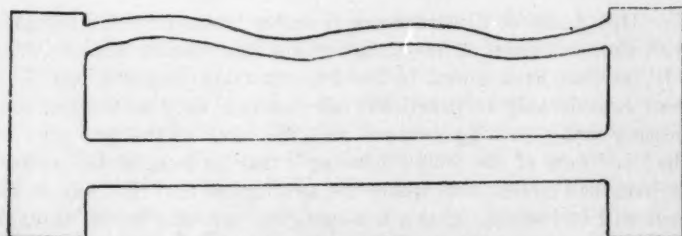
FIG. 6—SINE WAVE CONSTRUCTION.

The small member designed with a wave in it, under the tension stresses of the system, will be pulled toward the shape of a straight member and therefore dissipates the internal stresses without tearing, a condition that could not result if the connecting member were straight.

30. Whether or not this type of a design is adequate to care for the stiffness required is not known, and is a point that designers themselves must settle. However, from the foundry viewpoint, such a design is appropriate and very useful as a method of stress relieving in an enclosed system.

DESIGNING TO PREVENT CONTRACTION CAVITIES

31. When liquid steel solidifies, it contracts about 3 per cent in volume, and since steel solidifies progressively from the mold surface toward the center of the mold cavity, a pipe or contraction cavity will result unless the section is fed from some reservoir containing liquid steel. If, however, the section is unable to draw liquid metal from other sources because these inlets have completely solidified, then a defect will appear in the unfed section.

32. Castings with defects of this type, under vibratory stress aided by stress centralization, may develop cracks extending from the cavity to the casting face. If the casting is operating under pressure, leaks may develop.

33. In general, the problem of removing the possibilities of defects due to contraction results in the elimination of the hot spots in castings. These isolated hot spots are sources of great trouble in the manufacturing of steel castings, and anything that the designer could do toward eliminating them would be greatly appreciated by the foundryman.

Designing "L" Sections

34. A simple illustration will suffice to acquaint the designer with the seriousness of the situation and the possible remedy. The "L" section, as is shown in the accompanying diagram, Fig. 7, is used considerably in practically all castings, such as flanges, connecting and adjoining sections, and the like. If the two arms of the "L" were of the same thickness, it may be seen, by the method of inscribed circles, that where the two legs of the "L" join, a hot spot will be located. If this hot spot (S_2) can only be fed through the two legs of the "L" section, one may be certain that when the

casting has solidified, since the leg sections solidify first, there will
be a cavity somewhere near the center of the inscribed circle, S_2.

35. If, however, the "L" section were in a location where
the junction could be fed directly by an outside reservoir, instead
of attempting to feed through the legs, then the section at S_2
would be sound and contain no cavity.

36. There are numerous occasions unfortunately when such
a joining section can not be fed by an outside reservoir of metal
due to its inaccessibility, such as at the bottom of the mold or an
internal section of the mold. There is no special problem with
joining sections or hot spots that can be fed directly, and design
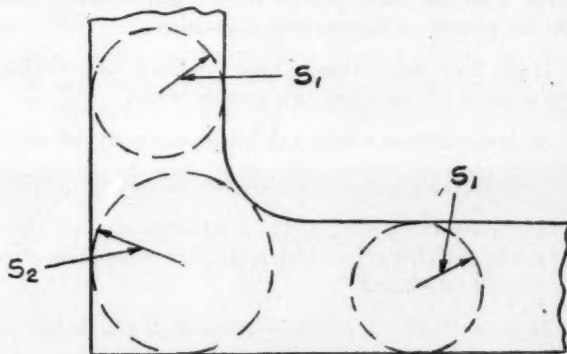
Fig. 7—Hot Spots.

features are not very important as far as safeguards against cav-
ities are concerned; but sections that can not be fed should be con-
sidered carefully by the designing engineer.

37. One of the most difficult features about designing to
prevent contraction cavities is to know when and where feed heads
may be placed since their position is quite largely a matter of
opinion among foundrymen. As a matter of fact, the casting may
be molded contrary to any plans that the designer may have had.
Thus, the casting may better be designed with the supposition in
mind that none of the metal junctions or hot spots can be fed
from outside reservoirs. With this point in view, experimentation
was carried on with joining sections to determine the extent of
the contraction cavity with various changes in design.

Studies on Various Joined Sections

38. There are only five ways in which sections may be joined. These may be exemplified by the following letters of the alphabet: L, T, V, X, and Y. All other modes of connection are merely modifications of the above possibilities. Designs in the forms of these letters have been studied in detail.

39. Patterns were 3 x 3-in. in cross-section. The arms were about 2 ft. long and headed at the extreme end. They were poured through the feed heads on the arms. In order that no variations should be overlooked during pouring, studies were made by pouring through one arm, and through two arms at the same time. Molding was according to the dry sand practice and plain carbon basic electric steel was used, poured from a teapot ladle. The sections were all poured in the horizontal position.

40. It has been stated that in these studies a 3-in. section was used. The reasons for choosing this section were:

(1) A large section would exhibit pronounced defects.

(2) A 3-in. section is not uncommon in structural castings.

(3) With the large section, the conditions of sand and steel would not be so critical and more comparable results could be obtained.

41. It is felt that the points illustrated in the design charts will be of a comparable order in sections other than those of 3-in., and that information obtained on a 3-in. section can be used in the design of other adjoining sections.

42. The sections were modified by using radii of different sizes in the corners and the effect of the changes was noted. For the purpose of this work, it has been assumed that the section studied is located in the casting in such a manner that no feeding is possible except through the arms.

43. The depth of the section has, of course, no bearing upon the shape of the defect in the plan view as long as the depth is great enough to permit a defect to form. Even if the depth of the section were infinite, the plan view of the defect should, theoretically, remain the same.

44. After the sections were cast, they were radiographed. Castings were radiographed with radium at a distance of 40-in.

with a 2-mm. source or focal spot in order that the defects would appear in their natural sizes on the film. The defects exhibited in the radiograph were then reduced photographically so that the defect drawn on the small diagrams presented in this report would be shown in their correct proportion. Reproductions of two radiographs are included to illustrate the general characteristics of the defects found (Figs. 8 and 9). Fig. 8 is a radiograph of a 3 x 3-in. "T" section and Fig. 9 is a radiograph of a section taken from a commercial casting.

45. It is now suggested that the designer turn to the charts illustrating the various sections. The charts are self-explanatory and from them may be obtained several features of design. See Figs. 10 to 26, inclusive.

46. In discussion of the "L" section, it may be said that if the outside corner at the junction is maintained, an increasing radius at the inner corner will bring about an increasing size of defect. If the section is uniform throughout, a defect will be found

FIG. 8—RADIOGRAPH OF T-SECTION.

if the inner radius is small. If the inner radius is increased, but a uniform section is maintained, the defect will become smaller until it develops into center line weakness, a condition apt to be found in uniform sections of any length.

47. Sections of the "L" type should not have less than ½-in. inside radius, and only in a few cases would a 3-in. inside radius be used. The best common practice, therefore, appears to be one in which radii of ½ to 1-in. would be used.

48. The practice of designing the section at the junction slightly smaller than that of the arms appears to give the best results.

"T" Sections

49. The only way in which a "T"-section can be designed, without being responsible for a contraction cavity, is to core a hole at the center of the junction of the two members. Depressions in the arm of the "T", while not eliminating the defect, do reduce it markedly from that found in the common "T" design.
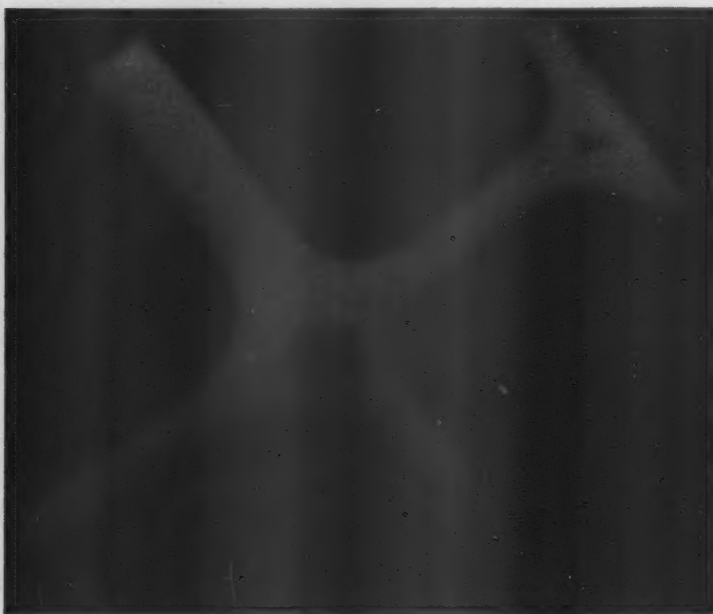
FIG. 9—RADIOGRAPH OF COMBINED X AND T-SECTIONS.

(104)—Inner corner—½ in. radius.
Outer corner—4½ in. radius.
Section at the junction is slightly smaller
than the arms.
No defect.
Best design.
No Defect.

(105)—Inner corner—½ in. radius.
Outer corner—3½ in. radius.
Very much smaller junction than arm
sections.
Drastic design not needed for complete
soundness.
No Defect.

(106)—Inner corner—3 in. radius.
Outer corner—6 in. radius.
Uniform section with large radius de-
velopment of center line weakness which
can be remedied by foundryman.
Good design.

Fig. 11—L Sections—104, 105, 106.

(101)—Square corners—no radii.
Section 3 in. x 3 in.
Poor design—also possibility of hot tears
at sharp corner junction.
Area or Defect—2.05 sq. in.

(102)—Inner corner—½ in. radius.
Radius slightly increases the size of defect.
Area or Defect—2.8 sq. in.

(103)—Inner corner—½ in. radius.
Outer corner—3½ in. radius.
Uniform section shows a small defect.
Design better than 102.
Area or Defect—0.6 sq. in.

Fig. 10—L Sections—101, 102, 103.

50.   Since the size of the defect increases with the size of the radius, it is recommended that a radius of ½-in., and not over 1-in., be used.
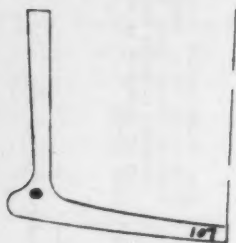
(107)—Section of a 14 in. Bitt casting. Defective—a poor design. To remedy—Maintain uniform section or core out hole at position of defect.
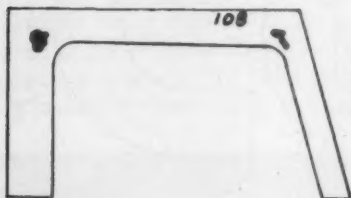
(108)—Double "L" from stern post casting. Defective—poor design. To remedy—Eliminate corners.

(109)—Double "L" from stern post casting. Poor design. Study also sections X and Y. To remedy—Make sections uniform and core out hot spot center.

FIG. 12—L SECTIONS—107, 108, 109.

51.   It has been suggested from time to time by those writing on the subject of design, that if the leg of the "T"-section were made smaller in section than that of the arm, the hot spot, and hence the size of the cavity, would be reduced. This is true, as is illustrated in 202, 209, 210 and 211 (Figs. 13, 15 and 16). It also should be pointed out, however, that as the leg section decreases, the temperature differential between the two sections increases, and stresses are formed that may result in the formation of hot tears. Thus, care should be used in designs of this type so as not to incorporate them in enclosed stress active systems.

(204)—Inner corners—2 in. radii.
Area of Defect—4.1 sq. in.

(205)—Inner corners—1½ in. radii.
Outside—5-in. radius—V-shaped.
Design much better than 204.
Area of Defect—0.2 sq. in.

(206)—Inner corners—1½ in. radii.
Outside—2 in. radius—U-shaped.
Defect larger than 205.
Area of Defect—0.8 sq. in.
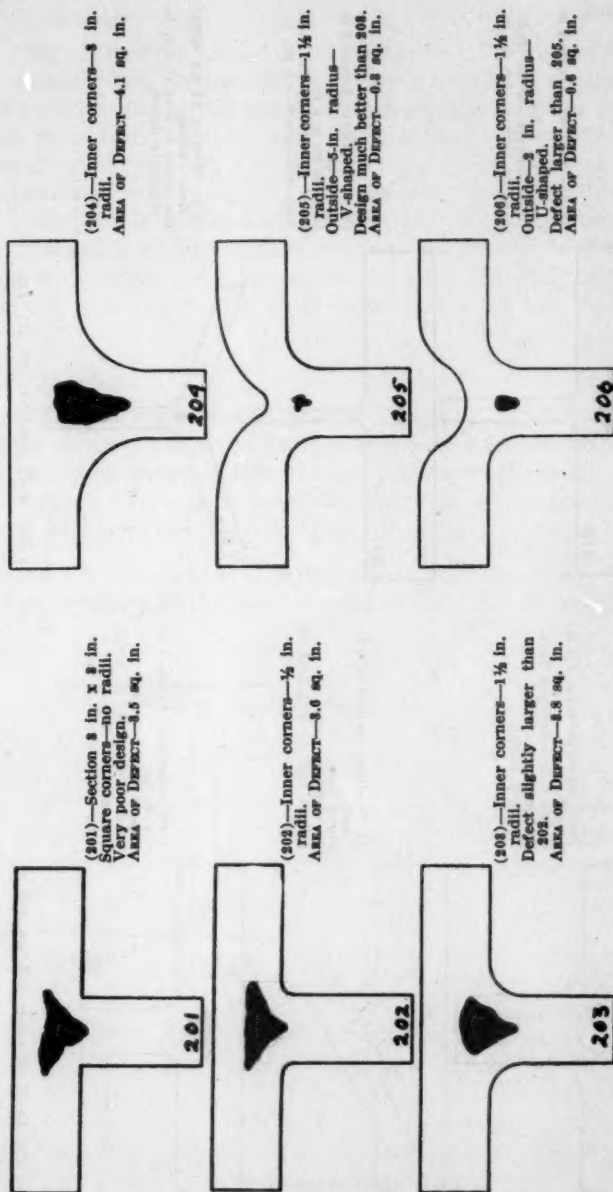
Fig. 14—T Sections—204, 205, 206.

(201)—Section 2 in. x 2 in.
Square corners—no radii.
Very poor design.
Area of Defect—3.5 sq. in.

(202)—Inner corners—½ in. radii.
Area of Defect—3.0 sq. in.

(203)—Inner corners—1½ in. radii.
Defect slightly larger than 202.
Area of Defect—4.2 sq. in.

Fig. 13—T Sections—201, 202, 203.

(207)—Inner corners—1⅛ in. radii. Cored 2 in. diameter hole. No defect. The best design.

(208)—Inner corners—1⅛ in. radii. Cored hole. No defect. Design not as practical as 207.

(209)—Sections: Arms—2 in. x 2 in. Leg—2 in. x 2 in. Inner corners—⅛ in. radii. Compare with 202. AREA OF DEFECT—1.8 sq. in.
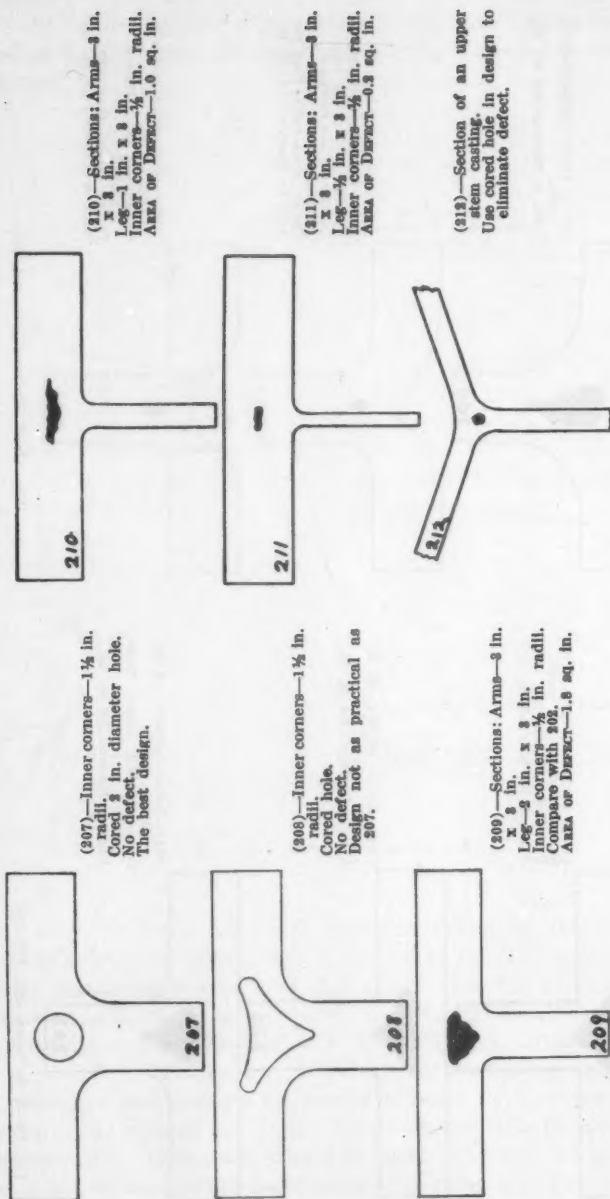
FIG. 15—T SECTIONS—207, 208, 209.

(210)—Sections: Arms—2 in. x 2 in. Leg—1 in. x 2 in. Inner corners—⅜ in. radii. AREA OF DEFECT—1.0 sq. in.

(211)—Sections: Arms—2 in. x 2 in. Leg—½ in. x 2 in. Inner corners—⅜ in. radii. AREA OF DEFECT—0.2 sq. in.

(212)—Section of an upper stem casting. Use cored hole in design to eliminate defect.
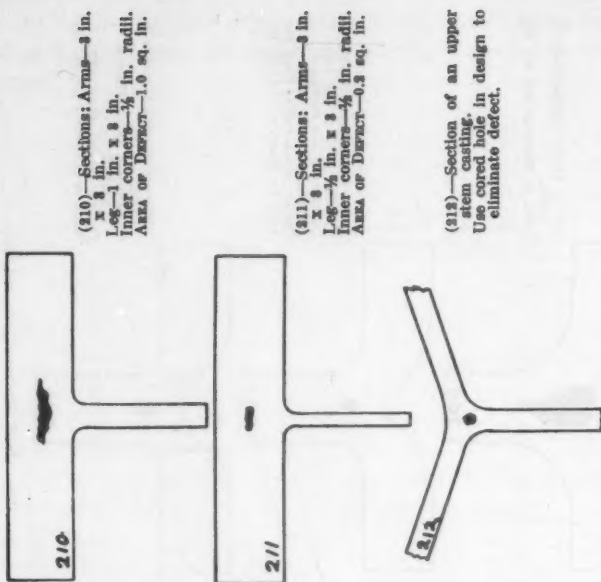
FIG. 16—T SECTIONS—210, 211, 212.

"V"-Sections

52.   A uniform section in the shape of a "V" will not be free from contraction cavities, since a synthetic hot spot is developed at the junction due to the fact that the sand is not able to conduct heat away from the three inside faces as fast as it can from the single outside face. It will be noticed that if the inner radius is increased, as in 305 (Fig. 19), that the defect will be smaller than in a comparable case such as 302 (Fig. 18). As in the case of the "L" design, a slightly reduced section at the junction of the members is necessary for a homogeneous section. In designing "V" sections, it is suggested that an inner radius not less than 1-in. be used.

"X"-Sections

53.   The designer will observe that an "X"-section could not be designed that could be made free from a contraction cavity, if the section were fed only through the arms. The defect could, however, be made quite small be designing the section with a cored hole as in 504 and 505 (Figs. 20 and 21).

54.   Much has been said in previous discussions upon casting design concerning offsetting the arms of the "X" so that the hot

(213)—Section from a structural casting.
Remedy same as 212.

(214)—Section from a stem casting.
To reduce size of defect use dip in arm at intersection as 205.
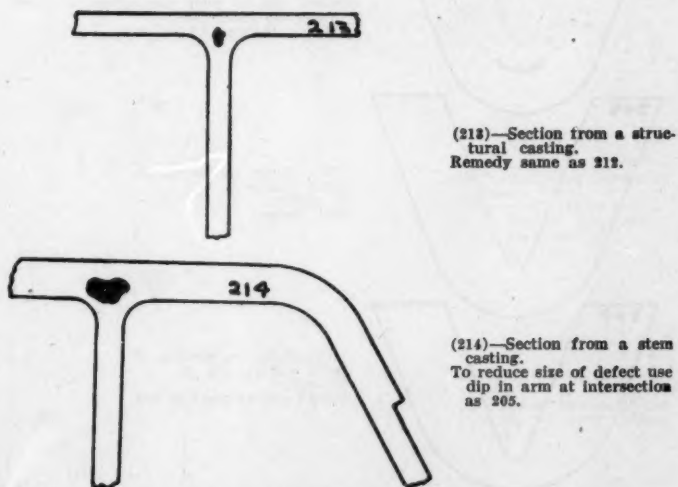
Fig. 17—T Sections—213, 214.

spot would be reduced. This is not true if the offset is similar to designs 506, 507, and 508 (Fig. 21). The defects formed by this type of construction are about the same size as would appear in the usual "X" type section. A section having one arm completely offset, as in 506, allows the foundryman to use external chills to advantage. Slightly better results could be obtained by spacing the offsets further apart, such as illustrated in 509 (Fig. 22). Such a design allows the foundryman ample freedom for the placing of external chills. Designs 510, 511, and 512 (Figs. 22 and 23) show how the defect may be decreased in size as the adjoining section is decreased. The same principles apply here as they did in the "T"-section that has previously been discussed.

55.   Unless the cored hole at the center of the hot spot is

(301)—Members intersect at 45 degrees.
Section—3 in. x 3 in.
No radii—sharp corners.
AREA OF DEFECT—3.3 sq. in.

(302)—Inner radius—½ in.
Outer radius—3½ in.
Uniform section.
Defect due to poor conduction of sand from the inner face.
AREA OF DEFECT—0.3 sq. in.

(303)—Inner radius—½ in.
Outer radius—4 in.
No defect.
Best design.

(304)—Inner radius—½ in.
Outer radius—4½ in.
No defect.
Design not as good as 303.

FIG. 18—V SECTIONS—301, 302, 303, 304.

advantageous to use, it is suggested that the one arm offset similar to 509 be used.

56. An interesting point encountered in the "Y"-section is that no matter how the features of the design are altered, the defect found was about the same size. Attention is called to 403, 404, 405, and 406 (Figs. 24 and 25). The only design without defect was 408. A triangular hole of this type is able to extend to all the hot spots that are possible in the section. Defects found in sections of commercial casting 409, 410, and 411 (Fig. 26) are exhibited for sake of comparison. Occasionally, there appears a design wherein an area of heavy metal is attached on all sides to members of much smaller thickness and so located that the foundry-

(305)—Inner radius—1 in. Outer radius—4 in. Uniform section — compare with 302.

(306)—Inner radius—½ in. Cored 2 in. diameter hole. Area of Defect—0.3 sq. in.

(307)—Portion of lower stem casting. Section could not be fed except through the arms. Design very poor.

(308)—Portion of stem post casting. Design should be modified to take off the lower portion of the v".
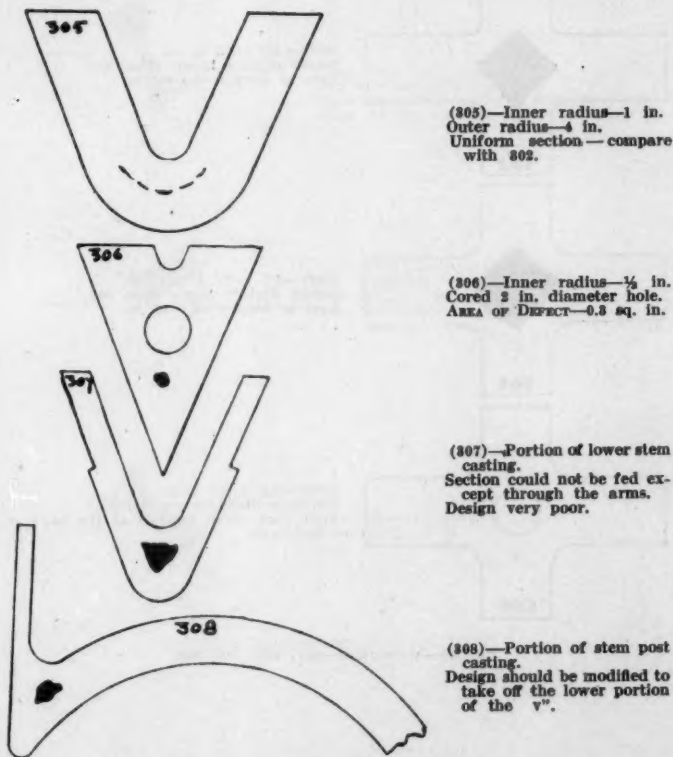
Fig. 19—V Sections—305, 306, 307, 308.

(501)—Section 3 in. x 3 in.
Sharp corners—no radii.
AREA OF DEFECT—5.5 sq. in.

(502)—All radii ½ in.
Defect slightly larger than 502.
AREA OF DEFECT—6.4 sq. in.

(503)—All radii 1 in.
Defect slightly larger than 503.
AREA OF DEFECT—8.1 sq. in.

(504)—All radii 1 in.
Two-inch diameter cored hole.
Four very small cavities at the head of
    each arm.
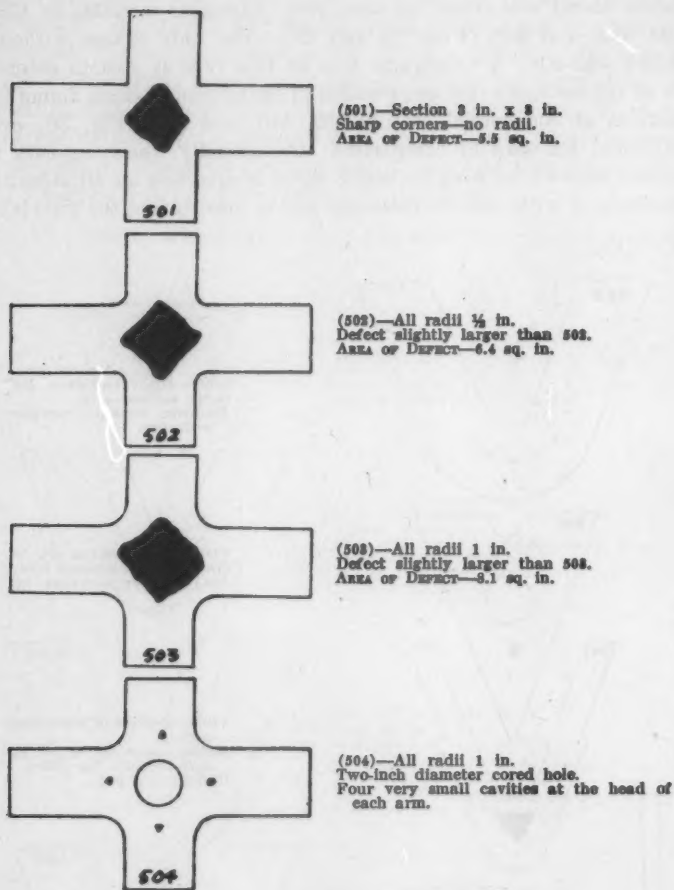
FIG. 20—X SECTIONS—501, 502, 503, 504.

(505)—All radii 1 in.
Cored rectangular hole cavities smaller
than those of 504.

(506)—All radii ½ in.
One arm completely offset.
Area of defect about the same as 503.
AREA OF DEFECT—5.5 sq. in.

(507)—Two radii ½ in.
Two radii 1½ in.
One arm ½ offset.
AREA OF DEFECT—5.3 sq. in.

(508)—Two radii ½ in.
Two radii 1½ in.
Both arms ½ offset.
AREA OF DEFECT—5.9 sq. in.
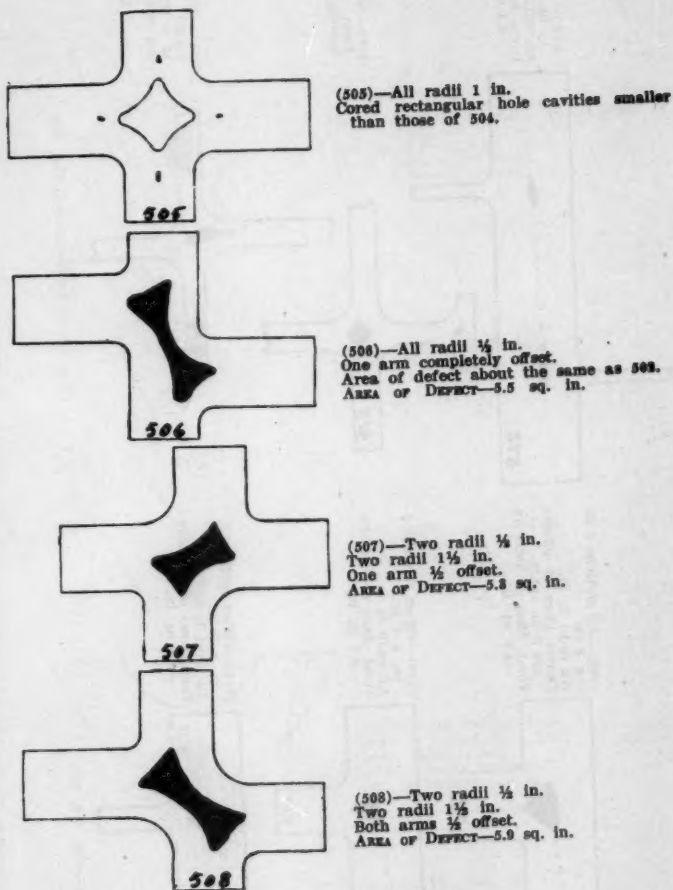
FIG. 21—X SECTIONS—505, 506, 507, 508.

(512)—Offset sections
⅞ in. x 3 in.
All radii ⅛ in.
TOTAL AREA OF DEFECTS
—0.4 sq. in.

(513)—Section taken
from a stern post
casting.

(514)—Showing defect
found in a portion
of a structural cast-
ing with arm offset.

FIG. 23—X SECTIONS—512, 513, 514.

(509)—All sections 3 in.
x 3 in.
All radii ⅜ in.
Center lines of offsets
are 3 in. apart.
TOTAL AREA OF DEFECTS
—4.3 sq. in.

(510)—Offset sections 2
in. x 3 in.
All radii ⅜ in.
TOTAL AREA OF DEFECTS
—2.0 sq. in.

(511)—Offset sections 1
in. x 3 in.
All radii ⅜ in.
TOTAL AREA OF DEFECTS
—0.9 sq. in.

FIG. 22—X SECTIONS—509, 510, 511.

(401)—Members intersect at an angle of 60 degrees.
Section—3 in. x 3 in.
Sharp corners—no radii.
AREA OF DEFECT—2.7 sq. in.

(402)—Inner radius—½ in.
Outer radii—3 in.
AREA OF DEFECT—2.7 sq. in.

(403)—Inner radius—½ in.
Outer radius—3½ in.
Lower radii—1 in.
Defect smaller than 401, 402.
AREA OF DEFECT—1.9 sq. in.

(404)—Inner radius—3 in.
Outer radius—6 in.
Lower radii—1 in.
Defect similar to 403.
AREA OF DEFECT—1.9 sq. in.
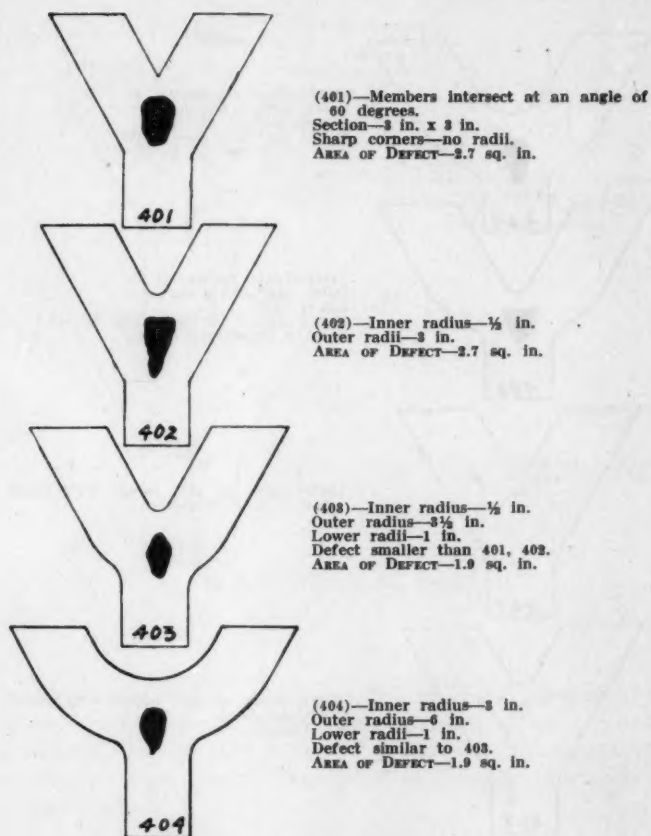
FIG. 24—Y SECTIONS—401, 402, 403, 404.
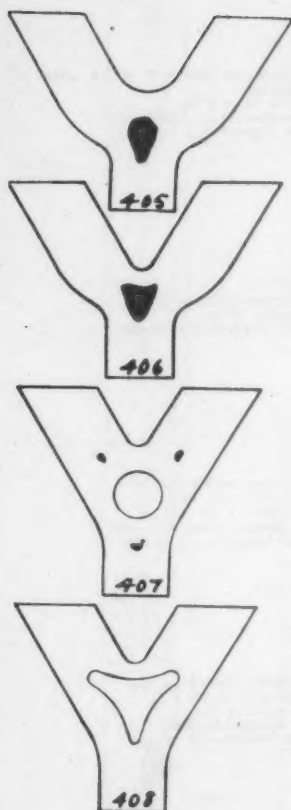
(405)—Inner radius—1½ in.
Outer radius—4½ in.
Lower radii—1 in.
Defect the same size as 403.
AREA OF DEFECT—1.9 sq. in.

(406)—Inner radius—½ in.
Outer radius—4½ in.
Lower radii—1 in.
Defect nearly the same size as 403.
AREA OF DEFECT—1.8 sq. in.

(407)—Same as 402, except with cored
hole 2 in. in diameter.

(408)—Same as 402, except with cored
triangular hole.
No DEFECT.

FIG. 25—Y SECTIONS—405, 406, 407, 408.

(409) — Portion from the upper stem structural casting showing type of defect found.

(410)—Section of the lower stem casting showing double "Y" and type of defect encountered.

(411)—Section of a structural casting showing type of defect found.
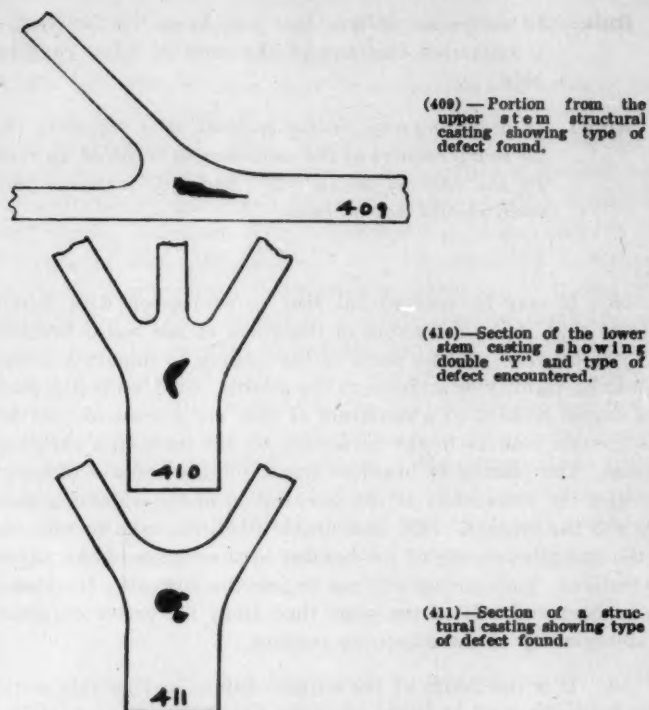
Fig. 26—Y Sections—409, 410, 411.

man has no opportunity to properly feed the heavy portions by means of a conveniently placed feed reservoir. Such a condition is distinctly one of poor design. The heavy section should either be cored or more closely investigated to ascertain if it could not be made lighter.

57. . A study of all these various sections leads to a few general rules:

Rule: *In designing unfed joining sections in "L" or "V" shapes, it is suggested that all sharp corners at the junction be replaced by radii so that this section becomes slightly smaller than that of the arms.*

Rule: *In designing sections that join in an "X"-section, it is suggested that two of the arms be offset considerably.*

Rule: *In designing any joining sections, it is suggested that all sharp corners at the junctions be replaced by radii. In the case of unfed "T" and "X"-sections these radii should not be large.*

Ribs

58. It may be pointed out that in connection with joining sections that it is the custom of designers to use webs, brackets, and ribs between various parts of the casting to impart a certain amount of rigidity or stiffness to the casting. Such stiffening members should be kept to a minimum as they are sources of considerable trouble both as to the formation of hot tears and shrinkage cavities. The placing of brackets across joining sections of course increases the mass effect at the intersection of the adjoining members and the bracket. This undesirable condition may be remedied by the extensive coring of the bracket in the region of the adjoining sections. Such coring will not impair the stiffening features of the bracket and will at the same time allow for better conditions of homogeneity of the adjoining sections.

59. It is the desire of the authors before leaving this section entirely, to impress upon the designer two things, one of which is that the sections which have been illustrated were not fed, and had it been possible to attach a reservoir of liquid metal to the sections they would have been free from contraction cavities. The foundryman will go to considerable trouble to feed these sections. It is only those that are located in impossible positions that he fails to reach. The second point to be stressed is that even though the designer is unable to design a section which if not fed will be free from contraction cavities, he can design it so that it will have, under the circumstances, the smallest cavity possible. Then, with this design the foundryman can apply the tools and knowledge at his command such as the use of padding, external or internal chills to make the section homogeneous.

60. It is suggested that the designer look over the charts of adjoining sections in the foundryman's portion of this report so that he may see what can be accomplished by chilling unfed sections.

THE EFFECT OF MASS UPON THE MECHANICAL PROPERTIES OF CAST STEEL

61. The effect of mass upon the mechanical properties of cast steel is a subject that is of practical importance to designing engineers, particularly those engaged in designing power plant equipment where steel castings are being installed in systems subject to greater pressures and higher temperatures than those previously used. Engineers have relied upon data collected on the properties of cast steel, such as that assembled by Lorig and Williams[5]. They realize, as was noted by Lorig and Williams, that such data are rarely complete in that the chemical analyses, size of casting, heat treatment, size of the specimen or the method of testing are not always known. As many of the data are obtained from laboratory heats produced under ideal conditions, the results may not be comparable to those obtained in a commercial casting. In commenting on this condition, Sisco, in his volume of the Alloys of Iron Monographs, pointed out that of all the variables which may affect the mechanical properties of cast steel, and one which is seldom mentioned in the compilation of data, the effect of mass is clearly the most important.

TESTING METHODS

62. A study of the previous work on this subject shows that no data have been reported upon the effect of mass on the mechanical properties of cast steels in both the as-cast and annealed conditions. As the designing engineer should have some indication of what may be expected as mechanical properties in various sections, it was deemed advisable to study this effect. It should be noted, however, that these data can refer only to the analyses studied and can only be indicative of the properties of other cast steels produced in similar sections.

63. In order to study the effect of mass upon the mechanical properties, well-fed coupons were cast horizontally in medium carbon steel and medium manganese steel. The coupons were all 12-in. long and those cast in medium carbon steel were ½ x ½-in., 1 x 1-in., 2 x 2-in., 3 x 3-in., 4 x 4-in., and 8 x 8-in. in cross section. Only the 1 x 1-in., the 3 x 3-in. and one-8 x 8-in. coupons were cast in medium manganese steel for these tests. Each coupon was top poured through the feed head. The feed heads, along the entire upper faces of the coupons, were large

enough to permit the shrinkage cavities to form within them
without extending into the coupons. The feed heads were sawed
from the coupon prior to heat treating. One coupon of each size
was annealed at 900°C. (1650°F.) for the customary time of one
hour per inch of section. The coupons were then cut and tensile
specimens machined from the locations shown in Fig. 27. The
specimens were all standard 0.505-in. diameter tensile specimens,
except those machined from the ½ x ½-in. coupon. These speci-
mens were 0.313 in. in diameter with a gauge length of 1.25 in.

64. After the tensile tests were obtained, specimens were
machined from the threaded portion of the bars for micrographic
studies, density measurements and chemical analyses.

65. In another heat, bars ½, 1, 2, and 3-in. in diameter and
coupons, similar to those mentioned previously, were cast from
low carbon steel. One-half of the bars and one-half of the
coupons were annealed as before. Tensile test bars were ma-
chined from the coupons and tested. The bars were tested in
the unmachined condition in order that some indication of the
actual strength of cast sections might be obtained.

MASS EFFECT ON MEDIUM CARBON CAST STEEL

66. The data obtained from the tensile specimens machined
from the medium carbon cast steel at the locations shown in
Fig. 27 are given in Table 1. It will be noted that the yield point
in the as-cast condition is listed in only a few instances. The rep-
resentative stress-strain curve (Fig. 28) shows that in the as-cast
condition, cast steel has no marked yield point. The yield point
must therefore be determined from a stress-strain curve. The
data of Table 1 together with the density data as set forth in
Table 2 are shown diagrammatically in Fig. 29.
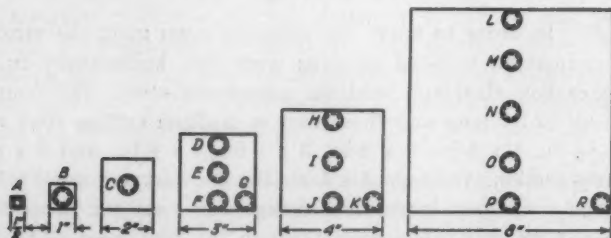
FIG. 27—LOCATION OF TEST BARS IN THE CARBON CAST STEEL BLOCKS.

67. The most outstanding feature exhibited by the curves is the sharp bends that are found in the tensile, yield and reduction of area curves in the neighborhood of the 2-in. section. In general, all of the mechanical properties in the 8-in. section were lower than those found in the ½-in. section. However, the loss was not as great in the annealed state as it was in the as-cast condition. The density curve also shows a drop in value as the mass of the test block increases.

68. If the properties of any one section are considered, it may be noticed that the lowest results are obtained at the center
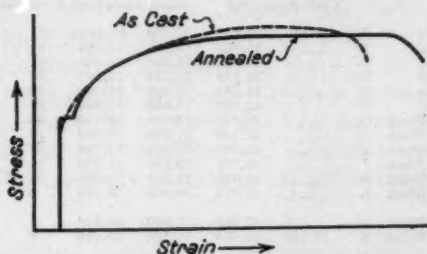
FIG. 28—TYPICAL STRESS-STRAIN CURVES FOR CARBON CAST STEEL.
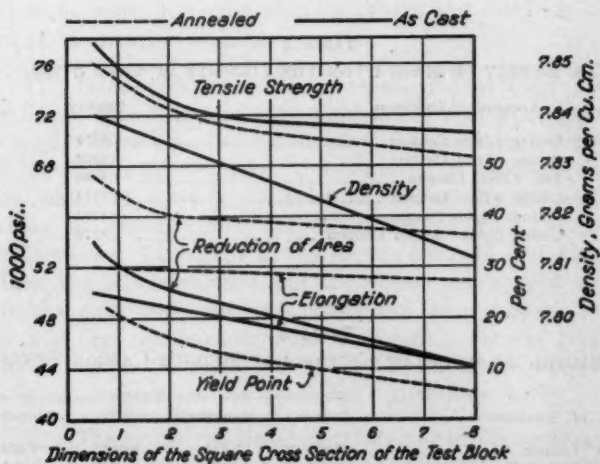
FIG. 29—EFFECT OF MASS UPON THE MECHANICAL PROPERTIES OF CARBON CAST STEEL AT THE CENTER OF THE TEST BLOCK.

of the section and the highest in the bottom corners where solidi-
fication has proceeded from two faces. The corner results are
usually higher than those obtained elsewhere along the bottom
where there is only one solidifying face. These conditions are
more pronounced, of course, when the steel is tested in the as-

Table 1

THE EFFECT OF MASS UPON THE MECHANICAL PROPERTIES OF MEDIUM CARBON CAST STEEL

Specimen				Yield Point Lbs. Per Sq. In.		Tensile Strength Lbs. Per Sq. In.		Per Cent Elonga-tion		Per Cent Reduction in Area	
				As Cast	Annealed	As Cast	Annealed	As Cast	An-nealed	As Cast	An-nealed
Center	½″	Block	A	38,200	49,000	77,600	75,450	25	30	35	47
Center	1″	Block	B	....	48,000	75,500	75,000	25	28	30	46
Center	2″	Block	C	....	46,750	73,000	74,500	21	30	26	40
Top	3″	Block	D	....	44,750	71,750	71,500	18	28	19	35
Center	3″	Block	E	....	45,200	72,300	72,000	20	29	23	39
Bottom	3″	Block	F	....	45,000	74,000	72,400	21	28	26	40
Lower Corner	3″	Block	G	....	45,250	73,500	73,250	23	29	30	42
Top	4″	Block	H	....	45,250	71,000	70,000	15	29	21	43
Center	4″	Block	I	....	44,500	71,200	70,500	19	29	23	39
Bottom	4″	Block	J	....	45,750	73,250	72,250	22	30	21	46
Lower Corner	4″	Block	K	....	45,500	73,700	73,000	25	30	33	46
Top	8″	Block	L	....	41,750	70,000	70,125	8	26	9	40
Center of upper half	8″	Block	M	....	43,000	71,000	68,250	9	26	9	40
Center	8″	Block	N	33,500	42,000	70,500	68,500	9	27	10	36
Center of lower half	8″	Block	O	....	41,750	74,250	68,750	16	28	18	41
Bottom	8″	Block	P	....	43,750	76,100	70,750	18	29	21	44
Lower Corner	8″	Block	R	....	43,500	76,750	72,125	22	29	30	44

Table 2

THE EFFECT OF MASS UPON THE DENSITY OF CAST STEEL

Location of Specimen	Density
Center 1-Inch Coupon	7.8379
Center 8-Inch Coupon	7.8306
Top 8-Inch Coupon	7.8106
Center 8-Inch Coupon	7.8114
Bottom 8-Inch Coupon	7.8122
Lower Corner 8-Inch Coupon	7.8175

Table 3

CHEMICAL ANALYSES OF SECTIONS OF MEDIUM CARBON STEEL

Location of Specimen	Per Cent			
	Carbon	Manganese	Silicon	Aluminum
Center 1″ Coupon	0.270	0.634	0.212	0.035
Top 8″ Coupon	0.265	0.634	0.229	0.028
Center 8″ Coupon	0.243	0.623	0.223	0.040
Corner 8″ Coupon	0.268	0.637	0.230	0.032

cast condition, but they are also noticeable in the heat treated material.

69. Since the mechanical properties varied to such a considerable degree, it was thought that perhaps segregation of chemical compositions was responsible for a large portion of the difference. Phosphorus and sulphur segregations were studied in the various sections by etching and sulphur prints, but no differences could be ascertained. This was perhaps due to the fact that they were both present in very low amounts in the steel as cast. As the steel was made by the basic process, these analyses were 0.02 per cent for the phosphorus and 0.03 per cent for sulphur. Segregation was also studied by chemical analysis. The centers of each section were analyzed for carbon, manganese, silicon and aluminum and the results are set forth in Table 3. The greatest difference appeared in the carbon content where 0.03 per cent was noted. The center of the 1-in. coupon contained 0.27 per cent, while the center of the 8-in. coupon contained 0.24 per cent. The manganese, silicon and aluminum analyses showed that there was practically no segregation of these elements.

70. The specimens also were studied microscopically in the unetched condition to see if the inclusions had formed an aluminum network which might tend to lower the physical properties in the larger cast sections. No network was evident. The inclusions were, however, slightly larger and fewer in the centers of the heavier sections.

71. It appears from the above study that there are three conditions that bear upon the mechanical properties as affected by mass. These are: (a) the density, (b) the carbon segregation, and (c) the microstructure. It is difficult to ascertain from the data which of these is the most important. As the section increases, the density drops, the carbon content drops and the grain size increases. In regard to the tensile strength and yield point values, the microstructure and the carbon segregation are probably the most important in causing the lower values. The density and the microstructure are responsible for the lower ductility found in the heavier sections. From this it appears that the microstructure is the dominant controlling factor.

72. The properties listed at the center of the section are the poorest ones present in this section. The average properties over the entire section are better than those exhibited by the center,

due to the preponderance of material having better properties, such as those exhibited by the corners and edges.

THE MASS EFFECT IN MEDIUM MANGANESE STEEL

73. The effect of mass upon the mechanical properties of the medium manganese steel studied (Fig. 30) is similar to that found in the medium carbon steel. In each case, the mechanical properties at the center of the test coupon decrease as the mass of the section increases. The variation in the tensile strength is, however, considerably greater in the manganese steel than it is in the plain carbon steel. The tensile strength in the plain carbon steel in the center of the 8-in. coupon is 5000 lb. per sq. in. less than that in the 1-in. coupon, a drop of 6.5 per cent. In the medium manganese steel, the decrease in similar coupons is 10,400 lb. per sq. in., a change of 10.5 per cent. It should be noted that the change in the tensile strength of the manganese steel decreases gradually and rather uniformly with an increase in mass, whereas the tensile strength of the plain carbon steel decreases rapidly at first and then at a decreasing rate as the mass increases.

74. The other tensile properties of the medium manganese steel also decrease gradually and at a more uniform rate than in the plain carbon steels.
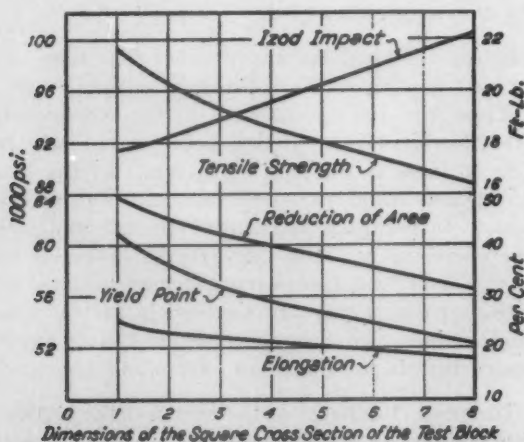
FIG. 30—EFFECT OF MASS UPON THE MECHANICAL PROPERTIES OF MEDIUM MANGANESE STEEL AT THE CENTER OF THE TEST BLOCK.

Table 4
THE EFFECT OF MASS UPON THE MECHANICAL PROPERTIES OF MEDIUM MANGANESE CAST STEEL

Specimen	Heat Treatment °F.	Yield Point	Tensile Strength	Elongation	Reduction of Area	Ft.-Lbs. Izod
		Lbs. Per Sq. In.		Per Cent		
Center 1" Coupon	Annealed 1650—1 hr.	60,750	99,000	25	49	17.7
Center 3" Coupon	Annealed 1650—3 hr.	56,750	94,750	22	42	18.6
Center 8" Coupon	Annealed 1650—8 hr.	52,250	88,625	18	31	22.3
Top 8" Coupon	Annealed 1650—8 hr.	52,750	88,125	18	31	......
Bottom 8" Coupon	Annealed 1650—8 hr.	56,000	95,000	23	41	......
Corner 8" Coupon	Annealed 1650—8 hr.	56,000	95,000	24	50	19.2
Midpoint 8" Coupon	Annealed 1650—8 hr.	55,750	92,250	21	35	......
*C-4	Annealed 1650—1 hr.	61,375	91,375	20	52	......
*C-5	Annealed 1850—1 hr.	56,000	88,650	20	50	......
*D-3	Annealed 2050—1 hr.	58,625	99,125	22	46	14.6
*D-4	Annealed 2250—1 hr.	60,000	99,675	17	36	......
*D-6	Annealed 1850—15 min.	60,650	96,125	22	42	......
*E-5	Annealed 2250— 5 min.	60,750	96,875	22	41	16.3
*E-4	Annealed 2050— 5 min.	....	99,650	18	40	......
*F-4	Annealed 1650—1 hr.	....	....	....	....	......
*F-4	Reannealed 1525—1 hr.	68,250	96,375	22	48	......
*F-5	Merten's Treatment	51,500	90,000	25	50	......

* Specimens previously annealed in coupon at 1650°F. for 8 hours.

Table 5

CHEMICAL ANALYSES OF SECTIONS OF MEDIUM MANGANESE STEEL

Location of Specimen	Per Cent			
	Carbon	Manganese	Silicon	Aluminum
Center 1" Coupon	0.340	1.46	0.322	0.057
Center 3" Coupon	0.323	1.46	0.320	....
Top 8" Coupon	0.320	1.46	0.319	....
Center 8" Coupon	0.297	1.41	0.312	0.056
Bottom 8" Coupon	0.333	1.47	0.322	....
Corner 8" Coupon	0.342	1.48	0.321	....

Table 6

THE COMPARISON OF THE MECHANICAL PROPERTIES OF WELL FED TEST COUPONS AND BARS CAST TO SIZE

Carbon 0.31   Manganese 0.60   Silicon 0.31

Section	Gauge Length	Yield Point		Tensile Strength		Per Cent Elongation A.C. Ann.		Per Cent Red. of Area A.C. Ann.	
		As Cast	Annealed	As Cast	Annealed				
½" Coupon	Standard Tensile Spec.	....	44,600	69,800	70,700	29	34	40	50
1" Coupon		....	42,500	69,000	69,700	28	33	38	48
2" Coupon		....	41,750	67,000	69,000	20	32	23	44
3" Coupon		....	40,750	64,500	67,750	18	32	24	40
½" Diameter Bar	2"	....	43,000	69,000	69,500	11	20	13	25
1" Diameter Bar	4"	58,500	40,000	67,100	64,600	12	16	12	30
2" Diameter Bar	8"	54,200	39,200	61,800	64,000	8	13	8	23
3" Diameter Bar	12"	....	39,500	....	63,550	....	14	....	13

75. The tensile properties of various sections of the 8-in. coupon vary in a manner similar to that found in the plain carbon steel. Table 4 indicates that the best tensile properties are found in the lower corner and the bottom of the test coupon. The top of the coupon, immediately under the head, exhibits the poorest properties and the properties then increase gradually toward the bottom of the coupon. The Izod impact and strength, however, increases as the mass of the section increases and is greater at the center of the 8-in. coupon than it is in the corner of the coupon.

76. Chemical analyses were obtained from the center of the 1-in. and 3-in. sections and from the top, center, bottom and corner of the 8-in. section in order that the amount of segregation could be determined (Table 5). Segregation of carbon can be seen in that the carbon content at the center of the coupon decreases as the mass of the section increases. The carbon content at the center of the 8-in. section is the same as that at the center of the 1-in. coupon. There is slight segregation of manganese but no segregation of silicon or aluminum.

77. This segregation of carbon and manganese probably will account for the decrease in tensile strength as the mass is increased.

COMPARISON OF COUPONS AND BARS CAST TO SIZE

78. One of the points upon which designers have desired information, is the comparison of the mechanical properties of cast steel sections and the properties exhibited by the well fed coupon. This point was investigated and the results are set forth in Table 6. In this experiment, bars were cast in the horizontal position while coupons were cast similar to those used previously. From the center of these as-cast and annealed coupons, standard tensile bars were machined and tested. These specimens likewise showed the effect of mass on the mechanical properties.

79. The bars were tested in the condition they came from the sand mold, except that one set was annealed before testing. The bars were not machined nor were the cast surfaces removed.

80. The yield and the tensile strengths were nearly the same in both bars and specimens machined from the coupons. The ductility, however, was much lower in the case of the bars. This

is undoubtedly due to the rough surface of the bars. The decrease in ductility with increasing mass is much greater with the bars than it is with the coupons.

81.   These data point out that while the strength properties in cast sections are nearly the same as those represented by the coupon, the ductility properties are much lower and should be borne in mind by the designer. The interesting point is that the ductility results are rather erratic and would depend a great deal on the surface of the casting, and probably on the size of the axial weakness.

82.   The purpose in presenting data on bars cast to size is to inform design engineers of the properties of castings that may have sections similar to those tested. It is requested that the data obtained on the machined test specimen be not directly compared to the data resulting from testing the cast bars, since the latter is a notable variation from the standard test bar and will not give comparable results. The machined specimen taken from the coupon gives an indication of the best properties of the metal. The data obtained from testing the cast bars give an indication of the properties of an actual cast section. It is well known that a rough surface will result in lower ductility values being presented, but it should also be pointed out that most castings are used in commercial installations with the cast surfaces exposed and that the actual mechanical properties of that section, as it stands in the installation, are more along the lines of the data as presented for the cast bars than they are according to the data collected from the coupon. Thus, the properties are reported as an indication of what the section will withstand and it is not at all a test of the most favorable properties of the steel, nor a direct comparison of two types of test specimens.

83.   The above section on the effect of mass on the mechanical properties of cast steel can be summarized as follows:

   (1)   There is a loss in strength and ductility, as measured at the center of the section, as the mass increases.
   (2)   In the carbon steel studied the loss is pronounced for the first 2-in. of cross section, after which it tapers off gradually. In the manganese steel, there is no decided knee to the curve.
   (3)   There is a decrease in density and carbon content, and an increase in Izod impact value as the mass increases.
   (4)   Microstructure, carbon segregation and density values

are responsible for the decrease of mechanical properties as the mass increases.

(5) Segregations of silicon, aluminum, phosphorus and sulphur were small and had no apparent effect as to producing low mechanical properties.

(6) The mechanical properties of sections as represented by cast bars show that strengths are quite similar to the optimum values as recorded by the machined test specimens, but that the ductility is much lower.

84. The above discussion in regard to mass effect leads to the following statement:

Rule: *In a general way, it may be said that the effect increasing mass has on the mechanical properties is not pronounced and probably is amply covered by the present factor of safety.*

Suggested Rules for Steel Casting Design

85. The suggested rules which have been presented in the first part of this paper are as follows:

(1) An attempt should be made to design all sections in a casting with a uniform thickness.

(2) It is not desirable to design cast steel structures with abrupt change in section.

(3) Sharp corners at adjoining sections are sources of defects, and, if possible, should be eliminated.

(4) When a design of a cast steel structure becomes very complicated or intricate, it is suggested that it be broken up into parts, so that they may be cast separately and then assembled by welding.

(5) In designing unfed joining sections in "L" or "V" shapes, it is suggested that all sharp corners at the junction be replaced by radii so that this section becomes slightly smaller than that of the arms.

(6) In designing sections that join in an "X"-section, it is suggested that two of the arms be offset considerably.

(7) In designing any joining sections, it is suggested that all sharp corners at the junctions be replaced by radii. In the case of unfed "T" and "X"-sections, these radii should not be large.

(8) In a general way, it may be said that the effect increasing mass has on the mechanical properties is not pronounced and probably is amply covered by the present factor of safety.

Part II — The Foundryman and Steel Casting Design

INTRODUCTION

86. In discussing the question of design from the foundry viewpoint, only the shape and size of the mold cavity will be considered. The composition of the mold may, of course, affect the soundness of the casting by producing hot tears, porous spots, or sand inclusions. This is in itself a most comprehensive subject, but one which is controlled largely by the type of materials available locally and the sand treating and sand testing equipment at hand. What may be expedient for one foundry may be entirely out of the question for another concern. The mold, therefore, will be considered merely as a medium in which a cavity similar to the finished casting may be produced.

87. It has been stated previously that the defects that may be attributed to the design of the casting are shrinkage cavities and hot tears. It should be noted that while many hot tears are due to mold composition and the manner of molding, a large portion of the hot tears found in castings are inherent in the casting design and cannot be eliminated by mold relieving.

88. Shrinkage cavities are produced by the contraction of the metal upon solidifying and, theoretically, they can be eliminated only in one manner, i. e., by "controlled directional solidification." Foundrymen have for many years used "directional solidification" and "progressive solidification" as synonymous terms. To the authors, however, each of these terms has a specific meaning and, for the purpose of this discussion, it may be well to define them.

89. *Progressive Solidification.* Any liquid metal cooling in a cavity formed in a refractory or metal mold will solidify *progressively* from the mold-metal interface toward the center of the cavity. Thus *progressive solidification* is present in every casting produced.

90. *Directional Solidification.* In addition to the *progressive solidification* from the mold-metal interface toward the center of the cavity, solidification will proceed *along* the cast member in the direction of the increasing temperature gradients produced in pouring the casting. These temperature gradients *along* the cast member appear in both the metal and the mold and are determined by the shape of the mold cavity and the method of gating and heading.

They are, therefore, *controllable*, whereas the foundryman has
no control over the direction of the temperature gradients produc-
ing *progressive solidification*. If this analysis is followed to its
logical conclusion it is evident that all shrinkage cavities in cast-
ings are the result of adverse temperature gradients, or *uncontrolled
directional solidification*. Only when the temperature gradients
within the metal and the mold are such that the "feed heads"
are the last portions of the casting to solidify, and thus can furnish
liquid metal to fill the voids formed during solidification, can a
"sound" casting be produced. As in most castings, desirable
temperature gradients can be produced by using tapered sections,
proper methods of gating and heading, and properly placed chills,
it follows that *controlled directional solidification* is the governing
factor in the elimination of cavities due to shrinkage.

CONTROLLED DIRECTIONAL SOLIDIFICATION IN THE FOUNDRY

91. In the first section of this paper devoted to "design for
the engineer," it was emphasized that designing engineers should
endeavor to maintain uniform cross-sections throughout the cast-
ing. The paragraph above on directional solidification appears to
be a direct contradiction of the first statement as it advocates the
use of tapered sections in producing "sound" castings. This
apparent contradiction has been placed in this discussion for the
reason that the designer cannot know how the foundryman will
produce his casting, and any tapering of sections that he may
provide might produce temperature gradients opposing those that
the foundryman finds essential to secure a good casting. If, how-
ever, the designer maintains fairly uniform cross-sections through-
out the casting the foundryman can produce much more easily the
temperature gradients he desires.

92. An illustration of how temperature gradients produced
by tapered sections can insure correct solidification is given in the
sketch of the turbine casing casting shown in Fig. 31. If this cast-
ing is poured flange-up, the sections tapered as shown and a
collapsible core used to forestall hot tears, it is comparatively easy
to produce a sound casting. It should be noted that the padding
on the cylinder bore is brought down below the steam chamber so
that the steam chamber can be fed directly. The casting was
poured through the head so that the last metal to enter the mold
(hot metal) was in the head. Now, if external chills are placed
around the ports of the steam chamber, it can be seen readily that
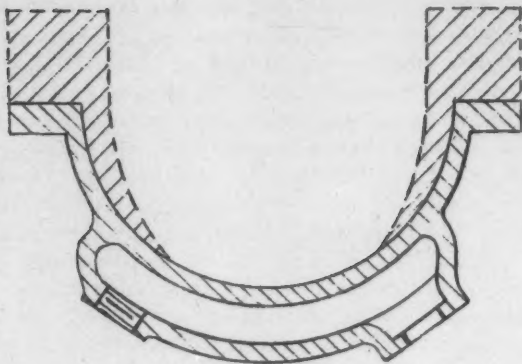
FIG. 81—APPLICATION OF PADDING.

solidification proceeds from the lower portion of the casting toward the heads. This *direction* of *solidification* has been controlled by producing favorable temperature gradients with tapered sections and this casting becomes a comparatively easy one to produce without defects.

93. Some objection may be raised to the excessive amount of machining required on the bore of the cylinder. This is merely a question of economics. Four of these castings were produced; each one was declared acceptable after a complete gamma-ray inspection. The loss of one casting, produced without the precaution of padding, would have cost more than the total extra machining charges on all four castings.

94. In many castings, particularly small and medium sized castings, the temperature gradients can be controlled by proper gating and pouring methods. These principles have been very well presented [6] by the late George Batty, and it would be superfluous to discuss them here.

MODELS AS AN AID TO FOUNDRY PRACTICE

95. In many cases, slight changes in design will enable the foundryman to produce sound castings even when it is almost physically impossible to make a casting without defects from the original design. It is, at times, difficult to obtain the consent of the customer for even slight changes in design. This is particularly true if the casting is complicated and rather difficult to visualize

from blue-prints. It may then be advantageous to prepare a model of the original design at an appropriate scale and use this model to point out the difficulties connected with the design and the modifications required to remove or elevate them.

96. This method of attack has been used to advantage by the Norfolk Navy Yard in the production of turbine castings. The models shown in this report (Fig. 32) to illustrate their advantages were prepared at the direction of S. W. Brinson, master molder of the Norfolk Navy Yard. Photographs of a completed casting prepared after the design had been modified are shown in Figs. 33, 34 and 35. If the photographs of the models and the castings are compared, it will be noted that while the changes are minor in character, as far as the engineering usefulness of the casting is concerned every effort has been made to obtain directional solidification and to simplify the production of the casting. It may be appropriate to point out a few of the changes that were made:

(A) The base casting was separated at *(A)* and the high pressure end was cast separately. This facilitated the production greatly.

(B) The thickness of the gland supporting web was increased and the cored "pockets" eliminated by changing the shape of the web.

(C) The bolt bosses were tapered from top to bottom to facilitate feeding.

(D) The thickness of the steam belt was increased and an opening provided for core anchorage. This opening was later closed by a plate welded on.

(E) The radius at the junction of the flange and web was increased to eliminate a sharp change in section and permit feeding of the sections at the bottom.

(F) The thickness of the web was increased and a larger radius provided to insure proper feeding.

(G) This outlet was cast separately and welded on.

(H) The wall thickness of the steam belt was increased and the outlet was cast separately and welded on. As the castings were to be used as a "right" and a "left" two openings were made to secure proper core anchorage. One of these openings was sealed by a welded cap, the other was welded to the outlet case separately. This permitted the use of identical cores for both the "right" and "left" castings.

(I) The web was cast separately and welded on to the main casting. This web, if cast integrally, would very probably have caused hot tears, which would have had to be repaired by welding after considerable chipping.

(J) The walls of the by-pass valve were brought straight down to the casing to eliminate sharp changes in direction of sec-
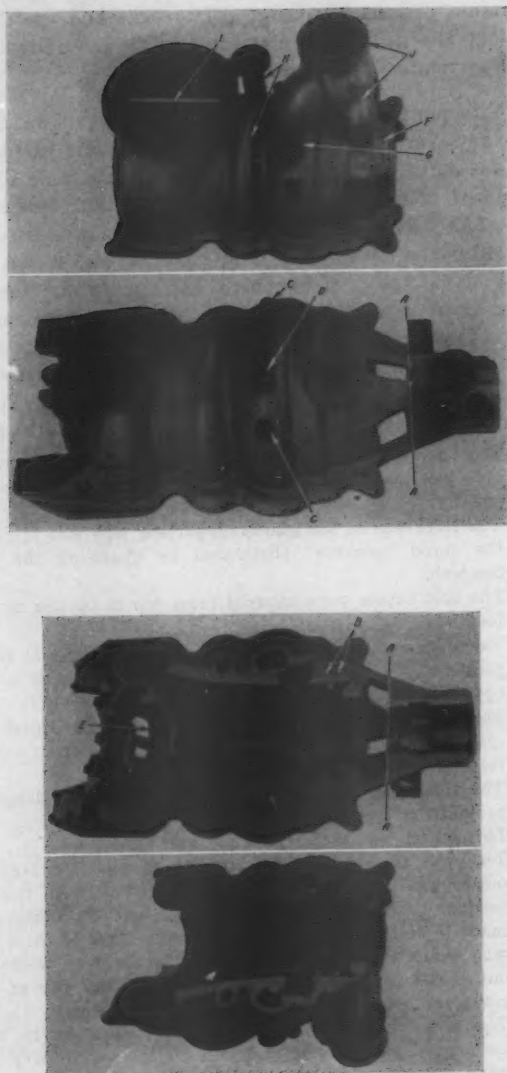
FIG. 32—MODEL DESIGNS.
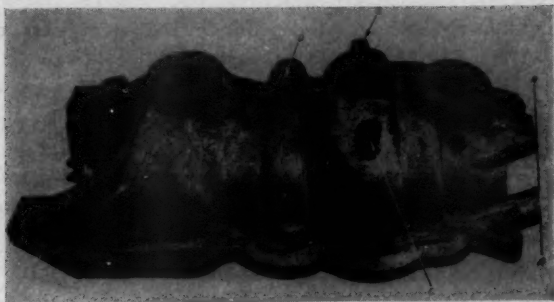
Fig. 33—Turbine Base.
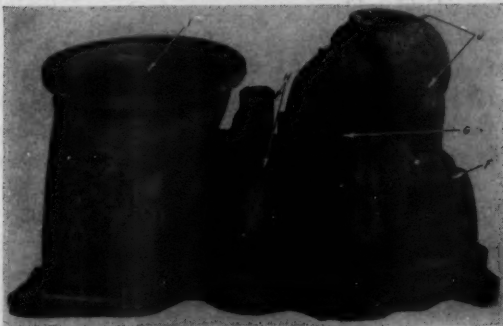
Fig. 34—Turbine Cover.
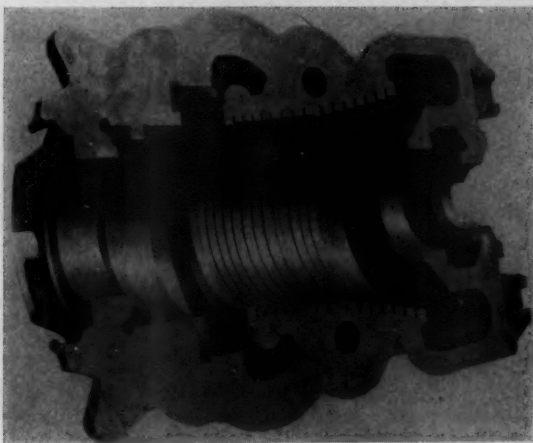
Fig. 35—Machined Turbine.

tion and permit better feeding. The outlet was cast separately and welded on.

97. In addition to the changes listed above, padding was provided to insure proper feeding. This extra metal was removed from the casting in the chipping and cleaning operation. These castings were cast flange-up in a manner similar to that mentioned previously. The castings were radiographed and were found to be exceptionally free from defects.

98. In this particular case, the use of a model of the casting as originally designed not only resulted in changes in design which facilitated production, but it also reduced the cost of pattern construction considerably.

RADIOGRAPHY IN THE FOUNDRY

99. If several castings are to be prepared from one pattern, radiographic tests become a distinct aid to the foundryman. In the case of the turbine castings mentioned above, one casting of each type was prepared and examined thoroughly before production was started on the remainder of the order. In this case, due to the changes in design, no changes were required. In the production of other castings, however, defects have been found in the first casting which were eliminated from all succeeding castings by slight changes in the pattern to facilitate directional solidification.

CHILLING OF STEEL CASTINGS

100. Proper chilling is also an important factor in the production of sound castings. There still are sections of castings which it is impossible to feed and in these cases the foundryman's only expedient in the production of sound castings is the chill. There have been many arguments concerning the use of external and internal chills. The authors prefer the use of external chills for several reasons. The size of the chill employed is important when either external or internal chills are used, as the effect of the chill may prevent feeding if too large, and fail to accomplish anything if too small. With internal chills, other factors arise. The chill must fuse into the cast matrix if a perfectly sound casting is to be obtained. The chill must be absolutely clean and dry if blow-holes or porous spots are to be eliminated. In green sand practice the chills can easily be inserted just prior to closing the mold and the cleanliness of the chill is comparatively easy to control. This is not always true in dry sand practice, for if the

chills must be inserted into the green mold, the fumes from the organic binders leave a deposit on the chills when the mold is dried, which invariably causes trouble in the finished casting.

101. The effect of internal chills of various sizes upon cast sections was studied by the authors in a very simple experiment. Ordinary well-fed coupons with cross-sections of 1-in. and 3-in. were molded in dry sand. After the molds were dried, chills 3/32, 1/8, 1/4, 3/8, 1/2 and 5/8 in. in diameter, respectively, were inserted in the mold. These chills had previously been cleaned in an acid, then in a caustic bath, and nickel-plated to insure cleanliness. The molds were bottom-poured so that the steel would rise evenly over the entire chill.

102. This experiment, of course, indicates the effect of chills only upon metal that rises over the chill. A chill placed so that a large quantity of metal would run past it before the mold was filled, would have an entirely different effect upon the metal as the chill would heat up slowly and a portion of it might be erroded.
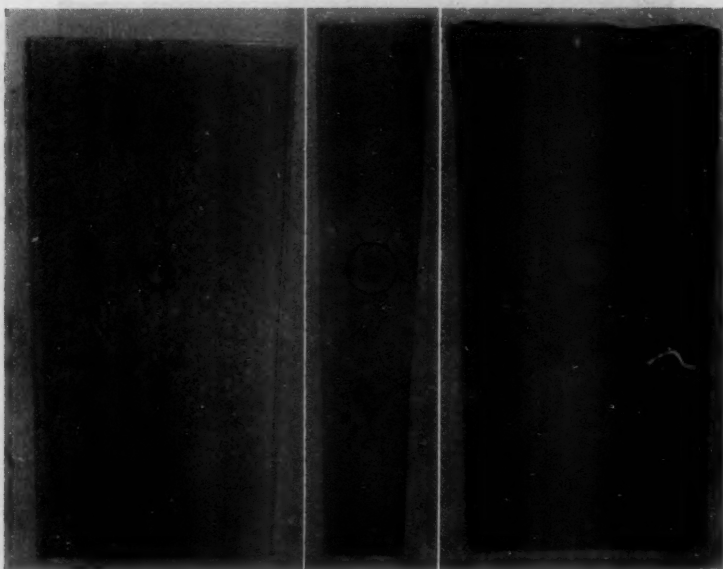
Fig. 36—(A) 1/2 in. Chill in 3 in. Coupon. (B) 5/8 in. Chill in 1 in. Coupon. (C) 5/8 in. Chill in 3 in. Coupon.

In these circumstances, a considerably larger chill would be required to accomplish the same purpose.

103. If the photomacrographs of the etched sections of some of these chilled coupons are examined, several interesting points may be noted. The ⅝-in. diameter chill did not "fuse in" in either the 1-in. or 3-in. section. The mechanical bond in the 1-in. section was so poor that upon deep etching the acid penetrated the crack and enlarged it. (B of Fig. 36.) It should also be noted that the chilled envelope, that first formed on the chill, cracked as the chill was heated and expanded. The dense chilled area extends across the entire section and, therefore, no feeding took place below the chill. A small secondary pipe may be seen about mid-way between the chill and the bottom of the casting. A radiograph of this section is shown as Fig. 37.

104. The ⅝-in. chill placed in the 3-in. section (C of Fig. 36) also failed to fuse in, but the mechanical bond was considerably better than that found in the 1-in. section. A slight crack was also formed in this case. The area of dense chilled material is
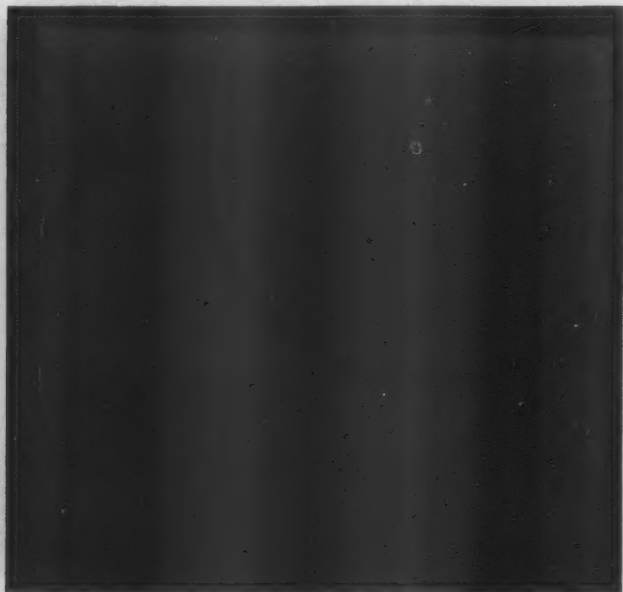
FIG. 37—RADIOGRAPH OF ⅝ IN. CHILL IN 1 IN. COUPON.

about the same size as that found in the 1-in. section. In this case, the chilled area is surrounded by a ring of porous dendritic material which is a semi-circular line similar to what is commonly called "axial weakness."

105. The ½-in. diameter chills also failed to fuse in properly in either section (A of Fig. 36 and B of Fig. 38) and the effect, while slightly less in magnitude, is similar to that of the ⅝-in. chill.

106. The ⅜-in. rod was the largest chill that showed any indication of fusion with the cast material in the 3-in. section. The photomacrograph of this section is shown as A, Fig. 38. In this case, the chilled area is quite small. An indication of a small circumferential crack may be seen just above the chill. A photomacrograph of the fusion zone in this section is shown as A of Fig. 39. The unfused chill-cast metal interface of the ½-in. chill in the 1-in. section (B of Fig. 39) is also shown for comparison.
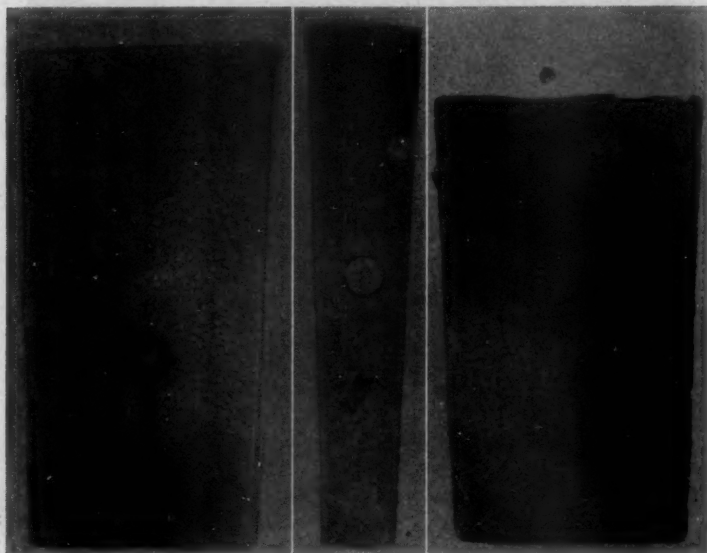
FIG. 38—(A) ⅜ IN. CHILL IN 3 IN. COUPON. (B) ½ IN. CHILL IN 1 IN. COUPON. (C) HORSE SHOE NAIL CHILL IN 3 IN. COUPON.

107. The ordinary horseshoe nail chill cast into a 3-in. section is shown in C, Fig. 38. In this case also, the chill is welded to the cast material but, again, the chilling action has been slight for the chilled area is very small. The large internal stresses brought about by chilling may be imagined by noting the diagonal cracks appearing in the head of the horseshoe nail. These cracks have, of course, been enlarged by the deep etching treatment but they were visible in the rough-ground unetched specimen. Similar cracks are visible in the chills shown in A, Fig. 36 and C, Fig. 40.

108. The ⅛-in. diameter rod was the largest chill that showed good fusion in the 1-in. cast section. The chilled area is also quite small in this case (A, Fig. 40). The horseshoe nail did
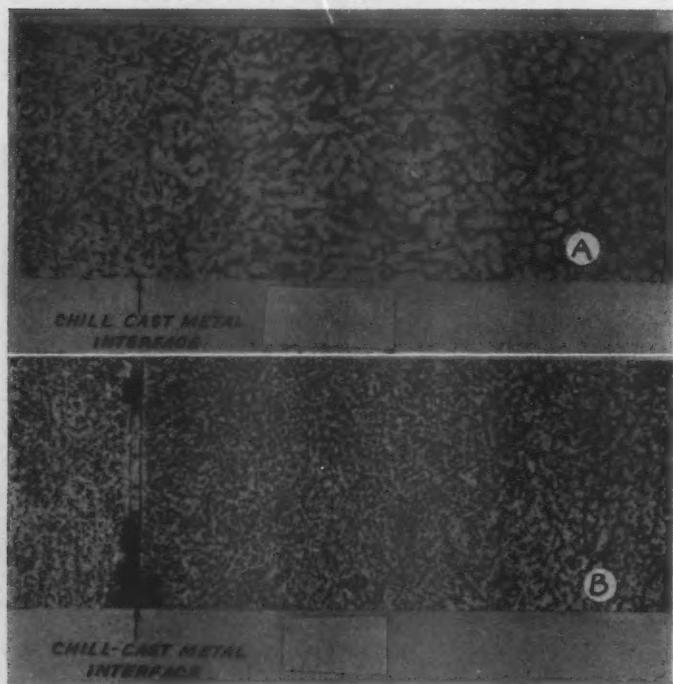
Fig. 39—(A) Fusion Zone of ⅝ in. Dia. Chill in 8 in. Coupon. (B) Unfused Chill-Cast Metal Interface of a ½ in. Chill in a 1 in. Section.
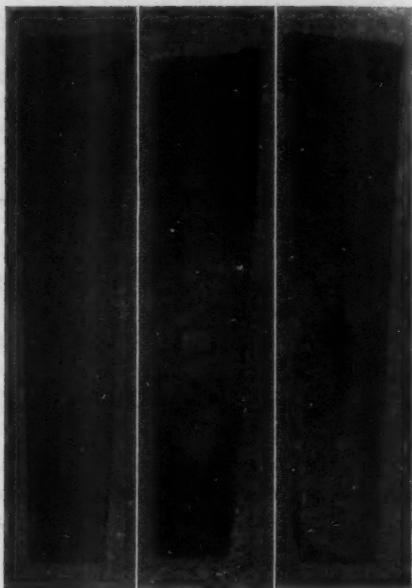
FIG. 40—(A) ½ IN. CHILL IN 1 IN. COUPON. (B) HORSE NAIL CHILL IN 1 IN. COUPON.
(C) ¼ IN. CHILL IN 1 IN. COUPON.
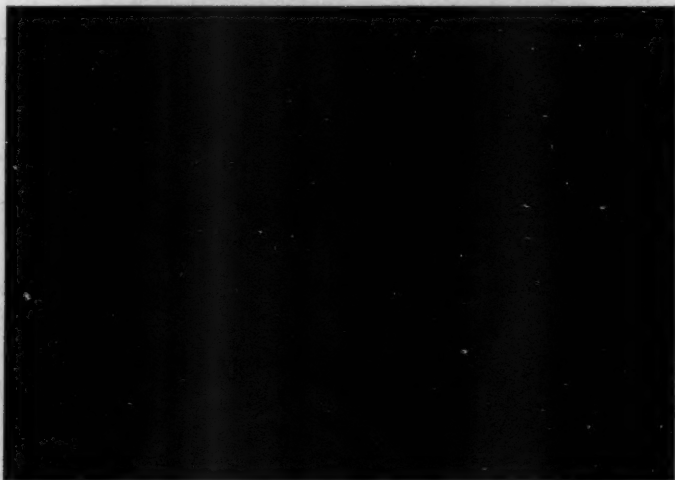
FIG. 41—RADIOGRAPH OF ¼ IN. CHILL IN 1 IN. COUPON.

not fuse into the 1-in. section at all (B, Fig. 40). The ¼-in. chill indicated partial fusion only (C, Fig. 40). A porous area and a secondary pipe are visible below the chill in the etched section and also in the radiograph (Fig. 41).

109. The two sizes of coil chills used seemed to fuse in very well. The small coil (3/32-in. dia. wire) used in the 1-in. section chilled but a small area of metal. The outline of this chilled area is visible in the etched cross-section of the casting (A, Fig. 42). The chilled area produced by the larger coil chill in the 3-in. section is not as well defined. This section is shown as B of Fig. 42. In spite of the precautions taken, this chill must have collected some dirt or water as there are porous areas above, as well as below the chill.

110. As the coupons containing these chills were available, test specimens were machined from them in such a manner that the chill passed through the center of the test specimen normal to the direction of testing. The data obtained on machined specimens of this type cannot be correlated for as the size of the chill in-
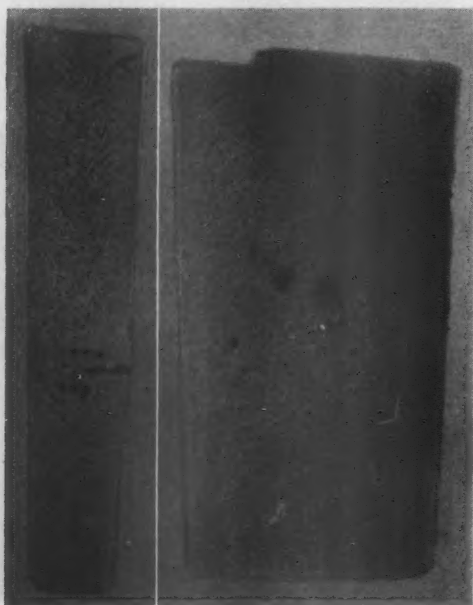
FIG. 42—(A) COIL CHILL IN 1 IN. COUPON. (B) COIL CHILL IN 3 IN. COUPON.

C. W. BRIGGS, R. A. GEZELIUS AND A. R. DONALDSON creases, the area of cast metal present in the cross-section of the specimen decreases rapidly. In the specimens containing the larger chills, the central portion of the test bars was practically all "chill," whereas in other specimens very little chill metal was present. The only point worthy of note indicated by these data was the fact that, provided the chill was actually fused in, the mechanical properties were not seriously affected.

111. The advantages in favor of the use of external chills are: (1) that fusion between the chill and the casting is not important in considering the size of the chill, (2) cleanliness is not as important, as the gases evolved can escape through the sand, and (3) in many cases the chills can be used more than once. External chills have the disadvantage that, for some applications, they must be cast, or machined, to fit the surface they are to chill, and of course, have the same difficulty as internal chills in that the choice of the proper size is most important.

112. The effect of external chills of various sizes has been studied qualitatively in a few of the adjoining sections mentioned previously. It should be noted again that these sections have not been fed except through the arms and that, therefore, they are typical only of the inaccessible junctions found in actual castings. It will be noted that in the case of the "T"-section with unusually large radii, some of the cylindrical chills employed are not those that might be used in actual foundry practice. They were included in the study in order that a comparison of the results might be made with those obtained with the "T"-section having smaller radii. The other unusual shapes were used so that the effect of the amount of surface in contact with the casting might be noted.

113. The cross-sectional area of the chill, the perimeter of the chill in contact with the casting (noted as the "effective perimeter"), and the cross-sectional area of the defect are given in each case. This method of illustration was chosen as it was deemed to be the easiest manner to visualize the effect of the chill upon the sections.

CHILLED "L"-SECTIONS

114. The effect of chills upon the defects found in unfed "L"-sections is shown in Fig. 43. A ½-in. diameter external chill placed at the radius of the inner corner reduces the size of the defect considerably. A chill 1-in. in diameter, or a small triangular

chill with about the same cross-sectional area, produced a sound section.

CHILLED "T"-SECTIONS

115. It is difficult to produce sound "T"-sections by the use of external chills alone. Eight different sizes of chills were used on "T"-section 202 and while each type produced a smaller defect than that present in the original casting, only one of the modifications produced a sound casting. The types of chills used are shown

(102)—Standard section.
AREA OF DEFECT—2.8 sq. in.

(102-1)—½ in. diameter chill.
Cross-sectional area—0.20 sq. in.
Effective perimeter—0.25 in.
AREA OF DEFECT—0.1 sq. in.

(102-2)—One inch diameter chill.
Cross-sectional area—0.78 sq. in.
Effective perimeter—0.78 in.
No Defect.

(102-3)—Small triangular chill.
Cross-sectional area—0.70 sq. in.
Effective perimeter—1.87 in.
No Defect.
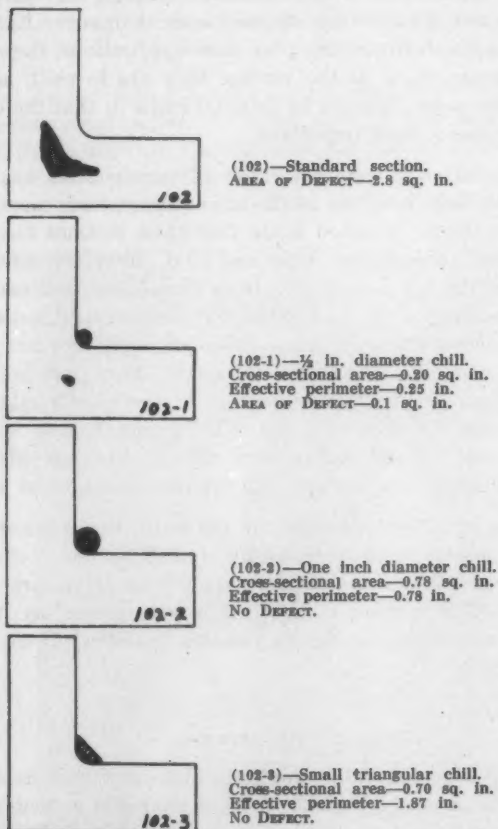
FIG. 43—CHILLED L SECTIONS—102, 102-1, 102-2, 102-3.

(202)—Standard section.
AREA OF DEFECT—3.6 sq. in.
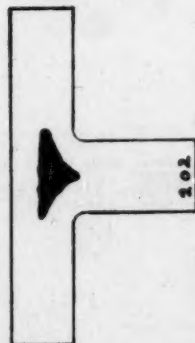
(202-1)—½ in. diameter
chills.
Total cross-sectional area—
0.40 sq. in.
Total effective perimeter—
0.50 in.
AREA OF DEFECT—2.1 sq. in.

(202-2)—1 inch diameter
chills.
Total cross-sectional area—
1.56 sq. in.
Total effective perimeter—
1.56 in.
AREA OF DEFECT—0.8 sq. in.
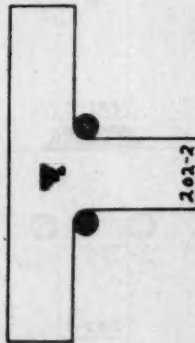
FIG. 44—CHILLED T SECTIONS—202, 202-1, 202-2.

(202-3)—1½ in. diameter
chills.
Total cross-sectional area—
3.54 sq. in.
Total effective perimeter—
2.36 in.
AREA OF DEFECT—0.3 sq. in.

(202-4)—2-inch diameter
chills.
Total cross-sectional area—
6.28 sq. in.
Total effective perimeter—
3.14 in.
AREA OF DEFECT—0.4 sq. in.

(202-5)—Small triangular
chills.
Total cross-sectional area—
0.70 sq. in.
Total effective perimeter—
1.87 in.
AREA OF DEFECT—0.9 sq. in.

FIG. 45—CHILLED T SECTIONS—202-3, 202-4, 202-5.

in Figs. 44, 45, 46, 47 and 48. Chills ½, 1, 1½ and 2-in. in diameter, respectively, were placed at the radii of the section. As the size of the chill at the radius was increased, the size of the defect decreased until a critical size of chill was reached.

116. The increase in chill diameter from 1½ to 2-in. also increased the size of the radius and this apparently counteracted the effect of the larger chill, as the size of the defect obtained is approximately the same in each case.

117. Two triangular chills were also used. The smaller of the two had a volume approximately equal to that of the 1-in. diameter chill with, of course, a much larger area in contact with the casting. This increase of the area in contact with the casting did not affect the size of the defect, as the result obtained with the 1-in. diameter is approximately the same as that obtained when the triangular chill is used.

(202-6)—Large triangular chill.
Total cross-sectional area—5.88 sq. in.
Total effective perimeter—7.6 in.
AREA OF DEFECT—0.5 sq. in.

(202-7)—1-inch plate chill (4 in. to 2½ in.)
Cross-sectional area—3.75 sq. in.
Effective perimeter—4.0 in.
AREA OF DEFECT—0.6 sq. in.

(202-8)—1-inch diameter chills plus 1 in. plate chill.
Total cross-sectional area—5.31 sq. in.
Total effective perimeter—5.56 in.
No DEFECT.
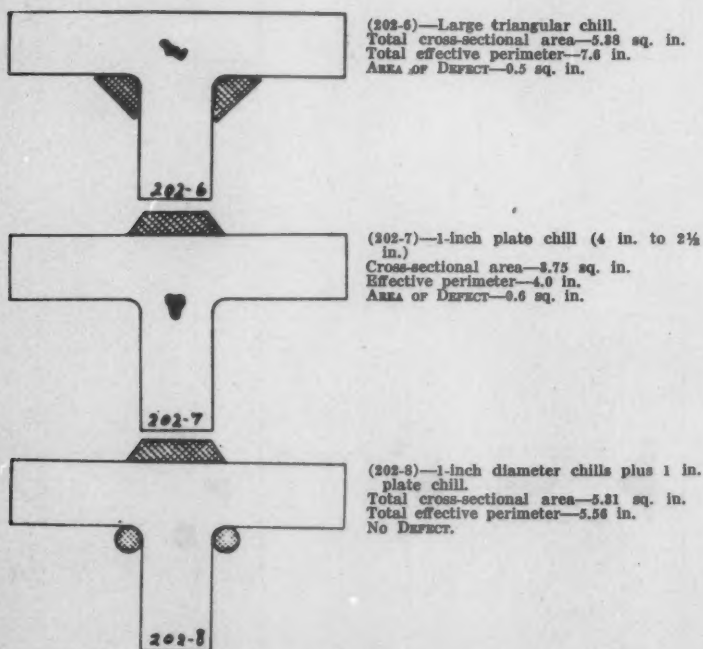
FIG. 46—CHILLED T SECTIONS—202-6, 202-7, 202-8.

118. It has been noted that chills placed at the radii of a "T"-section tend to move the defect toward the top of the "T." A plate chill placed across the top of the "T" has the opposite effect. This type of chill will decrease the size of the defect and move the defect down into the leg of the "T" (202-7, Fig. 46). If a combination of these two types of chills, *i. e.*, 1-in. diameter chills at the radii plus a plate across the top, is used, it is possible to produce a sound section (202-8).

119. If the leg of the "T" is smaller than the top of the

(210-1)—1-inch plate chill (4 in. to 2½ in.)
Cross-sectional area—5.75 sq. in.
Effective perimeter—4.0 in.
No DEFECT.

(204)—Standard section.
AREA OF DEFECT—4.1 sq. in.

(204-1)—1-inch diameter chills.
Total cross-sectional area—1.56 sq. in.
Total effective perimeter—0.25 in.
AREA OF DEFECT—3.9 sq. in.

(204-2)—1½ in. diameter chills.
Total cross-sectional area—3.54 sq. in.
Total effective perimeter—0.38 in.
AREA OF DEFECT—2.7 sq. in.

FIG. 47—CHILLED T SECTIONS—210-1, 204, 204-1, 204-2.

"T," a plate chill across the top is in many cases all that is required to produce a sound junction (210-1, Fig. 47).

120. In "T"-sections, excessive fillets are a liability to the foundryman, as it is very difficult to produce a sound section. This is illustrated by a series of studies with chills of various sizes with pattern 204. As in the "T"-section previously mentioned, an increase in the size of the chill tends to reduce the size of the defect. The size of the defect does not decrease as rapidly as in the previous case. Even a chill 4-in. in diameter does not reduce

(204-4)—3-inch diameter chills.
Total cross-sectional area—7.07 sq. in.
Total effective perimeter—1.0 in.
AREA OF DEFECT—1.1 sq. in.

(204-3)—2-inch diameter chills.
Total cross-sectional area—6.28 sq. in.
Total effective perimeter—0.50 in.
AREA OF DEFECT—1.3 sq. in.

(204-5)—Semi-circular chill (3 in. radius).
Total cross-sectional area—28.27 sq. in.
Total effective perimeter—9.42 in.
AREA OF DEFECT—4.1 sq. in.

(204-6)—3-inch diameter chills plus 1-inch plate chill.
Total cross-sectional area—10.82 sq. in.
Total effective perimeter—5.0 in.
AREA OF DEFECT—0.20 sq. in.

FIG. 48—CHILLED T SECTIONS—204-3, 204-4, 204-5, 204-6.

the defect to a size comparable to that obtained with a chill 2-in. in diameter in the "T"-section previously discussed.

121. When a very large chill that fits the entire radius is used (204-5), the defect, while of a different shape, is as large as that found in the casting made without chills. It was suggested by a foundryman who visited this laboratory that the defect shown in this section was much larger than one could expect and he mentioned the possibility that radiography had introduced considerable error. This casting was, therefore, sectioned on the center line and the defect revealed was compared with the original radiograph. The outlines of the defect on the radiograph and on the casting coincided perfectly. Calculations on the amount of liquid steel involved in producing a defect of this size have indicated that

(502)—Standard section.
Area of Defect—6.4 sq. in.

(502-1)—½-inch diameter chill.
Total cross-sectional area—0.80 sq. in.
Total effective perimeter—1.00 in.
Area of Defect—2.1 sq. in.

(502-2)—1-inch diameter chills.
Total cross-sectional area—3.12 sq. in.
Total effective perimeter—3.12 in.
Area of Defect—0.60 sq. in.
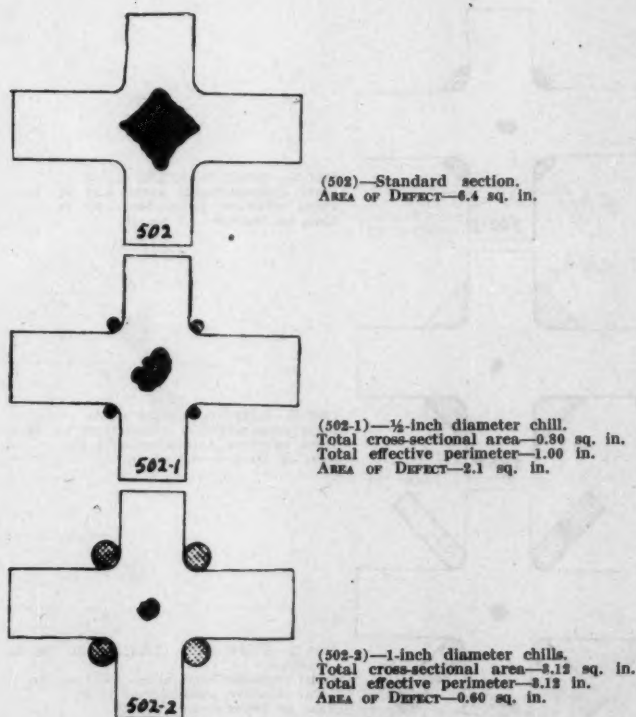
Fig. 49—Chilled X Sections—502, 502-1, 502-2.

with such a large chill it is theoretically possible to have a defect even larger than that found.

122. The minimum defect in this type of "T"-section was obtained when 4-in. diameter chills were used at the radii and a plate chill was placed across the top of the "T."

CHILLED "X" SECTIONS

123. The "X"-sections (Figs. 49, 50 and 51) were the most difficult sections to chill properly. In fact, it was not possible to produce a sound "X"-section by chilling alone. An increase in the size of the chill again produced a smaller defect. The best results were obtained with the large triangular chill (502-4). An increase
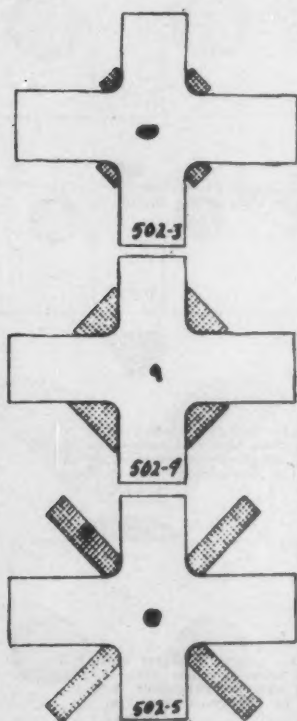
(502-3)—Small triangular chills.
Total cross-sectional area—2.80 sq. in.
Total effective perimeter—7.43 in.
AREA OF DEFECT—0.7 sq. in.

(502-4)—Large triangular chills.
Total cross-sectional area—10.76 sq. in.
Total effective perimeter—15.2 in.
AREA OF DEFECT—0.2 sq. in.

(505-5)—Plate chills (1 in. x 4 in.—½ in. radius).
Total cross-sectional area—15.57 sq. in.
Total effective perimeter—3.12 in.
AREA OF DEFECT—0.5 sq. in.

FIG. 50—CHILLED X SECTIONS—502-3, 502-4, 502-5.

in the radii at the corners of the casting, as in the case of the "T"-section (202-4), produced a larger, rather than a smaller defect (503-1). In this case, it appears that the ½-in. radius is preferable to the 1-in. radius.

124. The staggered "X"-section, unfed and unchilled, presents an elongated defect. Chilling this section with 1-in. diameter bars divides the defect into two smaller defects. Using chills larger in diameter increases, rather than decreases, the size of the defects produced. It would seem, therefore, as suggested in the section on design, that staggered section with some distance between the arms would be preferred (see 509, 510, 511, 512 of Figs. 22 and 23). These sections could be chilled as "T"-sections and produced without defects if feeding was not possible.

(503-1)—2-inch diameter chills.
Total cross-sectional area—12.56 sq. in.
Total effective perimeter—6.28 in.
AREA OF DEFECT—0.9 sq. in.

(506)—Standard section
AREA OF DEFECT—5.0 sq. in.

(506-1)—1-inch diameter chills.
Total cross-sectional area—3.12 sq. in.
Total effective perimeter—3.12 in.
AREA OF DEFECT—1.2 sq. in.
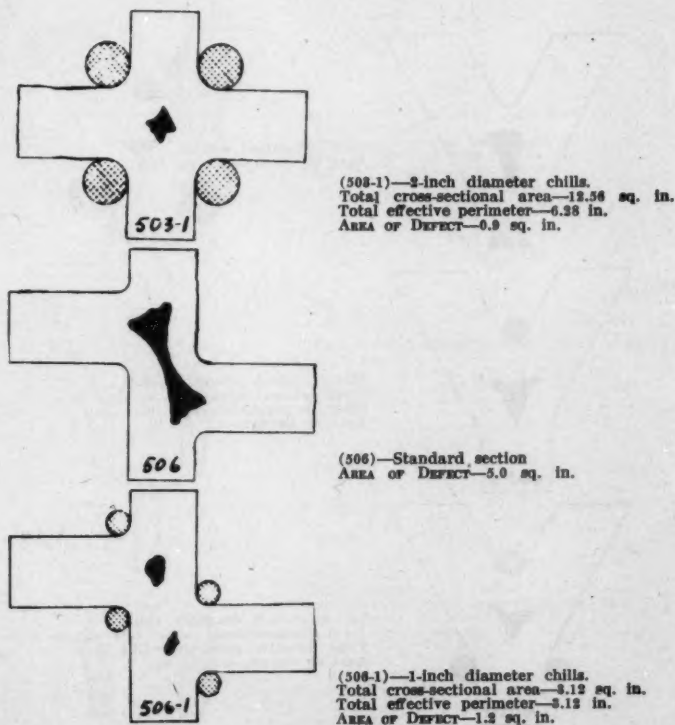
FIG. 51—CHILLED X SECTIONS—503-1, 506, 506-1.

CHILLED "Y" SECTIONS

125.   Sections in the form of a "Y" are not encountered as frequently as some of the other types. They are, like the "X"-sections, very difficult to produce without defects unless they can be fed directly. The results of various chilling methods are shown as Fig. 52 and 53. The best result obtained is indicated in casting 402-5.

126.   A summary of the effect of external chills upon the sections studied is presented in Table 7. It is interesting to note that in the castings studied the diameter of the chill was very important when an increase in this diameter meant an increase in the radius at the corner of the casting. For instance, in the "T"-section 202, a radius of approximately ¾-in. appears to be the

(402)—Standard section.
AREA OF DEFECT—2.7 sq. in.

(402-1)—1-inch diameter chill.
Cross-sectional area—0.78 sq. in.
Effective perimeter—0.78 in.
AREA OF DEFECT—2.1 sq. in.

(402-2)—1-inch diameter chills.
Total cross-sectional area—2.34 sq. in.
Total effective perimeter—2.34 in.
AREA OF DEFECT—1.6 sq. in.
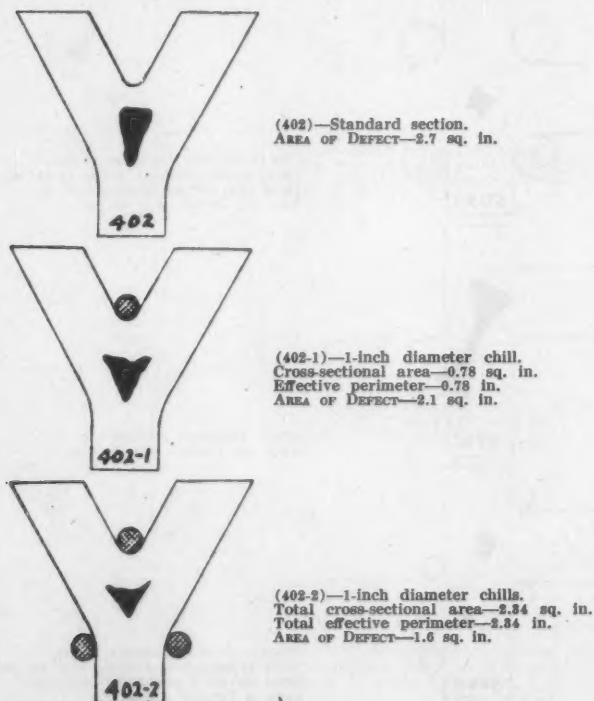
FIG. 52—CHILLED Y SECTIONS—402, 402-1, 402-2.

critical radius. An increase from a 1-in. diameter chill ($\frac{1}{2}$-in. radius) to a $1\frac{1}{2}$-in. diameter chill ($\frac{3}{4}$-in. radius)decreased the cross-sectional area of the defect from 0.8 to 0.3 square in. If, however, the diameter of the chill is increased to 2-in. (1-in. radius) the cross-sectional area of the defect becomes 0.4 sq. in. The same thing may be noted with the "X"-section. In this case, an increase in the diameter of the chill from 1-in. to 2-in. increases the cross-sectional area of the defect from 0.6 to 0.9 sq. in.

127. Another interesting fact is that the location of the chill is as important as the size of the chill. For instance, casting 202-6 (Fig. 46) is chilled with two triangular chills having a total cross-sectional area of 5.38 sq. in. and a total effective perimeter of 7.6 in. The defect obtained has a cross-sectional area of 0.5 sq. in. Casting 202-8, however, is sound, and total cross-sectional area and

(402-3)—1-inch diameter chill plus 2-inch diameter chills.
Total cross-sectional area—7.06 sq. in.
Total effective perimeter—3.92 in.
Area of Defect—1.0 sq. in.
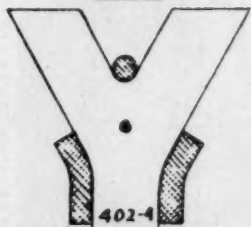
(402-4)—1-inch diameter chill plus 1-inch chill plates.
Total cross-sectional area—8.78 sq. in.
Total effective perimeter—9.05 in.
Area of Defect—0.2 sq. in.
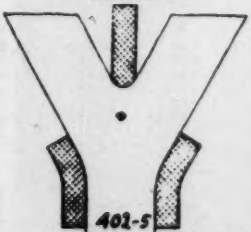
(402-5)—1-inch chill plates.
Total cross-sectional area—11.89 sq. in.
Total effective perimeter—9.05 in.
Area of Defect—0.1 sq. in.

Fig. 53—Chilled Y Sections—402-3, 402-4, 402-5.

effective perimeter of the chills are 5.31 sq. in. and 5.56 sq. in., respectively.

128. The effective perimeter and the cross-sectional area of the chill both govern the chilling effect of a particular chill, but the latter is the more important. The heat conductivity of steel is great enough so that the cross-sectional area must be quite large in proportion to the effective perimeter before the efficiency of the chill is impaired. This is illustrated by results obtained on castings 202-2, 202-5, 502-2, 502-3, 502-5, 402-4 and 402-5 (Figs. 44, 45, 49, 50 and 53). It will be noted that in each case where the 1-in. diameter chill and the small triangular chill are used, the size of the defect is approximately the same even though the effective perimeter of the triangular chill is over twice that of the 1-in. diameter chill. However, when the plate chill with a ½-in. radius is substituted for the 1-in. diameter chill, the cross-sectional area

Table 7
TABULATION OF EXTERNAL CHILLS

Casting Number	Corner Radius (in In.)	Types of Chills Used	Effective Perimeter (In.)	Total Cross-Sectional Area (Sq. In.)	Area of Defect (Sq. In.)
"L" Sections					
102	0.50	None	-----	-----	2.8
102-1	0.50	½-in. dia.	0.25	0.20	0.1
102-2	0.50	1-in. dia.	0.78	0.78	None
102-3	0.50	Small triangle	1.87	0.70	None
"T" Sections					
202	0.50	None	-----	-----	3.6
202-1	0.50	½-in. dia.	0.50	0.40	2.1
202-2	0.50	1-in. dia.	1.56	1.56	0.8
202-3	0.75	1½-in. dia.	2.36	3.54	0.3
202-4	1.00	2-in. dia.	3.14	6.28	0.4
202-5	0.50	Small triangles	3.74	1.40	0.9
202-6	0.50	Large triangles	7.60	5.38	0.5
202-7	0.50	1-in. plate	4.00	3.75	0.6
202-8	0.50	1-in. dia. plus 1-in. plate	5.56	5.31	None
204	3.00	None	-----	-----	4.1
204-1	3.00	1-in. dia.	0.25	1.56	3.0
204-2	3.00	1½ in. dia.	0.38	3.54	2.7
204-3	3.00	2-in. dia.	0.50	6.28	1.3
204-4	3.00	3-in. dia.	1.0	7.07	1.1
204-5	3.00	Semi-circular (3-in. radius)	9.42	28.27	4.1
204-6	3.00	3-in. dia. plus plate	5.00	10.82	0.2
"X" Sections					
502	0.50	None	-----	-----	6.4
502-1	0.50	½-in. dia.	1.00	0.80	2.1
502-2	0.50	1-in. dia.	3.12	3.12	0.6
502-3	0.50	Small triangles	7.48	2.80	0.7
502-4	0.50	Large triangles	15.2	10.76	0.2
502-5	0.50	1-in. plates	3.12	15.57	0.5
503-1	1.00	2-in. dia.	6.28	12.56	0.9
506	0.50	None	-----	-----	5.0
506-1	0.50	1-in. dia.	3.12	3.12	1.2
"Y" Sections					
402	-----	None	-----	-----	2.7
402-1	-----	1 1-in. dia.	0.78	0.78	2.1
402-2	-----	3 1-in. dia.	2.34	2.34	1.5
402-3	-----	1-in. dia. plus 2-in. dia.	3.92	7.06	1.0
402-4	-----	1-in. dia. plus chill plates	9.05	8.78	0.2
402-5	-----	1-in. plates	9.05	11.80	0.1

is increased considerably while the effective perimeter remains the same. The change in the size of the defect is very slight, indicating that plate chills of this type are very inefficient and not a great deal better than a 1-in. diameter bar.

129. Internal chills were used in an attempt to produce sound sections in cases where external chills could not do so. These attempts were not satisfactory, as in most cases the location of the chill was shown clearly on the radiograph by lack of fusion, dirt or gas around the chill. Typical radiographs showing the type of defects encountered are shown as Figs. 54 and 55. It should be noted that the type of defect is the same whether one coil chill or four are used, so that "over-chilling" cannot be responsible. A longitudinal view of two coil chills is shown as Fig. 56. In this case, practically every turn of the coils may be seen. The use of solid internal chills presented the same difficulties and it appeared that, for the same volume of chilling material, as good results could be obtained with solid chills as with the coil variety.

Fig. 54—Radiograph of Y Section with Internal Chill.

FIG. 55—RADIOGRAPH OF Y SECTION WITH INTERNAL CHILLS.

FIG. 56—RADIOGRAPH OF A BOSS WITH INTERNAL CHILLS.

130.   It is the opinion of the authors that if chills, similar to those mentioned above, are used on sections of less than 3-in. thickness, the results would be similar. The effect of these chills upon defects present, of course, will increase as the size of the section decreases.

Summary of Part II

131.   The first requirement in the production of a sound casting is a design to which the principles of "controlled directional solidification" may be applied. If this is not possible with the design as originally produced, a few slight changes may be required. Models of the original design at a suitable scale are very helpful to both the designer and foundryman, as the changes required can then be pointed out. These models are sometimes a distinct aid in convincing the designer or consumer that the alterations desired are reasonable and not injurious to the proper use of the casting.

132.   After the foundryman has obtained a workable design, he can use the tools at his command, such as tapered sections, proper gating and heading, chills, relieving blocks, etc., to preclude the possibility of shrinkage cavities and hot tears. These foundry arts have been used to advantage on poorly designed castings, but they cannot be expected to cure all the evils found in designs produced with little or no thought to foundry problems.

Acknowledgment

BIBLIOGRAPHY

1. Briggs, C. W. and Gezelius, R. A., TRANS. A.F.A., vol. 42 (1935), pp. 449-470.

2. Briggs, C. W. and Gezelius, R. A., TRANS. A.F.A., vol. 44 (1936), pp. 1-24.

3. Briggs, C. W., TRANS. A.F.A., vol. 45 (1937), pp. 61-81,

4. White, Clark, and Crocker, Am. Soc. of Mechanical Engineers, vol. 58, November 1936.

5. Lorig and Williams, American Foundrymen's Association— American Society for Testing Materials, "*Symposium on Steel Castings*," 1932.

6. Batty, G., Trans. A.F.A. vol. 42 (1934), pp. 237-258.

7. Heuvers, A., "STAHL UND EISEN," August 29, 1929, p. 1249.

DISCUSSION

Presiding: F. A. MELMOTH, Detroit Steel Casting Co., Detroit, Mich.

CHAIRMAN MELMOTH: This paper is going to require a great deal of quiet study and absorption. Design is one of the most important subjects we have at the present moment. If we have anything go wrong in the foundry, we generally kick about design, often quite justly so.

H. D. PHILLIPS[1] and E. C. TROY[2] *(Written discussion):* It is difficult to discuss a paper of this sort, containing as it does many scientifically ascertained facts, about which there can be little or no question.

In the first section dealing with steel casting design, it is pleasing to note that the rules laid down by the authors check so closely, in a general way, with those long advocated by foundrymen as a result of empirical observations. Rules 5 and 7, on page 643, however, do suggest a sound departure from previously recommended practices in the designing of L, V, T and X shapes.

Rule 8, on the same page, is especially helpful, coming as it does from an unbiased source. We, as foundrymen, are very grateful for this statement in a paper which shall find its way into the hands of designing engineers. We all have had knowledge of previous tests of this sort having been made, but the information secured has not been made available, and as a result there has been left in the minds of many, both producers and consumers, a feeling that possibly the results secured were not so good.

This investigation of the effect of mass on the properties of steel castings was well done, and is of considerable value.

In paragraph 70, the term "aluminum network" might be questioned. It is assumed that the authors refer to the eutectic type of inclusion encountered in some steels treated with a critical amount of aluminum.

[1, 2] Superintendent and Metallurgist, respectively, Dodge Steel Co., Philadelphia, Pa.

It is of interest to note that the inclusion form found in the centers of the heavier sections check the findings of others, and it is quite probable that the aluminum additions made to the basic electric steels investigated resulted in the formation of what Sims classes as number III type inclusions.

In attempting to study Table 4, it would have helped considerably had the authors included a sketch showing the location of tests C-4 to F-5.

A study of the test results secured indicates a low average Izod impact value for the medium manganese steel. This probably is due to the heat treatment given and, no doubt, could be improved by a normalize and temper. In further reference to heat treatment, the authors report that one coupon of each size was annealed at 900°C. for the customary time of one hour per inch of section. It is questionable whether or not this is the customary practice, it being the opinion of many that it takes no longer for diffusion to take place in the center of an 8 in. section than it does in the center of a 4 in. or 5 in. section, once they are at a uniform heat, or in soak, and recrystallization, of course, is completed as soon as the critical temperature is passed.

In the testing of bars cast to size, as the authors have pointed out, the surface of the bar and size of the axial weakness have a very potent influence on the ductility results secured and it would be interesting to know more of the procedure followed by the authors in molding, casting and cleaning these bars. It has been found that the method of blast cleaning, whether steel shot, grit or sand is used, has some influence on the results.

The chilling of steel castings by means of internal chills is a practice that might rightly be frowned upon, but it seems to be a necessary evil in the production of intricate, poorly designed castings.

In paragraph 129, the authors state that it appears that for the same volume of chilling material as good results could be obtained with solid chills as with the coil variety. Does it not appear reasonable though, that in the case of the coil chill, with its better distribution of chilling material, that the possibility of large internal stresses, as a result of the chilling action, is lessened?

The caption of Table 7 is "Tabulation of Internal Chills." Should this not be "Tabulation of External Chills?"

This paper is another excellent exposition of the advantages of radiography in the foundry, when it is used in a capable, intelligent manner.

In conclusion, we wish to congratulate and thank the authors for this paper of great value to the designer and producer of steel castings. Its value to the foundryman is enhanced by its clear style of presentation and it merits study by the most experienced producer and consumer of castings.

G. C. RIEGEL[3] *(Written discussion):* It has been the practice of our firm for many years to write into our steel casting specifications the following expression: "Vendor shall exercise care to avoid use of chaplets or internal chills." We have found very good confirmation of the practice recommended in this paper, that the use of nails, chaplets, and internal chills can only be expected to produce a structural quality like a "crutch to a cripple" on which the maximum dependability cannot be obtained in service. As a consequence, after a few initial castings have been made from certain designs, the sections most likely to show internal shrink are examined and re-designed, if possible, to promote soundness by a change of section where permissible, or the application of external chills, if otherwise unavoidable.

It is also quite possible in many instances, to change the location of the heading, or by the addition of more heads, overcome the natural shrinkage at the change of section. In one particular instance, we have cast an idler roller with a continuous head about the circumference of periphery which is removed by sawing in special equipment devised by the foundry. This circumferential head has practically eliminated the unsoundness at the junction of the rim and arms of the idler roller. It must always be borne in mind that a shrink cavity is simply a "yawning" expression of the solidifying metal for hot fluid metal which was not supplied.

C. E. SIMS[4] *(Written discussion):* The authors have tackled an extremely difficult and complex subject and are to be highly commended, not only for the great amount of painstaking work they have done, but for their courage in sticking to fundamentals.

It is, of course, fairly simple to design a casting that can be made sound if there are no other specifications, but there are often apparently irreconcilable conflicts between design of a casting that will be sound and one that will fit the purpose for which it is desired. Much can be done for improvement, however, and the examples given and the rules formulated, furnish considerable ammunition to the designer which he should use not dogmatically, but with discretion.

There is an old axiom to the effect that the soundness of a rule is not necessarily discredited by exceptions. In this respect, it will be noted that the first rule on page 608 is apparently contradicted by paragraph 48 and by the first rule in paragraph 57.

Uniform sections are advocated for the avoidance of hot spots, but perhaps it should be further emphasized that they furnish no effective remedy without the intelligent cooperation of the pattern maker and molder. For example, a long casting of perfectly uniform section may have a fatal hot spot if gated in the center.

The authors conducted their experimental work on 3 in. sections for very good reasons which they list. It should be pointed out, however, that different sized sections have different casting properties. This is

especially true in regard to corner hot cracks which are notably more prevalent in sections under 1 in. thick than in 3 in. sections.

Little mention is made of corner hot tears yet it is reasonably certain that many of the test castings would have had them had the sections been about ½ in. instead of 3 in. The reason for this difference, due to section, seems to be that the initial chilling action of the sand extracts a greater portion of the total heat of a lighter section and thereby produces greater differentials in temperature.

Contrary to a prevalent misconception, the corner hot crack or tear is not due to any restriction to movement from the sand. It is truly a shrinkage crack and cannot be cured by any method of mold relieving. In a simple angle casting, for instance, as in Fig. 10, No. 102, cracks will occur as readily with short legs as with long.

The reason is that the outer surface chills quickly, because of the large volume of sand to absorb the heat. The inner surface, especially right at the corner, on the other hand, has no such means of escape for the heat. This portion freezes and shrinks after the outside has become solid and rigid. The only way in which this shrinkage could be relieved is for the legs of the angle to converge to less than 90°.

If feeding by means of a riser is possible, these corner hot cracks can be prevented, but the only practicable preventative is an external chill, which corrects the cause.

There is a question about the sine wave construction of Fig. 6, and its ability to prevent tearing by relieving stresses. The authors have shown in previous reports that hot cracks occur above 2300° F. when the metal has nearly zero strength. This being so, it is possible that there would be enough resistance without tearing to straighten out the wave in a thinner and cooler section. The construction, moreover, creates an acute and an obtuse angle at the junction with the heavy section. The acute angle would be more susceptible to tearing than would a right angle, partly because of greater tensile stress and partly because of impaired cooling.

The authors show the apparent futility of attempting to eliminate completely shrinkage cavities in certain designs by re-designing, or by the use of either external or internal chills. Internal chills reduce the size of the cavities, but leave defects. In the case of a boss which is to be drilled and tapped, for instance, an internal chill may perform a very satisfactory service, even though some slight imperfection remains.

In the case of pressure castings, surface cracks may be a more potent cause of leakage than a well buried shrinkage cavity. External chills can be made to give surface soundness, by helping to equalize the cooling rate, and thus prevent leaking.

Shrinkage cavities at intersections may be so well buried as not to impair the structural integrity or strength of a casting in the slightest. There is a serious question whether to attempt the elimination of such cavities or to recognize their inevitability and their impotence.

It is hoped that this paper will stimulate thought and bring forth more data and other viewpoints and, last but not least, that the authors will continue their good work.

L. E. EVERETT[5] *(Written discussion):* This paper, in common with others by the same authors, is extremely interesting and constitutes a very valuable contribution, of tangible, factual data on steel casting design. In commenting on the paper, the writer can offer no experimental data to back up his comments but offers them more as observations based on several years of experience in trying to control practically the problems more scientifically and experimentally analyzed by these authors.

In addition to the section junctions listed by the authors, one of the most troublesome junction problems the steel foundry man has to face is that of a junction of vertical members such as L, T or X, supported on or covered by a plate section. As frequently designed, such section junctions offer problems in solidification almost impossible of solution without recourse to both internal and external chills.

In considering the authors choice of a 3 in. x 3 in. section upon which to base general rules of design, and considering the section two feet long from gate and head to the junction, it occurs to the writer that the defects found, and some of the results secured by changes in radii and later on by chill sizes, may be characteristic of the section size, (center line weakness) rather than of details of design. Except in the flanges and seats of pressure fittings, three inch section junctions are unusual in the ordinary run of electric furnace steel castings with which the writer is familiar, and the junction of three inch sections, unfed, is extremely unusual.

Referring to Fig. 8, the question arises in the writer's mind as to why the sound metal is so much thicker along the two inner radii of "T" section, than it is along the back. Experience indicates the back section should be at least as heavy as the sections along the radii.

In paragraph 47, the authors recommend ½ to 1 in. radius for the inside of three inch thick "L" sections. It is the writer's opinion that the radius to be used in any internal corner should never be less than one-half the section thickness, and 75 to 100 per cent of the section thickness is preferable. From a foundry standpoint, I believe their section 106 P-15 is much to be preferred to section 104, and would cause less stress concentration and susceptibility to hot tear formation in casting.

As mentioned in paragraph 48, the practice of designing an unfed junction slightly smaller than the connecting arms may be sound if the designer is certain how the casting is to be made. It is suggested that caution be exercised by designers in attempting to do this, except with the advice of the producing foundry, as conditions may be set up that would interfere seriously with proper directional solidification.

[5] Foundry Superintendent, Key Company, East St. Louis, Ill.

In paragraph 51 and Figs. 13 to 16, inclusive, and 44 to 48, inclusive, the question arises as to how these specimens were headed, gated and poured. In considering the question of coring out the hot spot in a "T" section, it occurs to the writer that if the metal left around the core is sufficient for the design strength, the three-inch section is unnecessary. Furthermore, a two inch diameter core may be satisfactory to use in a section three inches long—but if the section were to be increased to even six inches, let alone twelve inches or more, the securing of the core, and especially removing it from the casting, would be a mean problem. Particularly so if metal casing the core would not be permitted on account of the chilling effect of the casing.

In paragraph 52, the authors recommend not less than one inch radius for the inner radius for "V" sections, but it is believed that a radius at least 50 per cent, and preferably 75 to 100 per cent, of the section thickness would be preferable. In the manufacture of pressure fittings, it is desirable to keep such a radius at least 100 per cent of the section thickness.

In regard to the rule promulgated on page 643, regarding the effect of mass on physical properties, it is desired to call the authors' attention to Table 5, paragraph 9-F, of the Gulf Refining Company specification P-115 for chrome molybdenum steel castings, shown here as Table 8. This table was formulated after considerable experimental work to determine what could reasonably be expected of test bars from various section thicknesses. It is to be noted that the values given are for bars cut from blocks of the size noted—after heat treatment.

Table 8

MINIMUM LIMITS FOR PHYSICAL PROPERTIES OF TEST SPECIMENS
PREPARED FROM TEST BLOCKS OF SIZE INDICATED

	Range of All Blocks			Bars Cut From Castings
Test Block Section Thickness (in.)	1 to 1½	2 to 2½	3 to 3½	
Yield Point (Lb. per sq. in.)	80,000	75,000	70,000	75,000
Tensile Strength (Lb. per sq. in.)	110,000	100,000	90,000	100,000
Elongation (Per cent in 2 in.)	16.5	15.0	14.0	11.0
Reduction of Area (Per cent)	30.0	28.0	25.0	18.0
Brinell Hardness no.	180	to	240	

The above applies to castings made from steel containing approximately 6 per cent chromium, 0.50 per cent molybdenum and 0.25 carbon. The castings are valves, fittings and other castings for use in petroleum refining operations under corrosive conditions at elevated temperatures and high pressure.

Unless, as a result of careful experimental work, the Gulf Refining Company was convinced that the effect of mass did have a deleterious effect on the physical properties of test bars they would never have written the above into their specifications. As an indication of the comparison between test bar values, and the quality of metal in the castings the last column is included above to show the minimum values that test bar cut from any section of the casting must conform to. It may be also

said that the above specification has been in use since 1933 and castings are made regularly conforming thereto.

The writer is not qualified to discuss the use of internal chills, as they are not allowed to be used in any of the pressure work that constitutes our production, and while more expensive we feel that we can secure more consistent and reliable results with external chills.

Discussion of the results secured by the authors with external chills is somewhat difficult as we have no information as to the length of the chills used—consequently no volume figures are available. It is the writer's opinion that volume of a chill is of prime importance, and it frequently is necessary to know the actual volume of the defect, as well as its largest area, in order to treat it successfully.

For example, a section three or four inches deep that cannot be satisfactorily solidified by the application of external chills to its vertical faces only, such as the "X" or "Y" sections illustrated, can frequently be handled by a chill on the bottom or top surface alone or in combination with the side chills. This point is brought out here to emphasize that the results the authors have described hold true for the limitations of the sections and conditions they have selected—only, they have by no means exhausted the possibilities of external chill application to section junctures.

In regard to chill length or volumes, it is generally good practice to have the chills on straight surfaces extend beyond the casting line, at least one diameter. This prevents fins over the ends of the chills, and makes their removal, (if stuck) much easier. It also adds considerable heat capacity and volume to the chill.

The application of the types of external chills applied permits of very little discussion, in addition to that offered by the authors. The illustrations probably do not portray the actual conditions found, but as shown do not, to the writer's mind, illustrate good modern chill practices.

It is not considered good practice to have any re-entrant angles of sand between the chill and the casting. The plate chills used on the back of the "T" represents good chill practice, wherein the end of the chill as it leaves the casting is thinned and tapers toward the body of the chill, preventing a sudden line of demarcation between the casting surface exposed to the full thickness of the chill and that exposed to the sand. The use of round chills of the same radius as the radius on the casting is just fair. There is a small section of sand left between the chill tangent point and the casting that may be heated by both casting and chill. This sand, of very small volume and low heat capacity may get so hot so quickly, it will tend to heat the chill, thus destroying its effectiveness—at least partially, and also may create a hot zone on the surface of the casting. The use of small diameter round chills on large radii is not effective by the same reasoning—augmented. (Fig. 57 A and B).

The ineffectiveness of the large plate chills, 502-5, is probably
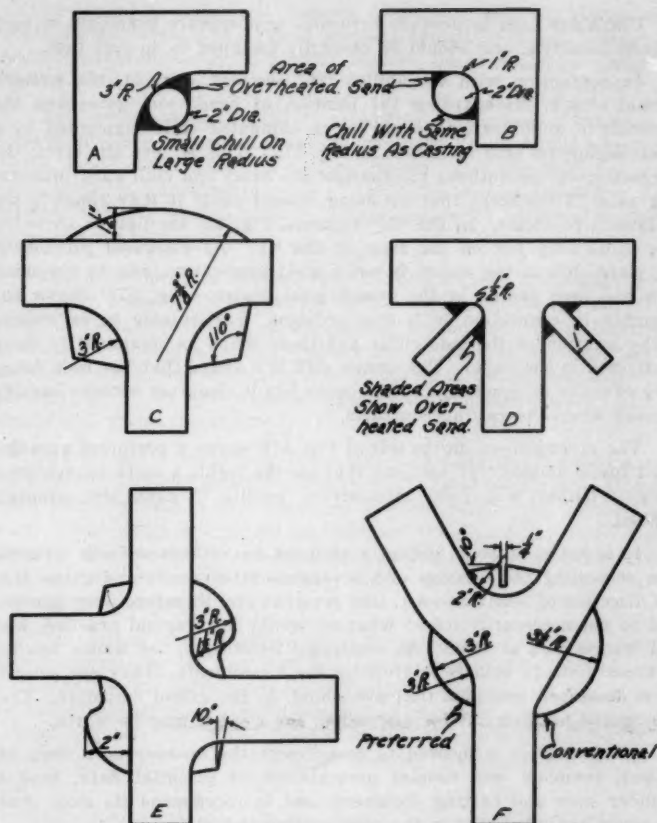
FIG. 57—SUGGESTED CHILLING PRACTICE FOR L, T, X, AND Y SECTIONS.

largely due to the small sections of sand left in the included angle between the chill and the casting. (See Fig. 57D of this discussion).

Trouble can frequently arise due to over-chilling, and referring to section 204-5, Fig. 48, it may be possible that the defect shown there, 4.1 sq. in. (which is the same size as that shown in the standard section —204 made without any chills) was caused by the large chills setting the legs of the "T" before any feeding could take place. It is the writer's opinion that chills of the same radius (3 in.) with a contact arc of 90°, but the included angle between the "radial" edges of about 110°, would represent a more practical attempt to solve the solidification problem. (See Fig. 57C). The same general design is believed to be preferable to and much more efficacious than round chills, in the other sizes as well.

Chills are used to produce favorable temperature gradients in very definite localities, and should be carefully designed to do just that.

In connection with the chilled "X" sections, Fig. 51, the writer's normal attack, disregarding the location of heads and gates and the direction of solidification, would be an elongated chill augmented by a small segmental chill as shown in Fig. 57E. This is very similar to the suggestion of the authors to separate the arms and chill each intersection as a "T"—except that by using special chills it may simplify the designer's problems. In the "Y" sections, Fig. 53, the figures show the side chills very low on the stem of the "Y." As discussed previously, the plate chill in the crotch is not a good application, nor to our mind is a one inch radius in the crotch good design. Fig. 57F shows our thoughts in connection with this problem. Particularly to be noticed is the location of the side chills and their shape as compared to those illustrated in the paper. The crotch chill is a shape that has been found very effective in practice, and the notch lets it compress without causing a crack when the casting contracts.

The arrangement on the left of Fig. 57F shows a preferred arrangement for a normal "Y" section; that on the right, a more conventional design. Either will yield satisfactory results if chills are properly applied.

It is not possible to design a chill set-up without definite information regarding the location and arrangement of heads and gates (i.e., the direction of solidification). Our remarks are, therefore, very general, and do not necessarily reflect what we would do in actual practice, with full information at hand. As mentioned heretofore, the writer has not technical data to submit reinforcing these comments. They are entirely from developed practices that are sound, to the extent employed. They are offered as material for discussion, for what it may be worth.

In closing, it is desired to compliment the authors on a very excellent, complete and concise presentation of essential data, both to foundry men and casting designers, and to recommend its close study by every one interested in the steel casting industry.

P. E. McKINNEY[6] *(Written discussion):* The authors are to be complimented on this paper and the manner in which this paper brings the very important subject of design to the attention of designing engineers and steel foundrymen.

Part 1 contains a really valuable compilation of data showing various cross-sections and indicating potential difficulties to be anticipated when design of steel castings departs too greatly from symmetrical lines.

While it may be agreed that some of the examples are exaggerated and not representative of the soundness to be expected in steel castings in actual practice when properly headed, these illustrations tend to demonstrate forcefully the inherent solidification characteristics of

⁶ Metallurgical Engineer, Bethlehem Steel Co., Bethlehem, Pa.

various metal sections and indicate what can normally be expected when design is such that these sections cannot be aided through the foundryman's ingenuity in placing feeders, heads and other devices to correct the conditions.

Reference is made in Part 1 regarding application of welding for joining intricate cast sections such as shown in Fig. 5. While a comparatively recent development in casting design, it is one that has many possibilities in improving quality of large mechanical assemblies which involve intricate shapes that in larger composite castings cannot be made 100 per cent sound.

The attachment of brackets and other members to castings which are simply intended for fastening the castings to bulkheads and do not enter into the functioning of the casting proper, may be so located as to make the casting a feeding member for the bracket thus bleeding the casting of much needed metal and contributing to the unsoundness of the casting proper. In such cases attachment by welding appears to be the logical and proper procedure.

The present development of welding technique has justified assembly of many wrought members in very complicated weldments. In many cases these weldments are advocated as substitutions for castings and many times welds are placed in locations which are far more highly stressed than would be the case in assembly as a weldment of cast members where the welded joints can be deliberately located in the lower stressed areas.

This method of construction involving assembly of intricate castings into weldments is not new but has been commercially practiced to a certain extent for quite some time. To those interested in this subject, we would suggest their careful reading of the article appearing in the November 1936 issue of the *Journal of the American Society of Naval Engineers* entitled "Ten Years of Welding Development" by N. L. Mochel.

The reference by the authors of their paper in Part 2 to temperature gradients as influenced by tapered sections in assuring proper directional solidification, is a very important point and of interest both to designer and foundryman. It appears to the writer that reference to these features of design might also be forcefully brought out in Part 1 of the paper. There is probably no single feature of design which leads to more difficulty than the tendency of designers to attempt to save weight solely based on the demands of design stresses through the removal of metal from areas in the casting which will result in isolated heavy sections without any thought as to directional solidification of these sections.

The fullest cooperation between foundryman and designer in working out the basic lines of a design will in many cases make the production of quite intricate shaped castings very practical. In actual practice there are many instances where foundrymen who have intelligently re-

developed lines on castings to assure proper feed, have later suffered a penalty for the removal of excess metal which was not shown on the original design and not provided for on the customer's patterns. Such practices on the part of casting users will naturally tend to discourage intelligent consideration of design problems by the steel foundryman.

As a result of recent discussions with European foundrymen, I have a very definite impression that the practice abroad is generally to allow the foundryman to dictate the fundamental lines of a casting during the process of design to a far greater extent than is generally practiced in this country.

The reference by the authors to the benefits to be derived from the use of models such as shown in Fig. 32, is very interesting and I believe that those foundrymen who have gone to the trouble of preparing models for difficult castings have found that this expenditure was fully justified.

There are many cases where it is impossible to visualize thoroughly all the ramifications of the design from the preliminary drawings but by the use of a model with the cooperative study by both foundryman and designer, changes can almost invariably be made to overcome many of the objectionable features which otherwise might not be discovered until several castings had been cast and machined.

The authors' reference to chills is extremely interesting and the various sketches illustrating the influence of chills on various commonly used cast steel sections, contains a great deal of original information which should be extremely valuable not only to the foundryman but should sound a warning to designers as to what may be expected when we depart too far from the natural laws of solidification of molten metal.

In reading this discussion of chills we get the general impression that the authors roundly condemn internal chills as having little or no justification in casting production. It would be quite unfortunate if such an impression became generally prevalent as there are many cases where the location of a difficult section can only properly be handled by internal chills due either to inaccessibility of some of the surfaces which prevent the use of external chills or the impossibility of applying small cored areas for equalizing metal cross sections.

Every foundryman who has intelligently studied casting technique, objects to the general use of chills. It can safely be stated that chills are a court of last resort to overcome the unsound features of casting design which are in many cases necessary regardless of all efforts of the designer to avoid such conditions.

While chills, either internal or external, are far from desirable, the intelligent use of chills by the foundryman will produce a casting with far greater utility than to avoid the use of chills and resort only to heading regardless of the adverse sections of the casting under consideration. It is safe to state that if definite prohibition against the use of internal chills were enforced in any buyer's specifications, there are

castings now being produced which are perfectly sound and suitable for service which would have to be drastically modified in order to assure a serviceable product.

In my opinion, some of the examples shown by the authors as typical of internal-chilled sections, are not fully representative of the best practices in internal chilling of steel castings as used today and, therefore, might be misleading to the users of castings and tend toward a definite prohibition of internal chilling, which would not be to the interest of either producers or consumers of steel castings.

Progressive steel foundrymen have, during the last few years, been giving considerable thought to the intelligent application of internal chills in sections of castings which are inaccessible either to external chilling or to proper feeding and in many cases special chills are designed and developed for a particular casting, which have been balanced between proportion of excess metal in a section not capable of natural feeding and that of chill metal.

Fig. 58 shows a few typical chills which have been selected from many coming within the writer's observation which have been used for chilling special sections. The use of internal chills is particularly necessary at times in heavy-sectioned castings where any amount of external chilling would be ineffective in making isolated heavy sections entirely sound. With the proper balance between chilling metal and section, very good welding of the chill metal and the casting metal can be accomplished. An example of this is shown in Fig. 59.
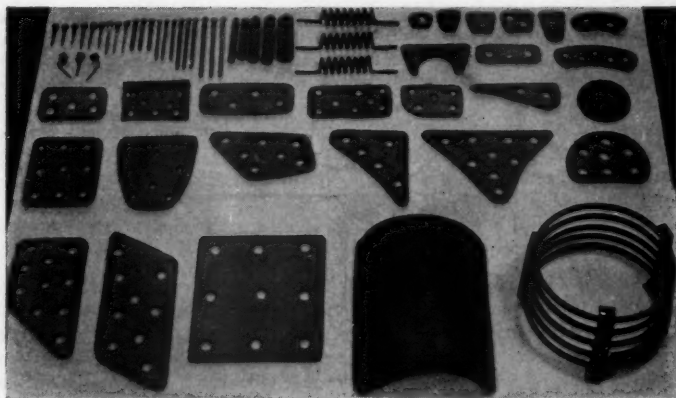
FIG. 58—VARIOUS TYPES OF CHILLS USED IN MAKING STEEL CASTINGS.
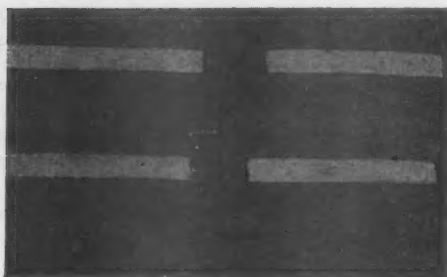
FIG. 59—WITH PROPER BALANCE BETWEEN CHILLING METAL AND SECTION, VERY GOOD WELDING OF THE CHILL METAL AND THE CASTING CAN BE ACCOMPLISHED AS ILLUSTRATED ABOVE.
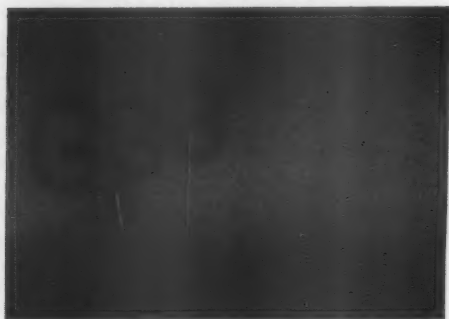
FIG. 60—RADIOGRAPHY OF SECTION CONTAINING INTERNAL CHILL. THE POSITION OF THE FILM AND SECTION IS SHOWN IN FIG. 61.
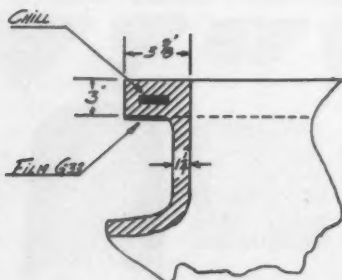
FIG. 61—LOCATION OF SECTION CONTAINING INTERNAL CHILL AND POSITION OF FILM.

While it might be argued that there is a dissimilarity of metal between the chill and the cast structure and that the strength of the chill section is not equal to that of a sound cast section of metal, it should be borne in mind that a section rendered solid by the intelligent use of internal chills is far better than a section with a porous center as a result of improper feeding or lack of entirely effective external chills.

In our study of chills we have made many examinations through sections containing internal chills where we have found no imperfections even as a result of radiographic examinations. A typical example of this is shown in Figs. 60 and 61 representing a radiograph taken through a rather heavily chilled area, which shows nothing alarming or no indications of discontinuity of metal.

It is the writer's opinion that until we more nearly approach the casting design with ideal lines of an ingot, the foundryman will have to resort to the use of every device available including internal chills to assure a finished product suitable for the service intended.

F. A. MELMOTH[7] *(Written discussion):* In the founding industry our conceptions of the effect of design, viewed particularly from the standpoint of the production of defects, have been evolved as a result of practical experience with actual castings and the effect of various methods used, proved of value in their elimination.

The demands on all enginering materials have been sharply increased by intensified working conditions, and by the modern demand for weight saving, and many things classed today as defects of consequence would have been dismissed as insignificant some few years ago. The value of the present paper is greatly increased by an acceptance of the fact that an intensification of this attitude is quite likely in the future.

It always has seemed to me that the first point for the designer to accept fully is that most of the defects referred to constitute the natural inherent behavior of solidifying metal, and in most cases, therefore, consist of a condition not wilfully caused by the shortcomings of the foundryman, whose skill enters into the question to the extent by which he can devise ingenious methods to overcome the perfectly natural phases of behavior. If this is appreciated by the designer, then he joins hands with the foundryman in the attempt to overcome this natural condition, whereas very often the situation becomes one in which the designer and producer are equally busy blaming one another for something for which neither possesses the complete remedy.

The authors have presented, in an organized way, a series of design comments and suggestions, many of which have been made during numerous discussions on the subject which have taken place at A.F.A. meetings. But, in addition, they have also endeavored to show why these things happen, and from their conclusions have evolved rules of design with which very few foundrymen could quarrel.

[7] Vice President, Detroit Steel Casting Co., Detroit, Mich.

While the paper is of real value to the founding industry, its greatest value will be the degree of acceptance extended to it by the designing engineer. It contains enough sound reasoning and demonstrations to affect materially the ease with which steel castings can be produced free from defects, if the conclusions drawn receive the ready acceptance of the designer, and if the latter applies them practically in future work.

The present writer suggests very strongly that every engineering college which includes in its curriculum courses on design, be supplied with a copy of this paper, and thereby given an opportunity to present to engineering students an enlightened viewpoint on the effects of design.

One might argue the practical significance of a defect in a steel casting being so placed as not to affect performance value, and suggest, therefore, that many of the defects illustrated are so placed and of such size as to be no detriment. There can be no doubt, however, that the soundest viewpoint is to realize that what may be fairly acceptable and of limited significance today, rapid changes in engineering practice in the near future may cause to become even of critical effect, and that, therefore, our aim should be absolute soundness, as a combination of foundry efficiency and ingenuity with intelligent design. The present writer merely repeats what he has said innumerable times during the last few years, when he insists that only by such combination can we ever hope to achieve success in steel casting production, which is the ultimate aim of all of us.

C. W. BRIGGS, R. A. GEZELIUS and A. R. DONALDSON (*Authors' reply to written discussion*): We wish to thank Messrs. Phillips and Troy for studying the paper so critically and calling to our attention that the title of Table 7 is in error. The title should be "Tabulation of External Chills."

The procedure employed, in regard to testing of bars cast to size, was to mold the bars horizontally in green sand. Four bars were cast in each mold. They were fed from one end from a common down gate. The other ends were free and contained only a small vent to allow for the escape of gas. After casting and cooling to room temperature, it was found that the sand peeled away from the casting, leaving a smooth, sand free surface. Only in one or two places was it necessary to complete cleaning with a wire brush. The bars were not blast cleaned.

Messrs. Phillips and Troy, in commenting on the use of solid and coil chills, have asked, "Does it not appear reasonable that in the case of the coil chill that the possibility of large internal stresses as a result of the chilling action, is lessened?" In reply to this question, we agree that it does appear reasonable, but we have no definite information on the subject and a mathematical treatise on the subject is not very enlightening as either method can be made to appear favorable by reasonable assumptions.

We were very much interested in Mr. Riegel's statements concerning chills and chaplets We believe that there are other organizations besides the Navy and the Caterpillar Tractor Co. that are like minded regarding their use.

We wish to thank Mr. Sims for his interesting and instructive discussion. He suggests that the first rule presented on page 608, is apparently contradicted by the first rule in paragraph 57. We were wondering if Mr. Sims would be a little less terse and allow us to suggest that the first general rule has been modified and exemplified by the more precise subsequent information.

We appreciate Mr. Sims' discussion of the corner hot tear in thin sections and we are grateful to him for reminding us that we had failed to include this point in the portion of the paper dealing with the foundryman and steel casting design since it is evidently something that cannot be controlled completely by the designer.

Mr. Sims has called attention to the probable lack of usefulness of the sine wave construction. His suggestions in regard to the illustration are to the point. Perhaps we were unfortunate in the selection of the illustration. We know of cases where such a type construction prevented cracking. It, however, should be added that the differences in the adjoining sections were very large. We also believe that wheels which have been designed with a single wave or half a wave spokes are quite successfully withstanding hot tearing whereas in many cases the straight arm design has not been so fortunate.

We also prescribe to Mr. Sims' statement that "shrinkage cavities at intersections may be so well buried as not to impair the structural integrity or strength of a casting in the slightest." There is one point that Mr. Sims has neglected to add. Shrinkage cavities, because of their very irregular and ragged outlines are excellent concentration centers of stresses under exacting operating conditions. Cracks opening up from these cavities may traverse the section. In high pressure, high temperature steam systems, such cavities have been known to be responsible for the development of leaks after an initial operating period. The cracks or weeps may be so fine as not to impair the strength of the casting, though they definitely have impaired its operating efficiency. It has also been our experience that the use of internal chills, because of their lack of fusion, may lead to similar difficulties.

We were very much interested in the opinions of Mr. Everett. We believe that it is quite possible to interpret a set of data in more than one way and, of course, everyone is entitled to his opinions. We, however, are sorry that Mr. Everett was unable to present any data to substantiate his opinions. Replies to discussions of this type are most impossible to formulate.

We believe that Mr. Everett has misinterpreted the data on design to apply only to sections shown in the horizontal position and then merely to a 3 x 3 in. section. We have carefully explained in the text (paragraph 43) that the depth of the section has no bearing on the

shape of the defect. And, whether the section is horizontal, vertical, or inclined at an angle, makes no difference in the rate of solidification providing the mold is filled quickly which is usually the case. In the few cases in which we studied vertical adjoining members, we found that the position and the size of the defect agreed very well with our basic set of data.

For reasons which we definitely stated in the text (paragraphs 40 and 41), we believe that we chose the best section possible for the study. The solidification of steel follows certain laws and the size and shape of the mold cavity does not play as important a part as might casually be considered. We should like to refer Mr. Everett to another publication of the author's* wherein the subject of solidification is treated at considerable length. The liquid contraction and the solidifying contraction are the same whether a steel is solidifying in a 3 in. section or in a 1 in. section, and cavities and center line weakness which may be present macroscopically in the 3 in. section may appear microscopically in a 1 in. section.

We disagree with Mr. Everett in that unfed 3 in. sections are extremely unusual. We suggest he study over designs of heavy construction equipment. Such designs are not infrequent in Naval shipbuilding. This we believe, however, is all beside the point, since our work has shown us at least that the points illustrated in the design charts will be of a comparable order in sections other than those of 3 in.

If Mr. Everett will again refer to Fig. 8, he will notice that the radiograph merely *"illustrates* the general characteristics." The test says nothing concerning the methods used. In this case, it so happened that chills were used at the inner radii.

Our recommendation in regards to the designing of unfed "L" sections is given in paragraph 57, and not paragraph 47. We did not recommend a ½ to 1 in. radius for the inside radii. We merely suggest that the best common practice appears to be one in which radii of ½ to 1 in. would be used. We are only mildly interested in Mr. Everett's rule in regards to the radius of the internal corner. In the first place, if only the inner radii is considered on an unfed joining "L" section and this radii is increased without regard to the modification of the outside radius or corner, it is very easily seen that the contraction cavity will increase as the inner radii increases. In the second place, it has been our observation that in most engineering designs on either fed or unfed sections, the radii are too sharp. It is our belief that the first step should be moderation backed by actual data. Mr. Everett believes that section 106 is to be preferred to section 104 and backs up this preference by stating that 106 would be less susceptible to hot tear formation. It is true that the 106 section is less susceptible to concentration of contraction stresses, but we should like to suggest to Mr. Everett that we are not considering this subject from only the foundry viewpoint. Uniform sections that

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* Studies on the Solidification and Contraction in Steel Castings — III — The Rate of Skin Formation, Trans., American Foundrymen's Association, vol. 43, p. 274, (1935).

are not padded will develop centerline weakness which will show up pronouncedly on radiography. In this section on design for the designing engineers, we are considering that the designer is not discussing his design with the foundryman and, hence, a design of 106 would probably not be considered. We should also like to state that we know that extensive padding is not being used at the present time in the industry, though it is being considered more frequently in the production of high temperature and high pressure equipment.

We are emphatically opposed to Mr. Everett's statement that "it is suggested that caution be exercised by designers in attempting" to use 104 design. We ask that this principle be uniformly adopted. If the foundry finds that such a design interferes with directional solidification, it is a very simple matter for the foundryman to pad the section.

In answer to Mr. Everett's question concerning the heading, gating and pouring of the "T" designs, this information is very clearly stated in paragraph 30. We are sorry, but again it is necessary for us to point out that contrary to Mr. Everett's statement, we did not prepare in paragraph 52 a recommendation in regards to the "V" section. Our recommendation in this regard appears in paragraph 57. We merely stated that from our work on the "designing of 'V' sections, it is suggested that an inner radius not less than one inch be used." We also wish to call the reader's attention to the fact that Mr. Everett's statement, "it is believed that a radius at least 50 per cent and preferably 75 to 100 per cent of the section thickness would be preferable" refers to his own personal opinion. We know of no reason why this arbitrary statement should be true. A variation of section makes little difference in the thermal conductivity of molding sand. It must be remembered that sand is a very good heat insulator. Re-entrant angles such as found at adjoining sections have a poor heat carrying ability and, hence, the metal section at this point has a reduced rate of skin formation. Such a condition is found in a section that is one inch in thickness as well as 3 in. in thickness. For confirmation of this fact, it is suggested that a study be made of Fig. 11 of the previously mentioned reference. We believe that we have been less dogmatic in our observations based on presented data than has Mr. Everett whose statements are based on personal opinion.

We were very much interested in the Gulf Refining Company's specification for chrome molybdenum steel castings. We believe if Mr. Everett had taken the time to study our data on the effect of mass, he would have observed that the results we presented and those tabulated by the Gulf Refining Company, are very similar, considering our data on the alloy steel classification consists of medium manganese and the results he presented are of a chromium-molybdenum classification. One would assume from reading this portion of Mr. Everett's discussion that we did not show that mass had an effect upon the mechanical properties of cast and heat treated steel. We, however, did show emphatically this condition to be true. It is a normal physical condition that exists in cast steel to a varying degree depending upon the compo-

sition. The Gulf Refining Company is to be complimented for recognizing this principle. Most consumers of steel castings believe that they should be able to obtain from any portion of a casting, regardless of the mass of the section, mechanical properties that are equivalent to those of a separately cast coupon. This, of course, is an error and the natural condition that exists due to the effect of mass should be publicized. Whether or not such facts result in a "deleterious effect" is a matter for each designing engineer to decide. Our study of the situation leads us to state that in a general way, the mass effect is probably being amply covered by the present factor of safety that is being employed in design work.

Mr. Everett is again mistaken when he says that "we have no information as to the length of the chills used—consequently no volume figures are available." In the first place, it was stated that the joining sections studied should be considered as having infinite length, actually for experimental purposes the depth or length of the section was 3 in. Thus the length of the chill was 3 in. We fail to see how this would make any difference, however, as the surface area of the chill is given in all cases and the volume figure can, of course, be obtained by multiplying the length of the chill by its surface area. We did not include this figure since we did not wish the data to apply to a section of a definite length, but preferred it to be considered as a condition of infinite length. It is, indeed, surprising that Mr. Everett was not able to recognize the relation between surface area and volume. However, we shall be more elementary and state that we have found, in general, "that the volume of a chill is of prime importance" as regards its chilling ability. We should also like to state, in order to prevent being misunderstood, that in some exceptions this is not true, for after the mass of the chill passes, a certain limiting value further increase in mass is not especially effective. For example, the plate chills in 502-5 could be made twice as long as they are shown in the sketch, and in this way doubling the volume. The defect obtained, however, will be practically the same as shown in the sketch.

We have made at no time a claim that we have exhausted the possibilities of external chill application to section junctures. We realize that we have merely scratched the surface. Of course, we did not discuss or study the placing of chills on the top or bottom surfaces of the junction, because we were assuming that the junction extended indefinitely. Under these circumstances, the junction did not have a top or bottom. It is evident that Mr. Everett can not reconcile himself to this situation.

In regards to Mr. Everett's dissertation on the design of chills, we wish to restate that some of the chills employed are not those that might be used in actual foundry practice. Simple shapes of perhaps unusual design were used so that the effect of the amount of chill surface in contact with the casting and the effect of the surface area (or volume) of the chill might be noted. The relation between surface area and effective perimeter of the chill was of utmost interest to us, as well

as to others that were interested in studying the subject from a technical viewpoint. We should like to ask in this regard if Mr. Everett believes that the large chill in Fig. 57E is one that is well adapted to study this relation? We do not believe that it is.

Mr. Everett has prepared four drawings showing the location and construction of chills for a T, X and two Y sections. These chilling methods he claims are preferred arrangements of good application for the production of sound sections. It is very easy to see that Mr. Everett has not studied his designs in detail. We have experimented on his suggestions and the illustration of Fig. 62 shows the location and the size of defects that are obtained when Mr. Everett's recommendations are followed. We should like to call to the reader's attention that none of the chilling methods produced solid sections and that the size of the defects was larger than that obtained by the chill studies we have carried on and reported in these pages.
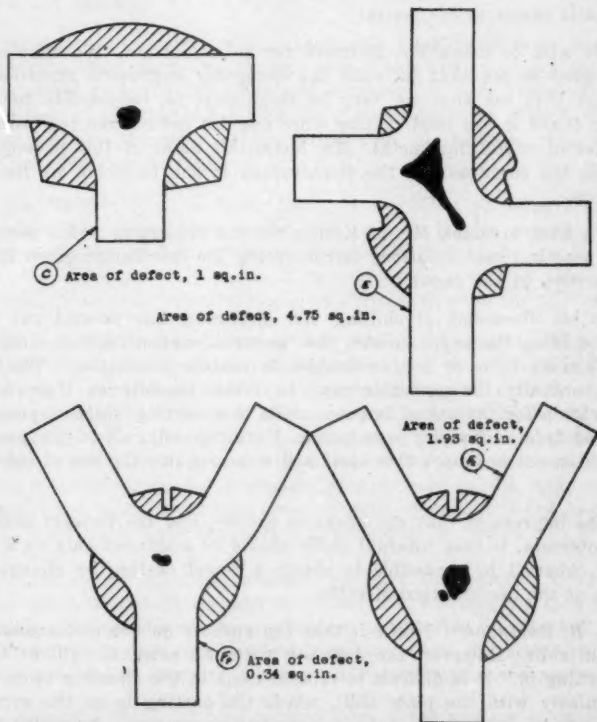
FIG. 62—RADIOGRAPHS OF SECTION USING EVERETT'S SUGGESTED CHILLS. (C) AREA OF DEFECT OF FIG. 57C. (E) AREA OF DEFECT OF FIG. 57E. (FP) AREA OF DEFECT USING PREFERRED SIDE OF FIG. 57F. (FC) AREA OF DEFECT USING CONVENTIONAL SIDE OF FIG. 57F.

It is not our plan, however, to write on the subject "Modern foundry chilling practice." And, since we have not done so, we are at a loss to see why Mr. Everett devoted so much to it in a discussion of our paper. It would seem to us that he could treat his subject more effectively in a paper prepared solely for that purpose.

There are only a few other points that we feel it is necessary for us to discuss. We are well versed in the rate of heat conductivity in sand and metal chills and have made extensive studies on the heat flow in molds at intervals away from the mold-metal interface and wish to state that without going into a lengthy discussion of the reasons that the supposition drawn by Mr. Everett in regards the small sections of sand as portrayed by Figs. 57A and B and 57C is, we believe, not correct. We interpret this condition as the characteristics of the surface area—effective perimeter relationship.

We are sorry it was necessary for us to reply at such length to Mr. Everett's discussion, but his misinterpretation of the facts left us with little choice in the matter.

We wish to thank Mr. Melmoth for his discussion. We were especially glad to see that he calls the designing engineer's attention to the fact that the engineer may in large part be responsible for the defects found in the steel casting since one can not change the inherent behavior of solidifying metal. Mr. Melmoth's point of full cooperation between the engineer and the foundryman is one to which we hardly subscribe.

We wish to thank Mr. McKinney for his discussion which presents many practical considerations corroborating the conclusions given in the first portion of the paper.

In his discussion of chilling Mr. McKinney has pointed out that in his opinion the paper creates the "general impression that—internal chills (have) little or no justification in casting production." We hope that eventually the quotation may be taken literally as there is no justification for the use of internal chills in a casting which is properly designed from a foundry standpoint. Unfortunately, all of the present designs do not approach this ideal and some require the use of internal chills.

The impression that we meant to convey, and the thought that we still entertain, is that internal chills should be employed only as a last resort, when it is impossible to obtain a sound casting by changes in design or the use of external chills.

S. R. ROBINSON[8]: I take it that the authors do not recommend the internal chill. However, the difficulty with the external chill at times is "burning in." It is difficult to remove them in the cleaning room, and particularly with the plate chill, where the casting is on the exposed surface to be painted. It makes a very bad appearance. I wonder what he has to say about that.

MR. GEZELIUS: You are quite right. We are not in favor of internal chills. We believe the internal chill is the last resort, because we have not as yet seen a radiographically sound casting produced by an internal chill. Sometimes by cutting through such a spot it appears to be sound, however, in radiographing you see every internal chill.

As to external appearance, we realize that commercially that is quite important. In the interest of the casting itself, we prefer to see a slight depression where an external chill was placed and know that the casting was sound, rather than have a very nice surface and know there is a hole underneath.

MEMBER: Have the authors had experience with the difference in conductivity of external chill of iron and copper? Have they noticed any difference in the action for the same surface of chilling?

MR. GEZELIUS: We have not tried anything but cast steel chills. It is logical to assume the copper chill might be more effective, but we cannot say exactly.

MEMBER: When the authors mentioned cored sections as eliminating defects, there were two sections mentioned, two cored sections. One was a round core and the other was a triangular core. Now, it seems in the case of the triangular core there might be other effects, such as stresses, due to the effect of the strength of the casting, even though no other effects would appear.

MR. BRIGGS: That is true. However, we have not suggested that the triangular shaped core is the best to use. It was merely carried along as a case of demonstrating the possibilities of design that are involved. We doubt very much if anyone would use it in commercial practice, unless under some varied conditions, such as in the "Y" section.

L. E. EVERETT: In connection with the use of cores for coring out junctions of sections, I do not entirely agree with the authors in that it is always a desirable procedure. In applying a small core in a heavy section of metal there is such a limited heat capacity in the core that the temperature builds up in the center of the section and practically the same shrinkage is gotten with the core in there than without it. Unless the sections are large enough, and the core can be large enough so that it can conduct the heat away rapidly enough, it is apt to defeat its purpose. Such cores are always troublesome to remove.

In the design of steel castings, we have the general problem of taking the heat out of the molten metal, and whether we take it out with the mold surface or the application of external or internal chills, fins or anything that will conduct the heat from the section that is causing trouble, the problem is the same irrespective of the means used.

Unless the factor of heat conductivity of the surface is considered in relation to mass, we do not get very far in trying to work with coring out recesses or applying chills either.

Also in connection with this paper, I wanted to ask the authors if in attempting to chill some of these junctions they have given thought to the necessity in general design of not over-chilling junctures. In other words, where we have to apply directional solidification, we cannot over-chill the junction, or we will stop all the feeding action at that point.

In the design of chills that point must be considered and is really quite a problem. If there is any way known to figure the size and content of a chill for any given section, I would like to know what it is. At present we use the hitandmiss method. It would be very helpful to be able to figure it out.

MR. GEZELIUS: I cannot agree with Mr. Everett that "the most important thing is to get the heat out." The most important thing is the manner in which the heat is dissipated, because just to get the heat out of a casting does not assure that it is going to be sound. Did you ever stop to think that it is absolutely impossible to produce a sound casting by uniform cooling?

MR. EVERETT: I meant to get the heat out of the heavy section.

MR. GEZELIUS: That is perfectly true. I agree with you, that it is practically impossible to calculate the size of the chill required. We have tried to do so in every manner we know, we turned it over to our mathematician and he spent hours and hours on it, and finally gave it up as impossible.

It would be an easy thing to calculate, provided you knew one fact. If you knew the temperature at which the chill and the metal reached equilibrium, it could be calculated very nicely, but there is no way of finding that out.

As to over-chilling, Mr. Briggs could not spend any time on it, but if you will look at Fig. 36C, a three-inch section fed from the bottom, containing a ⅝-in. chill, you will see that even that was large enough to stop the feeding underneath the chill.

CHAIRMAN MELMOTH: There is no doubt but that the future of steel castings is very largely affected by design. The efficiency with which we combat welding propaganda is largely in proportion to the efficiency with which we overcome the defects due to design. If we can produce parts as castings and eliminate defects either due to, or exaggerated by design, welding is one step further behind. On the other hand, if fabricators demonstrate they can weld and produce sufficient structural strength and not have present design defects, which we, as casting producers, allow the impression to get about are incorrectable in our castings, then steel castings are foredoomed to a very limited field of application.

W. J. CORBETT[9]: I am thoroughly in accord with the necessity and desirability of eliminating these so-called defects at junctions. I would

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[9] Works Manager, Atlas Steel Casting Company, Buffalo, N. Y.

like to ask what may seem a rather silly question, that is: how serious are these defects?

There is a possibility that there are many castings in service that have these defects at junctions, and I am wondering if the authors know whether increase in strength has resulted by changing the design only, disregarding the details of foundry practice, to eliminate these defects. That is, what is the difference in strength when the sections are subject to dynamic, static or fatigue stresses?

MR. BRIGGS: That is a very difficult question to answer, because how are you going to test a section of that type? We thought about it for a long time and decided that there was only one thing that could be done. This point we studied and the data is presented in Table 6. In this regard, we made a study of bars, ranging from ½-in. up to 3 in. in diameter in the as cast and the heat treated condition. These bars were not fed and, hence, contained center line weakness. In comparison with test bars with well fed coupons it may be noticed that the ductility has dropped off considerably, while the strengths are about the same. While this gives one some information on the properties of actual sections, it still does not fill the need for information on the mechanical properties of joining sections that contain contraction cavities. It appears that the designing engineer must tell us how important it is to have homogenous sections.

In the Navy, important castings, found by radiography to be defective, are either rejected or repaired to show no defects on further radiography, since we believe that the designing engineer bases his figures on a completely homogenous section. It is only in those cases where we are positive that the section is merely an unimportant tying-in member between other sections that we are inclined to allow the defect to remain.

Radiographic examination is daily receiving wider application. It is being used not only by foundries, but consumers as well. What is to be the opinion of these organizations when they uncover defects such as may be obtained by the lack of fusion of internal chills? The chill may be all right and the section may not be much below the physical property it normally would have if the section were sound, but, nevertheless, there does exist a defect, and in these high-temperature, high-pressure installations is it not possible that such a defect may be considered a detriment? We believe that it does.

R. L. COLLIER[10]: As near as we can discover the needs of the colleges today, the authors of this paper have brought a great deal of information that I think will fit into their desires very nicely. I certainly want to congratulate these authors on a most splendid paper, and one which should go a long way toward answering some of our problems.

MEMBER: There can be no question of the worth of the paper. Having seen some of the preliminary work at Washington, and having ques-

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[10] Secretary, Steel Founders' Society of America, Cleveland, Ohio.

tioned some of the results, I wonder if the authors would enlarge on the statement in paragraph 43: "The depth of the section has, of course, no bearing upon the shape of the defect in the plan view as long as the depth is great enough to permit a defect to form."

Also in paragraph 83, speaking of the effect of the mass, they have noticed, as reported in the paper to the A.S.M. last fall that there is a decrease in density and carbon content and an increase in Izod impact value as the mass increases. That seems directly contrary to what we have reason to expect. I think it would be interesting to have some further information on it.

MR. GEZELIUS: Answering your last question first, it appears to be contradictory, but it is a condition that exists. As you may remember, Dr. Lorig submitted some data at the same time, and on replotting some of Dr. Lorig's data, we found exactly the same condition. As the mass increases, the Izod value also increases; why, I do not know. I would expect it to be the other way, however, that is the condition we found.

It is rather difficult to explain verbally why we consider that the plan view is not affected by the depth of the section. However, if you will imagine a cross section of any one of these sections, and then in your mind visualize how the solidification progresses inward, you will see that if you discount the end sections, and have no end effect, it will make absolutely no difference how long that section may be. It may be three inches or it may be two feet, or it may be any other length, so long as it is long enough so that the end section, and therefore the end effect, does not interfere with the size of the defect. In the first inch on either side of a section there is no defect at all due to solidification of the ends. After you got beyond that point, the plan view should be exactly the same, no matter what the length may be.

Production of Short Cycle Malleable Iron

BY W. D. McMILLAN[1], CHICAGO, ILL.

Abstract

This paper covers the production of a fully annealed malleable iron which has been produced in a cycle of about 14 hours. The charging and molding practice, composition, annealing and the mechanical properties of this metal are discussed from a practical point of view. In this iron, the silicon is higher than the regular malleable iron. There are two factors which operate to shorten the annealing time, the higher silicon content which promotes graphitization and the high percentage of net metal in the annealing furnace made possible by the elimination of heavy pots. This type of iron shows a tensile strength of 55,000 to 60,000 lb. per sq. in. with an elongation of 15 to 20 per cent. The yield point is approximately 44,000 lb. per sq. in. This type of metal should be of general interest because of the lower annealing cost and the higher tensile properties. The paper includes a discussion of the possible suitability of this type of iron for the range of castings produced by the industry.

1. Annealing cycles for malleable iron vary from approximately 6 hr. to 6 days, considering only the time actually in the furnace or oven. The expression "short cycle" is relative and for that reason its use in this paper is briefly explained.

2. Pearlitic malleable, in which either cementite or pearlite or both are retained, is being produced in cycles of approximately 6 hr. Until recently 60 or 70 hr. have been required to produce a full anneal, that is, a structure consisting only of ferrite and temper carbon. Today iron of the conventional analysis, 1.00 per cent silicon and 2.50 per cent carbon, is being annealed in 30 hr. The iron to be discussed in this paper has been annealed in 14 hr. and 28 min., the resulting structure reflecting a full anneal with retention of no cementite or pearlite.

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3. The equipment has been operated for 8 months during which period approximately 5,000 tons of castings were annealed. These castings have shown satisfactory mechanical properties as well as satisfactory machinability.

CHARGING AND FOUNDRY PRACTICE

4. The iron has been melted in an air furnace with a normal capacity of 30 tons which was fired with pulverized coal. The furnace was operated 5 days a week with two heats a day. The iron is of a higher silicon content and the charge has been made up accordingly to provide a higher silicon iron at the tap hole, with allowances for melting loss. The analysis of the charge is varied according to the operation of the furnace but is about 2.00 per cent silicon and 2.85 per cent carbon. The charge is made up of pig iron, sprue, scrap and steel and is not essentially different in make up from a charge for conventional iron.

Fluidity

5. The fluidity of the metal has been determined by casting a spiral bar. A minimum figure was established based on satisfactory results in the molds. It was found that when the iron would give a full section in the test piece at 9, it was suitable for pouring into castings.

6. A spiral similar to that described in Mr. Joseph's article[2] presented at the 1936 A.F.A. Convention, is used, except in cross section. A brief description of the specimen is as follows: The section is a trapezoid, 1⅜-in. wide at the bottom, 1-in. at the top and 0.670-in. deep at the gate. This is tapered uniformly in the length of the bar to 0.700 x 0.590 x 0.200-in. The specimen is marked from 1 to 15, the marks being 90° apart. The gate is 1-in. in diameter. The metal is poured into a basin 2 x 4 x 2-in. deep, the top of which is 4-in. above the bottom of the mold. The outside diameter is 9¼-in. and the space between coils is 0.470-in.

7. While it is possible to pour a bad test piece with good iron, it is practically impossible to pour a good test piece with cold iron. Iron which shows a reading of 9 will be satisfactory but the test pieces on the short cycle iron will often run 11 or 12. The iron from the other furnaces melting a lower silicon iron will run 1 or 2 points further. Fig. 1 shows a test piece poured from iron which

2 Joseph, C. F., "*Measuring and Controlling Pouring Temperatures and Fluidity,*"
TRANS. A.F.A., vol. 44, pp. 103-132, (1936).

ran 1.73 per cent silicon and 2.20 per cent carbon. The specimen
on the right was poured with gray iron and is included to show the
full length of the spiral.

8. Because the characteristics of the standard analysis iron
are well established, we have endeavored to bring out the character-
istics of the short cycle iron more by comparison than by specific
figures.

Gating and Risering

9. With but few exceptions, no changes in gates, feeders and
risers have been necessary to obtain satisfactory castings with this
type of iron.

Yield

10. A check was made on a heat of short cycle iron in which
castings from every pattern on the floor were weighed with and
without sprues to determine the actual yield. All the metal in the
mold was sandblasted before weighing. The same procedure was
followed on a heat running iron of standard composition.

11. Normally the sprue on the type of castings checked will
run about 29 per cent. The average sprue on the two heats is 29.1
per cent. While these heats were not specially selected and should
be fairly representative, the short cycle iron shows a lower per-
centage sprue than the other iron. This is a check on only one heat.
Further checks for specific sprue indicate that the figures will
average practically the same on either type of iron.

FIG. 1—LEFT—FLUIDITY TEST PIECE POURED FROM AN IRON CONTAINING 1.73 PER CENT
SILICON AND 2.20 PER CENT CARBON. RIGHT—GRAY IRON FLUIDITY SPECIMEN.

12. The average net weight of castings shown (Table 1) is based on the castings in one mold from each pattern. It is not necessarily the average weight of all the castings poured, as no allowance is made for more molds of one pattern than another.

Table 1

YIELD DATA

	Short Cycle	Standard Composition
No. of Patterns Checked	53	65
Ave. No. of Castings per mold	2.74	2.50
Ave. Net Weight of Castings	1.86	2.63
Maximum Weight of Castings	11.87	15.25
Minimum Weight of Castings	0.38	0.42
Ave. Percent of Sprue	27.5	30.7
Maximum Percent	51.6	63.8
Minimum Percent	6.0	11.0

The average weight of castings throughout the foundry is about 0.9 lb.

Scrap

13. From the records on the last 38 operating days, the hard iron scrap on the short cycle iron has averaged 0.03 per cent higher than the scrap throughout the foundry during the same period.

Graphitization Tendencies

14. Because this iron has greater tendency to form primary graphite than the lower silicon iron, factors which promote graphitization should be given consideration. A decided drop in temperature before the iron reaches the mold or any condition which will permit slow cooling of the castings in the mold, will promote graphitization. The size of the section, a shape which pockets the green sand and the use of dry sand cores, particularly when the metal runs between cores, must be considered.

Metal Temperatures

15. This iron has been produced in a foundry in which about 200 tons a day are melted. The average weight of castings ranges from 0.7 to 0.9 lb. About 4 per cent of the tonnage goes into chain links which average less than 0.1 lb. To pour such castings, iron of any analysis must be hot. We have not found it necessary to maintain higher temperatures on this iron than the other type of iron. This is substantiated by the fact that the melting ratio on the furnace melting the short cycle iron is practically the same

as the iron on other furnaces melting the same tonnage per heat.
Temperature readings using a disappearing filament type optical
pyrometer range from about 2750°F. to 2850°F. It is important
that sufficient time is allowed after the charge is melted to obtain
through absorption of carbon and uniformity throughout the melt.

Composition

16. The analysis of the short cycle iron should come within
the ranges shown in Table 2.

Table 2

Composition of Short Cycle Malleable Iron*

Silicon	1.60 to 1.90
Carbon	2.00 to 2.40
Manganese	0.25 to 0.35
Phosphorus	Less than 0.12
Sulphur	Less than 0.10

* The sum of the carbon and silicon should be about 3.90 per cent.

Composition Control

17. The silicon content should not be less than 1.60 per cent
as it is difficult to obtain complete graphitization with silicon be-
low that figure. The silicon should not be higher than 1.90 per cent
regardless of carbon content, because of the possible formation of
primary graphite in some castings. Carbon as low as 2.00 per cent,
especially with silicon at the low side, is difficult to graphitize.

18. Carbon over 2.40 per cent will tend toward the formation
of primary graphite. Both elements should be controlled on the low
side to provide sufficient graphitizing influence to obtain a full
anneal and controlled on the high side to prevent the formation of
primary graphite.

Effect of Manganese

19. As with the standard analysis iron, the manganese and
sulphur should be properly balanced. We have obtained a full
anneal with manganese as high as 0.50 per cent with the sulphur
0.08 per cent. An excess of manganese will effect the elongation
to some extent but from a limited experience does not retard graph-
itization. A manganese content of 0.33 per cent with a sulphur of
0.13 per cent (2.6 to 1) resulted in the persistence of coarse
abnormal pearlite.

Effect of Phosphorus

20. The phosphorus content is held less than 0.12 per cent. We have had no experience with this type of iron with a higher phosphorus. It is believed that with a higher phosphorus content low temperature impact values would be lowered and that the iron would show lower impact values when cooled from within the galvanizing temperature range. This is discussed more fully under mechanical properties.

21. In addition to control of the composition, it is important that the metal be held in a molten state long enough for complete solution of the graphite and to obtain uniformity throughout the heat. It is equally important that the metal be thoroughly skimmed and that sufficient time is allowed to free itself from non-metallic inclusions. Because this metal has a greater tendency to precipitate graphite than the lower silicon iron, any condition which induces graphitization should be avoided. Silicon or carbon additions to the furnace should be given sufficient time for complete absorption.

ANNEALING

22. The equipment for annealing the short cycle iron is a gas-fired, radiant-tube, pusher-type furnace containing three rows or trays. There are 93 trays in the furnace, 31 on each rail. The trays weigh 130 lb. and are 20 x 24-in., the 20-in. dimension being in the direction of travel. There is a clearance of 25-in. above the rail but the trays are seldom loaded more than 15-in. high. The load at each push is 1200 lb. of which 800 lb. is net metal. At an interval of 28 min., the furnace will anneal 20 tons of castings in 24 hr. The actual time in the furnace on a 28 minute push is 14 hr. 28 min.

Furnace Control

23. There are five zones in the furnace and five thermocouples. The first three couples are connected with controlling and recording instruments. The fourth and fifth couples are connected with instruments which are recording only. No. 4 is arranged so that control may be obtained if necessary.

Cycle

24. The iron reaches a temperature of 1700°F. in about 3 hr. and is held at that temperature for about 2½ hr. It is cooled to 1475°F. in 2 hr. by air which is forced through tubes in the roof and floor of the furnace. It is cooled to 1200°F. in about 6 hrs.

and further reduced in temperature in the fifth zone by cooling tubes to facilitate handling; the second fast cool period being about 1 hr.

25. Scale is prevented by the use of an atmosphere which is forced into the furnace maintaining a slight pressure on the furnace. The atmosphere is produced from 800 B.T.U. gas by partial combustion and dehydration.

26. The equipment has operated on less than 2 cu. ft. of 800 B.T.U. gas per net pound of castings including the gas required for the atmosphere.

MECHANICAL PROPERTIES

27. Following the established practice, the mechanical properties have been determined principally on the ⅝-in. diameter, round test bar, A.S.T.M. specification A47-33. In addition, some tests have been made with castings on which values have been established when cast in the conventional analysis iron.

28. In general, the type of castings produced in this foundry is such that iron which will meet the A.S.T.M. specification A47-33, Grade 32510, is satisfactory. Tensile strength of 50,000 lb. per sq. in. and elongation of 10 per cent in 2-in. are considered minimum figures. The decision to produce this type of iron on a schedule of 20 tons a day was based mainly on the advantages which would result from the shorter annealing cycle. Preliminary tests indicated that this type of iron would show better mechanical properties but it was not adopted with the specific idea of producing a higher strength iron.

Tensile Properties

29. The results shown in the following table are taken from the production control records. They reflect the grade and range in properties of iron that "as cast" test bars indicate. They reflect all the imperfections to which foundry test bars are heir.

30. The results of the tests on the first 100 heats range in tensile strength from 53,250 to 67,900 lb. per sq. in. Incidentally, this bar later showed an elongation of 18.8 per cent in 2-in. The per cent elongation on the same heats ranged from 9.4 per cent to 22.7 per cent. Only one result was less than 10 per cent.

31. The results tabulated by weeks are the averages of all the bars tested. The range, however, is practically the same as

shown for the first 100 heats. For example, the results for the week
ending 7/10 which averaged 59,150 lb. per sq. in. tensile and 15.4
per cent elongation, ranged in tensile from 57,200 to 62,000 lb.
per sq. in., the elongation from 12.5 per cent to 21.1 per cent.
Tests on recent production show practically the same results. For
the week ending 12/10, the tensile ranged from 59,100 to 62,150 lb.
per sq. in. and the elongation from 12.5 per cent to 20.3 per cent.

Yield Point

32. In production control tests, the yield point is not de-
termined. However, a sufficient number of tests have been made

Table 3

RESULTS OF TESTS ON 190 HEATS

	Average Ten- sile Strength, Lb. Per. Sq. In.	Average Elong, in 2-in., Per Cent
First 100 Heats — 3/25 to 6/7/37	60,100	15.3
Week Ending 7/ 3/37 (10 Heats)	58,050	12.6
Week Ending 7/10/37 (10 Heats)	59,150	15.4
Week Ending 7/17/37 (10 Heats)	60,150	14.5
Week Ending 7/24/37 (10 Heats)	59,150	14.8
Week Ending 7/31/37 (10 Heats)	59,500	15.0
Week Ending 8/ 7/37 (10 Heats)	58,680	14.3
Week Ending 8/14/37 (10 Heats)	57,100	12.6
Week Ending 8/21/37 (10 Heats)	64,550	17.8
Week Ending 8/28/37 (10 Heats)	61,400	17.0
Average — 190 Heats	59,697	14.8

to indicate that the yield point is definitely higher than the yield
point of the regular iron. In 24 tests on "as cast" bars, the yield
point ranged from 41,000 to 46,000 lb. per sq. in. and averaged
43,600 lb. per sq. in. The yield point on iron running about
1.20 per cent silicon, annealed in a 100 hr. cycle, is about 39,000
lb. per sq. in.

Low Temperature Characteristics

33. Tensile strength tests were made on bars from the same
heats at room temperature and at 0°F. This was done to determine
if the percentage elongation would be less at low temperature,
particularly on iron which ran slightly more than 0.10 per cent
phosphorus.

34. The results of the five tests on this type of iron are given in Table 4.

35. In the above tests, there is a very slight decrease in the percentage elongation while the tensile strength at the low temperature has increased about 2.5 per cent. This iron showed a phosphorus content ranging from 0.106 to 0.126 per cent. The carbon

Table 4
Room and Low Temperature Properties

At 70-75°F.		At 0°F.	
Tensile Strength, Lb. Per Sq. In.	Elongation, Per cent	Tensile Strength, Lb. Per Sq. In.	Elongation, Per cent
60,800	18.8	62,000	21.8
63,000	17.2	64,500	21.8
59,750	19.3	61,450	18.0
59,750	18.0	61,100	14.1
55,000	14.8	56,800	11.7
Ave. 59,660	17.6	61,170	17.5

ranged from 2.12 to 2.38 per cent and the silicon from 1.67 to 1.90 per cent.

Drop Tests

36. No tests have been made for impact values with either Izod or Charpy specimens. Because it was expected that the impact values rather than the tensile properties would be less at low temperatures, drop tests were made on castings at room temperature and at 0°F. This type of casting was selected for the impact test because values already had been established on this casting from a series of regular routine tests. The test consists of dropping a 15 lb. weight from a height of 19-in. The point of impact is 3-in. from the support. Castings from this type of iron will average approximately 13 drops before failure. The tests made on these castings at 0°F. averaged 15.2 drops.

Adaptability and Potential Uses

37. Since this type of iron offers several advantages, among them higher tensile strength and higher yield point and lower ultimate cost, principally because of a lower annealing cost and

quicker delivery of castings, it is thought that there should be an interest in its application to castings other than the relatively light section implement parts.

38.   Because of the higher silicon content, there is a greater tendency to form primary graphite than with the conventional 1.00 per cent silicon iron. To determine the size of section in which this iron would be clear and also to determine the effect of primary graphite on the tensile properties, a study was made utilizing a wedge bar and the standard tensile test bar. The test bars were examined in the ¾-in. section after anneal. The wedge bars were examined as cast. A brief description of the wedge bar follows.

39.   The bar is 1 x1-in. at the butt and tapered in 13-in. to 3/16-in. At spacings of 1-in. beginning at the heavy end, there are "V" shaped grooves which run across the top and bottom of the bar. These grooves are to facilitate breaking the wedge in the hard state. The first groove represents a section 1-in. thick and the sixth groove represents a section 11/16-in. thick. This bar is gated between the second and third grooves and the gate is approximately 1 x ⅜-in. There are two bars in the mold and they are fed by a gate which separates the bars by approximately 2.5-in.

40.   Examination of the wedge bars indicates that this type of iron will be clear in a 1-in. square section when the rate of cooling is not retarded abnormally.

41.   Wedge bars from 57 heats were examined and 46 of them were clear in the 1-in. section. There were 11 heats which were not clear in a 1-in. section, 7 of which showed a silicon-carbon total of over 4.00 per cent. Nine of the 46 heats which were clear in the 1-in. section showed a silicon-carbon total of over 4.00 per cent. Of the 16 heats which were higher than 4.00 per cent silicon-carbon total, nine were clear and seven were mottled. This indicates the importance of a proper pouring temperature.

42.   The importance of pouring temperature and of sufficient time in the molten state is emphasized further by the occurrence of a mottled condition in a 1-in. round sprue when the silicon-carbon total was as low as 3.80 per cent. It is reasonable to assume that, with proper degree of super-heat and with a composition within the desired range, this type of iron would be clear in a 1-in. square section. At the same time consideration must be given to factors which retard the rate of cooling in the mold.

43. The extent to which this type of iron may be used in heavier castings involves also the grade of iron specified and the minimum tensile figures which are acceptable. When a tensile strength of 50,000 lb. per sq. in. and an elongation of 10 per cent is acceptable, such castings may be produced from a relatively high carbon iron with 1.00 per cent silicon, in which case the castings normally will be free from primary graphite.

44. With the conventional type of iron, it is considered almost essential that castings be free from primary graphite. With a 1.00 per cent silicon iron, primary graphite is the result of an excessively high carbon at the same time influenced by pouring temperature. The effect of primary graphite is to lower the tensile properties. To maintain tensile strength of 50,000 lb. per sq. in. with a high carbon iron, the complete elimination of primary graphite is essential. The structure, as well as the composition, is controlled to give the desired tensile properties.

45. The tests made on the higher-silicon iron indicate that it is entirely practical to maintain 50,000 lb. per sq. in. tensile strength and 10 per cent elongation with iron which shows some primary graphite. As this type of iron will normally show 58,000 to 60,000 lb. per sq. in. tensile strength and 15 to 20 per cent elongation, some primary graphite may be permitted. While the tensile properties will be lowered, the resulting tensile values will still be greater than 50,000 lb. per sq. in. and 10 per cent elongation. As with the conventional type of iron, the structure, as well as the composition, is controlled to give the desired tensile properties.

46. The extent to which primary graphite affects the tensile strength and elongation, has been determined on 72 heats. The ¾-in. section of the standard test bar was examined under the microscope after annealing and the number of graphite spots or areas was counted in the full section.

47. Practically the same results were obtained on bars which showed as high as 12 distinct areas of primary graphite as on those bars which showed none. In bars which showed from 13 to 20 areas, the tensile figures were satisfactory when the primary graphite was fine. When the carbon was high the primary graphite was coarse, the elongation ranged from 6 to 8 per cent while the tensile strength was less affected and averaged about 47,000 lb. per

sq. in. It has been observed that when the graphite formation is very fine, it invariably may be traced to temperature. The relation of coarse and fine primary graphite in this iron might be compared to coarse and fine graphite flakes in the low and high tensile gray cast irons.

48.   The annealing equipment was installed to handle 20 tons of castings per day. Mainly because there has been sufficient tonnage in small castings to operate the furnace on this basis, there has been no definite program to determine the maximum size of section which may be poured from this type of iron. There is considerable development work yet to be done with this iron with regard to both the type of castings and the annealing cycle. Experience indicates that this type of iron may well be considered for other malleable iron castings in view of the advantages of a lower cost and a high quality product.

DISCUSSION

Presiding: H. A. SCHWARTZ, National Malleable and Steel Castings Co., Cleveland, O.

H. BORNSTEIN[1]: I wish to compliment the author of this paper. What he tells us in this paper we would have thought impossible just a few years ago, that is, the production of malleable castings in such a short annealing time. It certainly shows we are making very definite progress in the malleable iron industry.

There are a few points in this paper I would like to discuss. In paragraph 2, the author points out that with the conventional analysis, one per cent Si. and 2½ per cent C., an iron can be annealed in 30 hours. We have done a little experimental work on irons of various analyses in relation to the annealing time and our experience indicates with an analysis such as that, it would require closer to 50 hours rather than 30 hours to complete graphitization. I think value could be added to the paper if we would have more details as to the charge used with the analysis as given in paragraph 4. I would like to see some figures in reference to the amount of pig iron and scrap and so on, both on the conventional analysis and on the analysis to produce the short cycle iron. That would give some of us an opportunity to compare costs which relate both to the production of the white iron and the annealing.

In the paper, reference is made to the atmosphere. I think we would like to know just how that atmosphere is made up, volume and analyses of the gases going into the furnace, as well as an analysis of the atmosphere in the furnace. Some reference is made in the paper as to costs. I think costs are very important because the economics will determine whether you are going to produce short cycle annealing

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[1] Director of Laboratories, Deere & Co., Moline, Ill.

on a production basis. We realize that you do save money on annealing
by reducing the cycle. If you are putting in a furnace, say, to handle a
40-hour cycle, you have got more equipment involved, more initial cost,
than you have with a 14-hour cycle, and fuel costs are lower on the
shorter cycle. On the other hand, how much additional cost have you
as far as the charge is concerned? It is the cost of the equipment,
the detailed annealing costs and the detailed melting costs which will
determine for you what you are going to produce. Also the composition
and, consequently, the annealing cycle will depend to some extent upon
the type of castings which you are to produce.

J. H. LANSING[2]: I would like to state, as has Mr. Bornstein, that
this paper has been very interesting. However, there are one or two
items to which I would like to invite attention. One is in paragraph 2,
the statement that pearlitic malleable is being produced in cycles of
approximately 6 hours. That is undoubtedly a fact but may not apply
to more than one of the several types of material. The only reason
for calling attention to it is that the impression may not be given
that six hours is a standard period for the production of most pearlitic
malleable, which is not the case.

Another item is with regard to the comparison of yield based on
other than identical patterns. I believe that the experience of a great
many of us has been that when one makes a comparison of yield, it
is necessary to use identical patterns, in order to indicate whether the
amount of feeding with one metal may be greater than the amount of
feeding required with another metal. If you are using what we might
call similar patterns, but if they are not identical, I believe there is
considerable room for variation.

One other item is in connection with the drop test, particularly at
low temperature. I believe in that connection it would be of decided
interest to have a comparison made on some basis such as the wedge
test, which has long been used in the malleable iron industry and with
which most producers are familiar. In its case there is a definite
type of specimen used and a definite weight dropped a measured
distance. This should give some comparative figures of interest.

ENRIQUE TOUCEDA[3] *(Written discussion):* In the last few lines
of the abstract of this paper, a statement is made that it includes a
discussion of the possibility of this type of iron for the range of
castings produced by the industry. In this connection the speaker
would point out that unquestionably great progress has been made in
the shortening of the second step in the manufacture of malleable iron
and the makers of these ovens deserve a great deal of credit. It is an
important development, and in certain cases has made possible the
furnishing of castings to the customer even more rapidly than normally
would be required. However, prompt deliveries are but one of the

things desired by the customer and in the speaker's opinion it is second to assurance that the castings will have long life in service.

There is danger, however, that the customer will get an entirely erroneous idea of what may prove practical in the case of his castings, on reading some of the advertising matter that has appeared in the technical press and in papers that have been written on the subject. It is for this reason that the speaker has quoted the last few lines of the abstract. In short, there are a very limited number of plants in the industry that can make use of a hard iron in which the silicon content might vary between the limits of 1.60 and 1.95 per cent, and the shortness of the anneal is dependent upon the silicon content of the hard iron.

The author of the paper has considerable to say regarding the effects of primary graphite on the tensile properties, but fails to furnish data covering its effects on impact value and it is in the latter connection that primary graphite causes its greatest damage.

The speaker is willing to go on record as stating that in the past there is nothing that has caused more damage to the reputation of the product than the presence of primary graphite. Ordinarily its presence is fatal to long life in service.

If the regular jobbing plant is considering the matter of the adoption of an annealing process that will enable him to make deliveries more promptly, he would be foolish indeed to take a chance in connection with hard iron composition, unless he has forgotten the many headaches he and others have experienced in the past. Fundamentally, he must adhere to the composition that has and will prove satisfactory to his clients, and then decide on the particular process that will anneal this composition in the shortest time.

This may prove to be a car type oven fired with pulverized coal or an electric oven in which the anneal can be completed in from two to somewhat less than four days. It must not be forgotten that while there is demand for prompt deliveries, it is contingent upon an assurance of long life in service.

J. T. JANS[4]: Mr. Bornstein asked for some information regarding the atmosphere in the furnace that Mr. McMillan described. The gas preparation unit of that furnace is a rather simple unit which takes about one volume of the 800 B.t.u. mixed gas, available at that plant, which is a mixture of natural gas and coke oven gas, and about four to five volumes of air. These are passed through fixed orifices so that the proportion of gas to air is constant. These are partially burned and then passed through a water condensing and drying compartment before they enter the furnace. Most of the atmosphere gas is added to the two ends of the furnace, to take care of door and vestibule losses. The condition in the body of the furnace is stagnant, experiments in the laboratory and in the field having proven this to be the best condition.

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[4] Engineer, Holcroft & Company, Detroit, Mich.

Your vice-chairman, Mr. Wright, of the Michigan Malleable Iron Company, has one of the best explanations for the formation of a pearlitic ring that I have heard. These irons with which Mr. McMillan's paper deals are high silicon irons. The high silicon promotes graphitization. Gases that will attack carbon and tend to promote decarburization, will attack silicon even more readily. In that way, the outer surface of the casting, if attacked by gasses of this nature, would graphitize slower than the interior of the casting, which is an explanation of the pearlitic ring.

Professor Touceda has a discussion, which you just heard, on the formation of primary graphite. So far, I do not believe that anyone has recommended an installation for high silicon iron with real short cycle, such as 14 hours on sections that run more than 1 in. thick. Now in the paper that Mr. McMillan gave, I believe the average he gave me was 0.9 of a pound. It was not the top range of castings. They ran considerably higher.

There is one other thing. There has been a number of tests run which indicate that you can successfully anneal iron on short cycle which will meet the railroad specifications, which are a minimum of 53,000 lb. per sq. in. tensile, 35,000 lb. per sq. in. yield, and 18 per cent elongation. By increasing the second stage of the annealing cycle somewhat, annealing can be successfully done with these high silicon irons in around 24 to certainly not over 30 hours.

MR. McMILLAN (*Written closure*): I would like to say that I was asked to write a paper giving my experience in the actual production of castings fully annealed in a short cycle, in this case 14½ hours. It was my purpose to set down the procedure followed in making the castings and to avoid technical, theoretical details.

The portion of the paper dealing with the effect of primary graphite on the tensile properties was included with the knowledge that the subject was controversial, because it is an angle that should not be overlooked. The condition of primary graphite in the casting is undesirable. However, it is not something to give in to and evade by using a lower silicon and longer cycle. Rather, it is somewhat of a challenge to control the metal in the furnace and to produce clear castings. This is not impossible and our experience has indicated that it is entirely practical for the size of section we have run. Under unfavorable conditions we get a little primary graphite. While the presence of some may be tolerated we aim to keep all sections clear. Out of 466 heats on the air furnace we had four which showed primary graphite in sections which should be clear to produce a satisfactory casting. This does not mean that all the castings from these heats were mottled. Almost always the mottled condition can be tied up with temperature.

With regard to the 1.60 to 1.90 range in silicon, this range was reported because the records of the heats showed this range. Most of the analyses are within about 18 points of silicon and on the low

side. As stated in the body of the paper, the sum of the silicon and carbon must be controlled. We have poured iron with 1.90 silicon and 2.00 carbon which fully annealed. This does not mean that we strive for such an analysis.

The average weight of casting is about 0.9 lb. A good many will run as high as 15 or 20 lb. We have a very high percentage of castings which weigh about 0.1 lb. which has a misleading effect on the overall average.

With regard to yield, we have not made any drastic changes in gating. Patterns which have run in the conventional iron have been switched to the short cycle iron without change. This applies to relatively light weight castings. It is quite probable that heavier castings will require regating.

Mr. Jans' remarks have covered the question brought up concerning the atmosphere. This is essentially a matter of controlling the CO-CO_2 ratio, and the H_2-H_2O ratio.

With regard to the cost of the charge, we have about 35 per cent return and use about 50 per cent pig iron. Pig iron with 3 per cent silicon carries an extra of about 65 cents a ton. However, if a lower percentage of pig iron is used, 46 per cent instead of 50 per cent, the 65 cents extra is overcome. Less pig iron is advisable in the interests of lower carbon and a charge may be made which costs no more than a similar charge if the pig iron is 2 or 3 per cent less.

In closing, we have produced about 6000 tons of castings. Many of these have been in actual service on power operated equipment long enough to prove that they are entirely satisfactory. This type of iron offers distinct advantages. There are problems to be worked out with any kind of iron and in any foundry. Its application on other types of castings will bring up new problems which must be worked out separately as they come up. The brief recital of experience with this type of iron can hardly be expected to do more than stimulate interest in short cycle malleablizing. If this has been done it may have then served its purpose.

Design of Straightening Equipment for Malleable Iron Castings

By C. W. Weedfall[1], Saginaw, Mich.

Summary

This paper is perhaps the first attempt made to record the knowledge gained by experience and study on this subject and submitted to the industry in written form. Some of the important points discussed are: (1) Causes of warp and deformity in malleable iron castings and the advantages gained by the use of straightened equipment to remove these defects; (2) the effects of straightening and coining pressures upon the structure of malleable iron, which are well illustrated by means of graphs and microphotos; (3) numerous formulas and sketches explain how to obtain the correct pressures on various sections of castings; (4) factors which govern the design of straightening equipment, such as, customer requirements, pattern construction and foundry technique are outlined in general for all types of equipment. These factors are further expanded in a more detailed discussion of the various types of equipment under separate headings, as follows: (a) Straightening fixtures, (b) forming dies, (c) squeeze dies, and (d) hammer dies. These various types of straightening equipment are well illustrated with formulas, charts and drawings. An attempt is made to show when and why to use a certain type of equipment in preference to another. The advantages and disadvantages of each type of equipment are brought out, along with the fundamentals of the use of mechanics in design.

Introduction

1. The use of straightening equipment for processing malleable iron castings in the foundry is increasing, due to a constant effort to reduce costs and to produce a more uniform product.

2. Design of straightening equipment for malleable iron cast-

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ings, which is the subject of this paper, is based upon a number of factors. Knowledge of these factors is necessary to design the proper type of equipment which will produce uniform castings consistenly, and at the least possible expense.

3. Malleable iron castings become warped or deformed when heated to the high temperatures used in the annealing process.

Cause of Warpage

4. Warpage is caused by the relief of internal strains within the casting. When a casting cools in a mold, internal strains are produced. These strains result from unequal cooling of the various sections making up a casting and from the resistance of shrinkage, due to the shape of mold and cores used.

5. Internal strains are neutralized or relieved during the annealing cycle because of the reduction of strength in iron at annealing temperatures. The effect of this neutralization of strains within a casting, which is undergoing an annealing treatment, causes it to change shape or warp. The degree of warpage depends entirely upon the nature of the casting. It can not be prevented by any known packing method used in the annealing process.

Deformed Castings

6. Deformed castings are those which have been bent or twisted out of shape by some force applied externally to the casting during anneal. The amount of deformity is in direct ratio to the care used in packing hard iron castings into annealing pots preparatory to annealing. By referring again to the high temperatures to which castings are subjected during annealing, it readily can be seen that a reverse action to warpage takes place. As the hard iron castings are heated to annealing temperatures, expansion takes place within the confinement of the annealing pots and produces strains which can not be resisted by the weak metal at this temperature, and they, therefore, yield to these forces which produce deformity. Deformity, in castings also is produced by the castings own unsupported weight, or the weight of castings placed on top of one another.

7. Deformity in annealed castings can be lessened to a considerable extent by packing hard iron castings tightly in sand or some other inert material which acts as a support for the casting so that it can withstand the strains that produce deformity. This method cannot be depended upon to produce accurate castings and,

in most cases, straightening equipment still must be used to produce a uniform product.

8.   If straightening equipment must be resorted to, then packing castings with a packing material becomes an unnecessary expense.

9.   To straighten a casting which has become warped or deformed during annealing, requires the application of a blow or pressure sufficient to bend it back to its original shape and produce a permanent set in the metal so that it will retain this shape. Such a force, used in straightening malleable castings, has little effect upon the structure of the metal.

RELATION OF PRESSURE TO HARDNESS

10.   Fig. 1 shows graphically the relation of pressure to deformation and hardness.

11.   The proportional limit, or point where malleable iron begins to take a permanent set sufficient for straightening, as shown on the deformation curve (Fig. 1), is approximately 17,000 lbs. per sq. in. For all practical purposes, a pressure of 20,000 lbs. per sq. in. should be sufficient to straighten the average casting. There is little appreciable change in the hardness curve at this pressure which should indicate no damaging effect to the structure of the iron.

12.   It would be interesting to have a tensile and elongation curve which shows the effects of pressure on malleable iron. We can, however, draw a comparison from the hardness curve. Straightening pressures produce the same results common in the cold working of metals. The hardness and tensile strength increases while the elongation decreases when higher pressures are applied per square inch of material.

STRAINED OR SWELLED CASTINGS

13.   Castings, which have strained or swelled surfaces due to defective molding, or have finish allowed for coin pressing, require the application of higher pressures per square inch on these surfaces than used to straighten a warped or deformed casting. When it is necessary to coin a casting, the pressure applied must produce a permanent set in the metal of approximately 6 per cent. This requires a pressure of 60,000 lb. per sq. in. Castings which are

swelled or strained, will require a pressure corresponding to the per cent of swell or strain to bring them back to pattern dimensions. These required pressures may be read from the deformation curve, and the resulting hardness in the material may be read from the hardness curve.

14. Castings which are compressed to any extent in one direction will grow in another direction. This often changes the dimensions of the casting, and should be considered in die and pattern construction. Excessive coining sometimes results in machining difficulties due to the increase in hardness.

EFFECT OF PRESSURE ON MICROSTRUCTURE

15. Figs. 2 and 3 show micrographs of the grain structure of malleable iron. Fig. 2 shows ordinary malleable iron, while
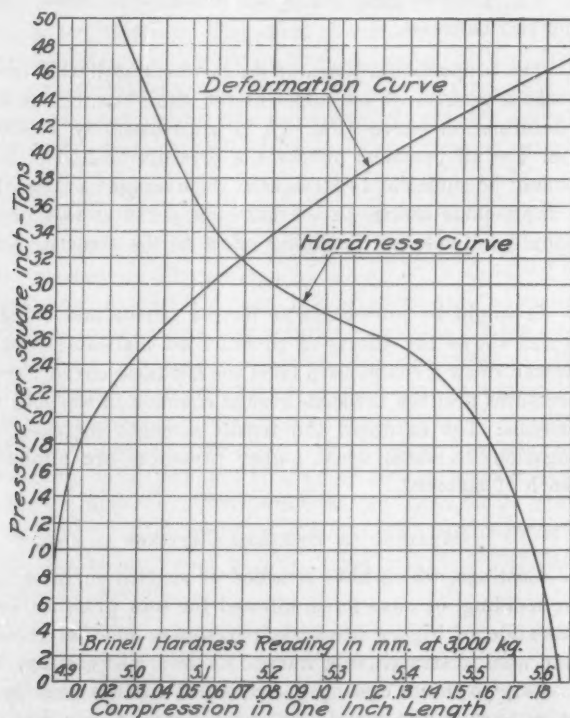
FIG. 1—DEFORMATION AND HARDNESS CURVES SHOWING THE RELATION OF DEFORMATION AND HARDNESS TO PRESSURE.

Fig. 3 shows malleable iron after being subjected to a pressure of 100 tons per sq. in. Fig. 3 clearly shows an elongation or flattening of the grain boundaries and the evidence of slippage within the grain boundaries. This evidence of slippage has been instigated by some authorities as the reason for hardening of metals during cold working.

FORMULAS FOR OBTAINING REQUIRED PRESSURES

16. Fig. 4 shows formulas for obtaining pressures required
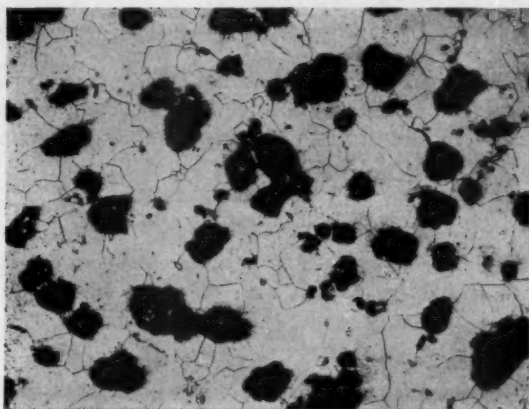
FIG. 2—MICROGRAPH SHOWING THE GRAIN STRUCTURE IN NORMAL MALLEABLE IRON.
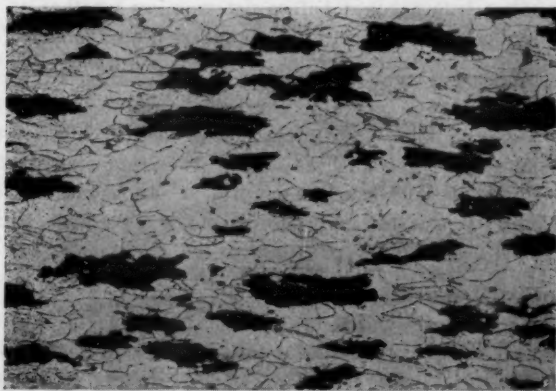
FIG. 3—MICROGRAPH SHOWING THE GRAIN STRUCTURE OF MALLEABLE IRON AFTER BEING SUBJECTED TO A PRESSURE OF 100 TONS PER SQ. IN.
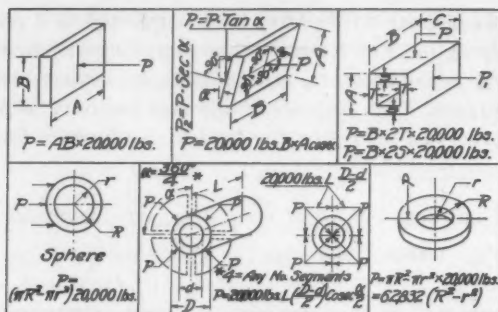
FIG. 4—PRESSURE FORMULAS FOR STRAIGHTENING VARIOUS CASTING SECTIONS.

to straighten various sections of castings. These formulas also can be used to determine the pressure required to obtain any per cent of deformation by substituting the corresponding pressure obtained from Fig. 1 in place of the 20,000 lb.

FACTORS IN EQUIPMENT DESIGN

17. Factors responsible for the different types of equipment designed to straighten castings are: (1) Customer requirements, (2) pattern construction, and (3) foundry technique.

Customer Requirements

18. Customer requirements pertain to quantity of production, order of machine operations in relation to locating or set-up points, types of finish and tolerances allowed. All customer requirements must be taken into account before determining whether to design a straightening fixture, forming die, squeeze die or hammer die.

19. Whatever type of equipment is designed to process a casting, it should be made to straighten and hold all finished surfaces and stopping points in relation to the locating points used for the customer's first machine set-up. It is important that these locating points be held exact and uniform and should not in any way be defaced, due to improper die contact, or have variations due to pattern or foundry practice.

Pattern Construction

20. Details in pattern construction must be known. Clearances in dies usually are made for parting lines and gated portions of castings due to the variations obtained during their removal. Draft angles also should be taken into account, as a casting is

straightened to its pattern dimensions rather than to those shown on customer blueprint.

21. Neglect to consider the above pattern construction points usually results in excessive pressures where these variations occur. This may prevent the die from closing properly, or cause an unsupported casting wall to cave-in.

Foundry Technique

22. Variations, such as, shifts, swells, strains and core variations, must be anticipated. Provisions must be made to handle the average run of castings. Care must be exercised in foundry practice to prevent defects occurring at locating points.

23. The different types of straightening equipment to be discussed in this paper are: Straightening fixtures, forming dies, squeeze dies, and hammer dies. Factors affecting the design of these various equipments will be outlined under their respective headings.

STRAIGHTENING FIXTURES

24. Straightening fixtures are hand-operated devices, consisting of cast or machined blocks conforming with the shape or contour of castings to be straightened, and with or without clamping means to assure positive locating points. Hand hammers are used as the straightening medium.

25. Straightening fixtures are used to process low production castings, which have light sections and are easily bent, and to act as a gauge for finished castings.

26. The accuracy of castings straightened in this manner is entirely up to the judgment of the operator because he must assume when casting is contacting the gauge at the necessary points.

Disadvantages

27. Production is low and costly. Quality is hard to maintain and is a source of customer complaint due to hammer marks and un-uniform castings. This is especially true of heavy castings which are difficult to bend where desired.

Locating Points

28. It is important that a straightening fixture be so designed that it will have the exact locating points in respect to the casting as the fixture used in the machine shop for machining the casting.

If this is not done, there will be trouble machining the castings accurately. This will be true no matter how accurately the castings are straightened in a straightening fixture which disregards customer locating points.

Example

29. When castings must be struck a heavy blow with a hammer, or when a pinch bar is used, the casting to be straightened must be clamped in position, usually at some vital locating point. A good example of this type of straightening fixture is one used successfully to straighten clutch and brake pedals, as shown in Fig. 5. The faces of the pedal boss are the working or locating points and are either machined or coined by customer. If coined, the boss is placed under a coin press and squeezed. No relation is held to any other part of the casting.

30. This means that the face of the boss on the casting must be square with the rest of the casting. If the casting is to be machined, the first operation is to drill the boss. This operation is accomplished by clamping the boss between a flat plate and a cup center through which the drill is fed. This shows how vital the
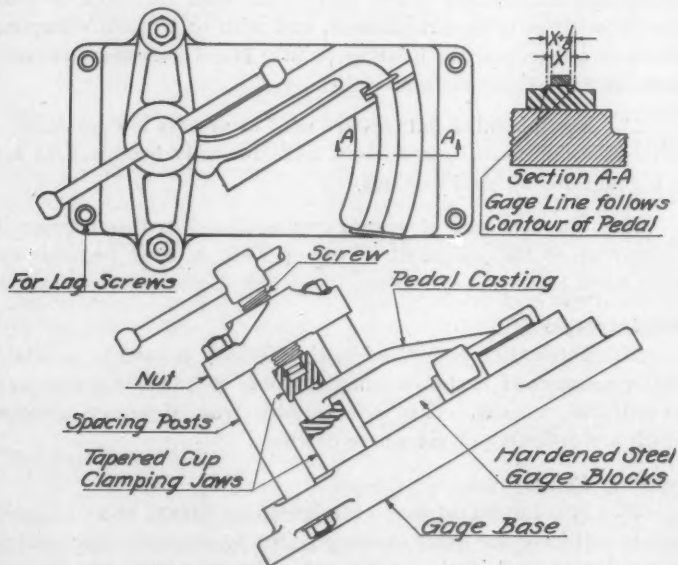
FIG. 5—CLUTCH OR BRAKE PEDAL STRAIGHTENING GAGE.

squareness of the faces of this boss is to the remainder of the casting. An error of 5° on the face of the boss would hardly be noticed with the eye, but it would cause an error of 1 in. at the outer end of the pedal.

31.   To eliminate this error, the faces of the bosses are clamped in a clamping device, which is in relation to the hardened steel blocks built up from the base of the fixture to conform to the angle and contour of the lever arm and foot pad.

32.   Preliminary straightening should be done outside of this fixture in a vise or with the use of a pinch bar, and the final straightening finished with a lead hammer while casting is in the fixture.

FORMING DIES

33.   Forming die consist of a top and lower die made up of stationary parts. The lower die is placed on the bed of a press, while the top dies is secured to the ram. The top and lower dies open and close with the ram movement and usually are held in relation with one another by guide pins.

34.   The purpose of a forming die is to distribute the vertical force of a press into direct or resulting angular forces upon the surface of a casting that is straightened.

35.   The use of forming dies for straightening castings is determined by the following factors:

(1)   Accuracy of casting required.

(2)   The number of casting surfaces upon which pressure can be applied.

(3)   The distribution of area upon which pressure is applied.

36.   Explanation of the above factors will be made during the following description of forming dies shown in Figs. 6 to 9.

37.   Fig. 6 illustrates a simple forming die in which vertical pressure is distributed uniformly over all top and bottom surfaces of the casting. The semi-circular section in the lower die is relieved to prevent the casting sticking on die.

38.   Castings can be formed accurately in this type of die if all warpage or deformation occurs only in the direction of squeeze. There are no means of squaring up a casting which may have been deformed out of square.
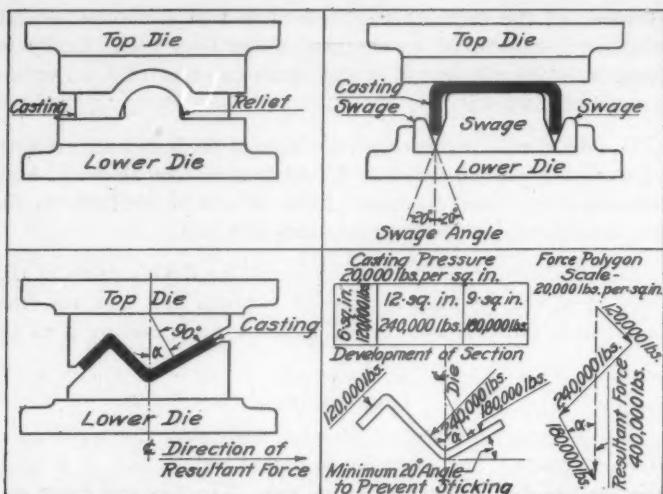
FIG. 6—UPPER LEFT—FORMING DIE SHOWING UNIFORM DISTRIBUTION OF PRESSURE ON CASTING. FIG. 7—UPPER RIGHT—FORMING DIE WHICH APPLIES A RESULTING HORIZONTAL PRESSURE BY THE APPLICATION OF A SWEDGE. FIG. 8—LOWER LEFT—FORMING DIE USED TO STRAIGHTEN CASTINGS HAVING ANGULAR SURFACES, SUCH AS A "Z" SECTION. FIG. 9—LOWER RIGHT—CONSTRUCTION OF A FORCE POLYGON USED TO DETERMINE PRESSURES AND THE RESULTING FORCE FOR DIE CONSTRUCTION SHOWN IN FIG. 8.

Application of Swedges

39.   Fig. 7 illustrates a forming die which applies a resulting horizontal pressure by the application of a swedge. Direct horizontal pressure on vertical casting surfaces can not be obtained with forming dies.

40.   Accuracy of castings straightened in this manner is not dependable, due to the inability of producing sufficient permanent set in the metal. Objectionable flat spots some times occur on castings where sufficient set is obtained. Channel or yoked sections which have sufficient tolerances are well adapted to this type of die.

Straightening Angular Surfaces

41.   Fig. 8 illustrates a forming die used to straighten castings which have angular surfaces, such as a "Z" section.

42.   Pressure applied to such surfaces is the resultant of the vertical pressure applied by the press. These resultant pressures must be applied so that there is a uniform pressure per square inch at right angles to the surface of the casting. The resulting

horizontal forces also must be in equilibrium so that the top and lower dies will center when pressure is applied.

Design

43.   In designing a die of this type, it is necessary to calculate the area of each surface contacted and multiply it by the pressure per square inch which you wish to apply, then draw a force polygon as shown in Fig. 9. The resulting force will give the amount of press pressure required. It also will give the direction in which the force should act in relation to the casting to obtain the pressures required, and at the same time hold the die in equilibrium.

44.   Horizontal angular contact should be greater than 20° to overcome stick angles and permit the die to slide into equilibrium.

Male and Female Swedges

45.   Fig. 10 illustrates a forming die which consists of a male and female swedge and is used to straighten light, springy castings, such as pedals, lever arms, etc.

46.   The female swedge serves as a lower die, and consists of
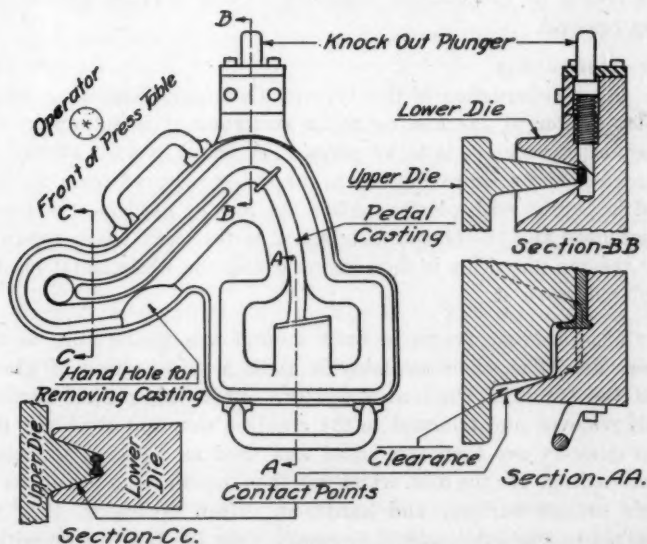
FIG. 10—FORMING DIE CONSISTING OF A MALE AND FEMALE SWEDGE AND IS USED TO STRAIGHTEN LIGHT SPRINGY CASTINGS, SUCH AS PEDALS, LEVER ARMS, ETC.

an impression of the casting to be straightened with side walls extending upward and outward from the contour of the impression. This acts as a swedging means for guiding and forming the casting to its proper shape.

47. The male swedge acts as a top die and is designed to fit the female swedge when die is closed.

48. Upright portions of the casting which cannot be formed with the lower die, are swedged to shape by suitable swedges built into the top die, as shown in section "*CC*," Fig. 10.

49. The lower die is not fastened to the press table, as it must be allowed to float, conforming with pressure within the die until it closes.

50. Sticking of castings within the lower die is an inherent characteristic with this type of forming die. A knockout plunger, such as shown in section "*AA*," Fig, 10, therefore, is placed conveniently under the approximate center of casting, where it can be struck with a hammer. Due to the springiness of the casting formed in this type of die, and the small amount of clearance allowed on the side swedges, the hammer blow is transmitted through the entire length of the casting, loosening it and making possible its easy removal.

Die Construction

51. Construction of this type of die is quite simple. A white metal master of the casting to be straightened is set up in the position in which it is to be pressed. A frame is built around it, which conforms approximately to the shape required for a top die and is poured with plaster. After the plaster hardens, the frame is removed and the plaster is trimmed to the exact shape required for the top die. This is then inverted and the white metal master replaced upon it.

52. Another frame is built around this conforming to the lower die. The whole assembly is given a thin coating of grease and the frame is filled with plaster. When this plaster hardens it is removed and trimmed to the required shape of the lower die. The plasters are then shellacked and used as patterns to obtain steel castings for the dies, which are then machined to templates on their contact surfaces and hardened. After hardening, they are checked to templates, and if necessary, they are touched up with a hand grinder.

SQUEEZE DIES

53.   Squeeze dies consist of a top and lower die, either or both of which are made up of a number of movable wedge shaped parts. These parts are machined in relation with one another to form a cavity. The sides of this cavity, which come in contact with the casting, straighten the casting when the die is closed. Springs usually are used to hold these wedge shaped pieces apart to provide clearance for loading and unloading the die. Actuating and holding means are provided to keep the moving parts in relation with one another so that the entire die can operate as a self-contained unit.

Field of Use

54.   Squeeze dies are used when it is necessary to apply both horizontal and vertical pressure to a casting at the same time. Since presses used in the foundry are only capable of delivering vertical pressure, it is necessary to develop the horizontal pressure within a die.

55.   This is done by utilizing the vertical force within the press, and at the same time converting this vertical force into horizontal forces, by the application of the wedge principle within a die. This type of die design permits pressure to be applied to the casting in any direction at the same time during one cycle of press operation.

Mechanical Advantage

56.   The wedge has a mechanical advantage similar to the lever, pulley or screw. This mechanical advantage, neglecting friction, is equal to the distance through which a force is applied, divided by the distance through which the resulting force acts; or in other words, the depth of wedge divided by the width of the wedge.

57.   If the depth of the wedge is 10 in. and the width of the wedge is 5 in., the mechanical advantage is 10/5 or two. This means that a press capable of delivering 500 tons pressure vertically could be made to deliver 1000 tons horizontally with the proper equipment.

PRESSURE FORMULAE

58.   Fig. 11 illustrates the mechanics of the wedge principle and shows formula for horizontal and vertical pressures with and without friction.

59. Friction must be considered in squeeze die design. The common fault of die sticking is due largely to friction and, since the pressure per square inch on sliding parts within a die amounts to 4000 or 5000 lb., the coefficient is very high.

60. The coefficient of friction is defined as the tangent of the angle φ at which a body begins to slide down an incline.

61. When the width of wedge divided by the depth of wedge is less than the coefficient of friction, we have a stick angle. Such a condition within a die would give trouble unless sufficient spring pressure was put under the wedge to overcome the difference between the proper and improper angle.

62. Through experience, we have determined this friction or stick angle to be 15°. This means that 27 per cent of the horizontal power is absorbed by friction. We have always had trouble with dies having included angles of less than 15°.

Design Considerations

63. Before a squeeze die can be designed, the following analysis must be made of the casting to be squeezed:

a. After pattern construction and foundry variations are taken into consideration, determine where the pressure
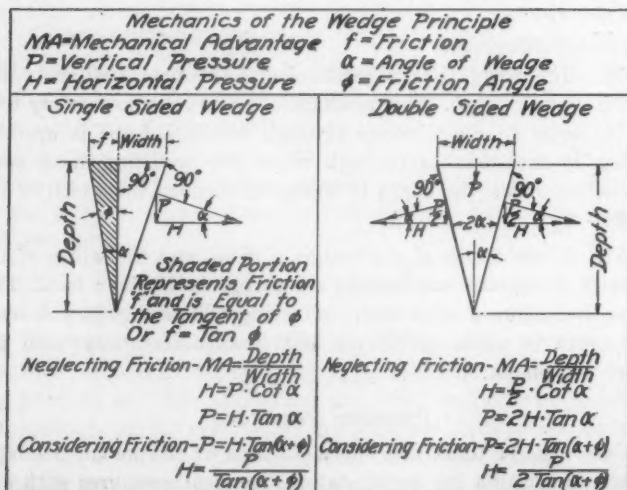
FIG. 11—ILLUSTRATES THE MECHANICS OF THE WEDGE PRINCIPLE, AND SHOWS FORMULAS FOR HORIZONTAL AND VERTICAL PRESSURES WITH AND WITHOUT FRICTION.

should be applied to the casting to put it into shape and hold
it in relation to customers locating points.

  b. Determine the direction of these pressures and try
to combine two or more which can be applied with one moving
part.

  c. Determine pressure required to straighten the various
sections of the casting to be squeezed by applying the pressure
formula contained in Fig. 4.

  d. Calculate the total pressure requirements to straighten
a casting, and design die to distribute these pressures so that
they will come within the capacity of the press and operate
with the least vertical slow travel of the press as is possible.
30° wedge angle is good average practice. 40° wedge angles
have been used successfully on certain classes of work, where
the die actuates the wedges for horizontal pressure. When
wedges are actuated by pressure applied through a casting,
attention must be given to the distribution of pressures so that
the proper wedge angles can be selected. More pressure must
not be applied through a casting than is required to coin or
straighten it.

Maximum Pressure

  64. The maximum pressure which is practical to apply upon
a casting in a squeeze die is total press pressure vertically and
twice the total press pressure horizontally.

  65. If certain parts of a casting require more pressure than
the press is rated, apply horizontal pressure to these parts. This
may enable you to increase the capacity of your press 100 per cent.

Distribution of Pressures

  66. Distribution of press pressures within a die is shown in
Fig. 12. Formulas for calculating pressures are derived for each
case.

  67. Dies designed to obtain maximum vertical press pressure
operate the squeeze mechanism through the casting straightened,
as in Figs. 2 and 3.

  68. Dies designed to obtain maximum horizontal press pres-
sure operate the squeezing mechanism through the top die without
contacting the top of casting, as in Fig. 6.

69. The maximum vertical pressure on a casting is reduced by the amount required to operate a squeezing mechanism, such as in Figs. 5, 7 and 8.

Accuracy and Maintenance

70. Accuracy and maintenance of forming dies are dependent upon the proper consideration of the following factors in die design:

1. Elimination of dirt and abrasive material in moving parts.

2. Proper selection of construction materials.

3. Adjustment of wearing parts.

71. Castings are either milled, sand or shot blasted before being straightened. The department for doing this work is called the cleaning department in a foundry. The term "cleaning" is a misnomer. Castings which leave the cleaning department contain sand, steel shot, chips and abrasive material which put extreme hardships on press and die equipment used to straighten castings. Considerable thought must be given to the design of a die to eliminate the admittance of dirt and abrasive materials in the moving parts, even if particular attention is paid to the proper cleaning of castings with the equipment used.

72. Whenever possible, a die should be designed so that the
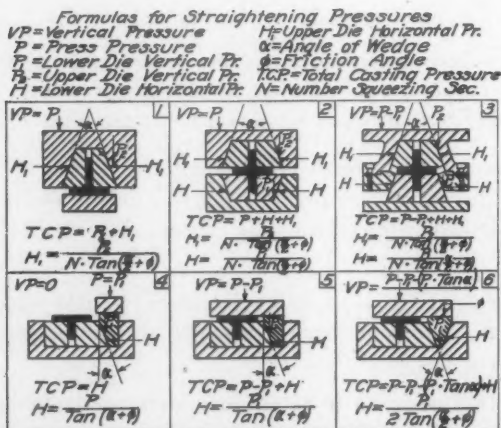
FIG. 12—ILLUSTRATES THE DISTRIBUTION OF PRESS PRESSURE WITHIN A DIE, AND GIVES FORMULAS FOR CALCULATING THESE PRESSURES.
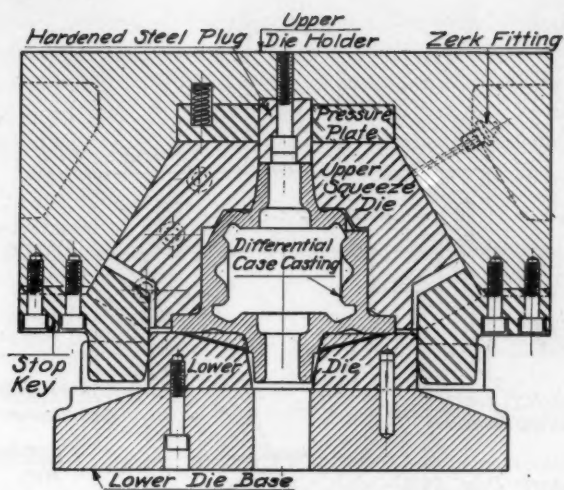
Fig. 13—Differential Case Squeeze Die.
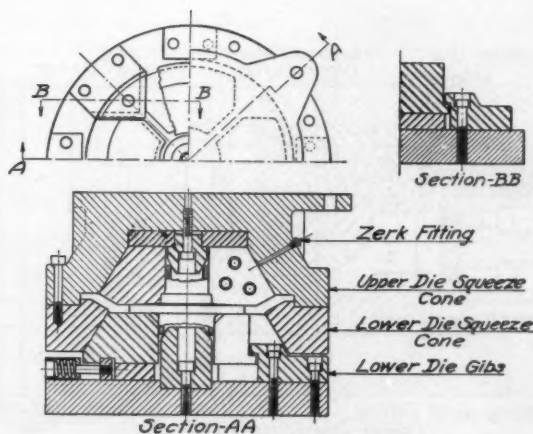
Fig. 14—Front Wheel Hub Squeeze Die.
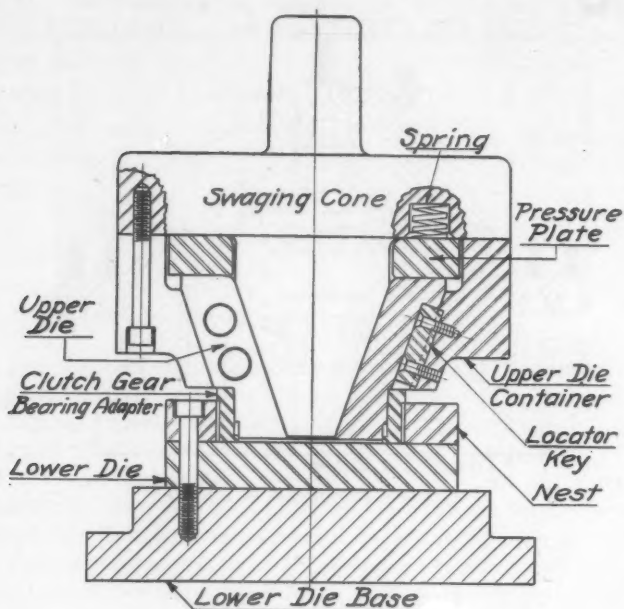
FIG. 16—EXPANSION SQUEEZE DIE.
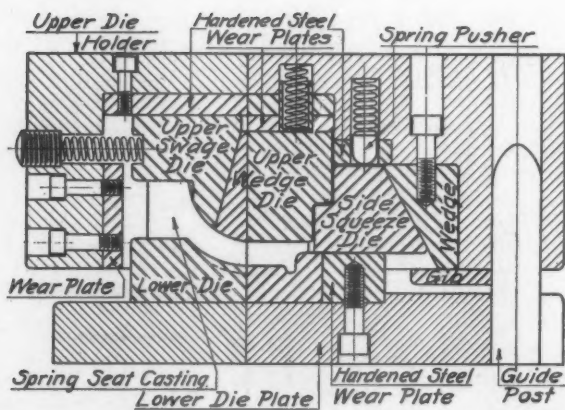
FIG. 15—SPRING SEAT SQUEEZE DIE.

top die will contain all the moving parts. Keep the lower die free of deep pockets and moving parts, if possible, so that it can be blown out readily. When lower dies must contain moving parts, actuate them from the top die so that they will not be enclosed in a container. Provide clearance spaces under the moving parts in the lower die so that dirt will have a place to go and can be blown to the outside of the die.

73. Figs. 13 to 16 show typical illustrations of squeeze dies.

Selection of Die Materials

74. Materials used in die construction are dependent upon the use to which a particular die part is subjected. This is usually a factor of pressure per square inch of surface, wear and breakage.

75. Die parts which contact the casting over considerable area and are not subjected to pressures over 30,000 lb. per sq. in., can be made of cast steel containing 0.35 to 0.40 per cent carbon and 1 to 1½ per cent nickel. This steel has a tensile and compressive strength of approximately 113,000 lb. per sq. in. when carburized at 1650°F., quenched in water and drawn at 325°F.

76. Die parts which are subjected to more pressure due to irregularities in castings, heavy production, or coin operations, must be made from a better material. When the sections are rugged and contact surfaces can be ground, a good grade, carbon-vanadium tool steel made of good base material, free from sponginess and non-metallic inclusions, with the following composition, gives good results:

	Per Cent		Per Cent
Carbon	0.66-0.75	Phosphorus	0.025 max.
Manganese	0.10-0.30	Silicon	0.20-0.35
Sulphur	0.025 max.	Vanadium	0.15- .25

77. A small amount of vanadium added to the steel toughens it and causes it to harden more uniformly with less distortion, compared with carbon tool steels. This steel has a compressive and tensile strength of 300,000 lb. per sq. in. when hardened in water and drawn at 400°F. to 58 brinell.

78. When the above die parts are not rugged nor ground, a steel having the following composition gives the best results:

	Per Cent		Per Cent
Carbon	0.95	Chromium	0.50
Silicon	0.22	Tungsten	0.50
Manganese	1.30	Vanadium	0.10

This steel has a tensile and compressive strength of 300,000 lb. per sq. in. when hardened in oil and drawn at 400°F. to 58 brinell.

79. Pressure plates, and surfaces over which moving parts slide, should be calculated to enable the selection of material to withstand the pressures and wear to which they are subjected.

80. Good results are obtained in die construction when allowing 5,000 lb. per. sq. in. for soft machine steel; 30,000 lb. per sq. in. for carburized machine steel; and 100,000 lb. per sq. in. for hardened tool steel.

HAMMER DIES

81. Drop hammer dies were developed to improve hand straightening methods, and were the forerunner of modern press dies.

82. Drop hammers are used chiefly for striking dies, although some squeeze dies have been used successfully. Hammer dies must be built stronger and of better material than press dies, and parts must be held together more securely, as the force of the blow tends to loosen them.

83. The energy of a hammer is equal to the weight of the hammer multiplied by the height through which it falls. The energy of the falling ram of a fast travel hydraulic press is equal to the weight of the ram multiplied by its velocity at the end of the fall. The difference is due to the fact that the press ram is not a free falling body.

Force of Hammer Blow

84. The average force of a hammer blow is equal to the number of inch-pounds of energy divided by the amount of penetration plus the weight of the falling body.

85. The efficiency of the force of blow is dependent upon a number of variables, such as:

1. Imparting motion to body struck.

    2. Penetration against friction.

    3. Resistance to shear or deformation.

    4. Crushing or heating of both the falling body and the body struck.

86. The distance through which these resisting forces act is generally indeterminate. Experiments conducted by R. H. Thurston on pure copper plugs and A. W. Moseley on bessemer steel plugs showed efficiencies of 90 per cent and 70 per cent respectively. Further tests by W. T. Sears showed that after a certain velocity of the falling weight had been attained, the speed had little effect on the compression of a lead plug, which he used in his experiments. This speed was fixed at 10 feet per second, but its exact value is uncertain. This speed would be equivalent to a drop of approximately 19 in.

87. The speed at which the force is applied with a drop hammer is undoubtedly the reason for the surface effect of the blow on castings. If the force could be applied slower, the pressure would penetrate to the center of material as is characteristic with hydraulic presses.

Examples

88. Fig. 17 shows a typical hammer die used for striking bearing caps. The solid construction of die blocks and holders
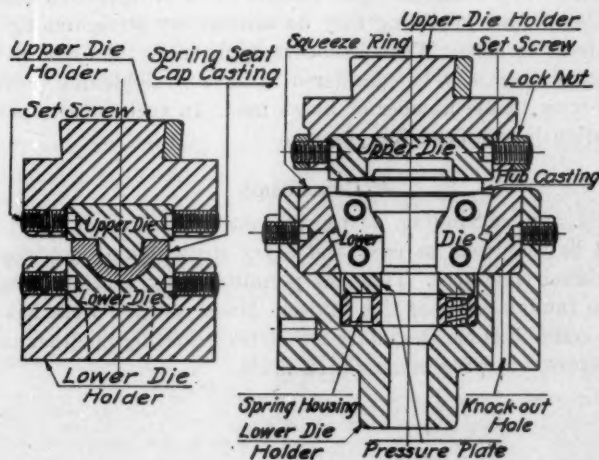
FIG. 17—LEFT—SPRING SEAT CAP HAMMER DIE. FIG. 18—RIGHT—HUB HAMMER DIE.

should be noticed, as well as the method for holding die blocks in die holder.

89. Fig. 18 shows a hammer squeeze die construction for straightening hubs. The squeeze ring has a solid stop in it to correspond to a similar stop on the squeezing sections. When the squeezing sections are forced into the squeeze ring by collapsing the springs between the sections, they will remain in place until the springs in one section are broken.

Disadvantages

90. The disadvantages of drop hammers for straightening castings are:

    (1) Limitation to flat symmetrical castings.

    (2) Difficulty in obtaining a solid flat strike.

    (3) Fatigue produced by noise.

    (4) Difficulty in setting dies accurately.

    (5) Maintenance cost of equipment.

Advantages

91. The advantages to drop hammers might be summed up by mentioning:

    1. Speed of operation.

    2. Unlimited straightening power.

92. Drop hammers have a short cycle of operation and are considered fast, providing they do satisfactory straightening with one strike or blow. When hammers apply more than one blow on a casting, they may be considered to have straightening power in proportion to the number of blows used. In such a case speed of operation is sacrificed for power.

CONCLUSION

93. In concluding, many interesting designs of equipment could be discussed, as each and every straightening job requires individual treatment. Time has permitted only a brief discussion of the fundamentals of this subject. Much research work has been done and numerous ideas have been tried, resulting in either success or failure, but progress has been made.

Report of Committee on Foundry Sand Research[1]

By H. Ries[2], Ithaca, N. Y.

Members of the American Foundrymen's Association:

During the past year several worth while pieces of work have been accomplished by your Committee on Foundry Sands Research.

Revised Edition of Sand Testing Standards

There has been prepared a new edition—the fourth—of the pamphlet on Standard Tests of Sands and Clays. This edition contains a number of new tests, and is the first to include testing methods for foundry clays. The Association is to be complimented on the attractive form in which it has been issued, and I wish to thank the members of the Committee on Standard Tests for the assistance which they have given freely and generously.

Properties of Sand

There has also been completed the manuscript of a separate pamphlet on the properties of sands, which may serve as an introduction to the testing pamphlet for those who desire information on this subject. In the preparation of this, Messrs. Dunbeck, Dietert, Randall and Reichert, as well as the late Mr. Aptekar, have given much assistance by the preparation of special chapters.

High Temperature Investigation

A most important line of work which has been initiated has been a study of the properties of steel sands at different temperatures. This work was recommended by the Sub-Committee on Steel Sands and the Board of Directors of the A.F.A. made special provision for this work.

The work outlined consisted in selecting a series of steel sands and making up mixtures of these with bentonite, fire clay, cereal, silica ester and oil binders. These sands are to be tested with one binder alone, and also with a combination of binders. These mixtures are to be tested for their compression strength and expansion at different temperatures up to 2800°F. if necessary.

The Sub-Committee on Steel Sands at its meeting in Milwaukee a year ago, prepared detailed instructions for the work

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with the understanding that some of these might be subject to some revision as the experiments proceeded, and some change was later found necessary. The work is being carried out at Cornell University, and the A.F.A. has designated H. L. York to conduct the experiments. As in all such cases, it is sometimes difficult to foresee troubles which may arise, and this particular investigation has not been free from them. It was the intention at first to utilize an electric furnace for the experiments and one was acquired, but for various reasons which need not be mentioned here, this type of furnace was found to be unsatisfactory and after some experimenting, Mr. York designed one which is heated by natural gas. This furnace is giving good satisfaction, not only because the desired temperatures can be reached, but it is easily controlled, and the temperature in the testing zone remains constant and uniform throughout that portion of the furnace where the sand specimen is being tested. While considerable delay was caused, at the start, due to our troubles with the electric furnace, and certain modifications which had to be made in the gas heated furnace, the work is now proceeding in a satisfactory manner. It seems likely that when the investigation of steel sands is finished, that it may be desirable to extend the work to sands of other types.

We believe that it is fitting to express our appreciation to the authorities of Sibley College of Mechanical Engineering for the facilities, materials and use of mechanics which they have put at our disposal in the course of the work.

Future Work

Lest it be thought that this investigation finished the work which the Sand Test Committee has in mind, it may be well to mention a few suggestions that have been made for future work.

More should be known regarding the permeability of sands at different temperatures, and the relation of hot to green permeability.

The deformation of green sands under pressure is known to take place, but this property has not been studied in detail. It should be investigated in connection with the other properties of sand. One German investigator has claimed that if sand deformed under a pressure of less than 85 per cent of its compressive strength, the sand is not to be regarded as suitable for molding purposes. This point should be investigated.

Deformation may be related to flowability and the factors governing the latter have still to be determined; in fact, we have as yet no standard test for this property.

Strange as it may seem, there are two properties of sand whose influence on other characters of the molding sand is not definitely known. One of these is the effect of grain shape on such properties as strength of the sand. To carry on this work means not only testing the effect of different shaped sand grains, but the work must also be done at first with grains of uniform size.

The other is the relation between particle size, composition, and bonding value of clays containing particles of less than 20 microns. In this connection a moment's thought will, we believe, indicate to us that we might have two naturally bonded sands having the same grain distribution, the same A.F.A. clay content, but different strengths. This must obviously be due to the character of the bond, and right here is a great field for study. It may even be found that the minerals in the clay bond exert an influence.

Durability is another property which thus far seems to have eluded the efforts of those who are endeavoring to develop a laboratory test which correlates with foundry practice. Incidentally, a study of the bonding material may throw light on the cause of durability.

Several suggestions have been received that there should be a satisfactory method for the evaluation of bentonites, as it would be of value to producers and consumers.

It will be obvious, we believe, that some of these proposed investigations will not only take much work and also funds, but the time and money that may be spent on them will be well worth the results of practical importance that are likely to be obtained.

Respectfully submitted,

H. RIES, *Technical Director*

DISCUSSION

MEMBER: What is the present status of the knowledge on the question of particle shape:

DR. RIES: So far as I know, there is no unanimity of opinion. Some persons believe that round grains give greater strength and others say the sharp grains give greater strength.

MEMBER: Why should there be a question between round grains and angular grains when it has been accepted that the round grains give more strength and more permeability?

DR. RIES: This question has been raised and several have suggested that we ought to investigate it. Do angular grains or round grains give the greater strength? Some years ago some tests were made and they seemed to indicate that angular grains will hold the material together better and, therefore, give greater strength. As a matter of fact, round grains are very rare. Most sand grains are angular or sub-angular.

CHAIRMAN HARRINGTON: I believe that many of our previous conceptions have changed and it is quite possible that our conception in regard to this problem of grain characteristics may have to be changed in the light of further research.

MEMBER: What effect does the grain shape have on permeability?

DR. RIES: Merely expressing a personal opinion, I would say the round grains might help to increase the permeability, assuming that they are all the same size.

MEMBER: What is Dr. Ries's opinion as to when his committee will start studying durability.

DR. RIES: This question of durability has been tackled by several different investigators and the committee is hoping to develop a laboratory test to determine the durability of the sand. Then, when we get a laboratory test, we will want to have one that will check up somewhat with practice.

Different tests for durability have been suggested so far. It may be remembered that Professor Schubert of the University of Illinois gave a paper[3] on that subject and none of the three tests seemed to be in agreement. I suppose that the safest way to do is to ultimately make tests which will simulate as closely as possible the foundry conditions. We have a sub-committee on durability.

MEMBER: In discussing the round and the angular grains, we hear a great deal about the angular grain sands of Pennsylvania and of the rounded grain sands of Illinois. The sand producers have their ideas on the amount of bond to use to develop maximum strength in each of these types of sand. How is the committee going to issue a report and what effect is it going to have on these two groups, with their diversified opinions as to what these different sands will do.

DR. RIES: To begin with, I do not believe the committee would want to issue a report which might appear to be ultimately favoring one producer as against another producer. In such an investigation, an attempt will be made to determine the effect of the round grains and the angular grains. Now, in working out a problem of that sort, you have to start with grains of uniform size. That is a difficult matter to begin with, because in order to get grains of uniform size, one must sift a whole lot of sand and separate out the grains of one size. I have done something along that line and know some of the troubles.

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[3] Schubert, C. E., *A Correlation of the Physical and Chemical Properties of Clays with the Durability of Molding Sands*, Trans., A.F.A., vol. 45 (1937) pp. 661-688.

What Happens To a Core in a Core Oven

By Jasper Willsea,* Rochester, N. Y.

Abstract

In this paper, the author presents specific statements on the importance of water in drying and baking of a core in an oven. The action of water molecules is discussed with an explanation of how they cause humidity. How water affects green strength and green permeability during drying is explained. Water or moisture as affecting oven temperature control is discussed with the importance of dry oven air emphasized. The author then advances his theory of "even exchange temperatures." Ideals for fast and economical drying are presented. A discussion of volatile matter follows. Finally, the author advocates coke as the ideal fuel.

1. The American Foundrymen's Association already has covered the standardization of core sands and core binders with scientific thoroughness. However, water, a naturally standard material, has received little attention by most investigators. This paper shows that, during the baking of a core, water is the one big problem.

2. All facts in this paper on the make-up of cores are from average, general, foundry practice, and should be well known to every foundryman. We shall consider only oil sand cores to get a clearer picture of the entire water problem.

3. *An Experiment Which Shows the Water Problem:* A thermometer bulb in the center of a core during baking shows a rise in temperature until 212°F. is reached. The temperature remains at 212°F. while there is any water. As soon as the temperature rises above 212°F., it is known that all the water is gone. The length of time the core is held at 212°F. depends on how soon all of the water is dried out of the core.

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4. *Hardening the Core Is Fast and Automatic:* Hardening is fast and automatic as soon as the water is gone. An oven which bakes cores faster than others, is simply an oven which takes out all of the water quicker than others.

WATER MOLECULES

5. *Heat Energy in Water:* The smallest particles of water are molecules. Each molecule has more or less heat energy, although the effect is too small to see. Ordinary water actually is a seething mass of energetic molecules in constant motion. The surface has a lot of peaks that are rapidly rising and falling.

6. *"Drying" or Evaporation:* The molecule on top of a high peak is six layers above the surface layer. Each molecule on top of a peak is the one with the greatest amount of heat energy. It pushes itself away from the molecules that are clinging to it, and jumps out from the water. Any moist surface, such as the water coating on the sand grains of green cores, has this action going on constantly.

Humidity

7. *Moisture in an Oven, or Humidity:* Moisture in an oven is actually separate water molecules full of heat energy, bouncing into one another, and too energetic to stick together.

8. *Moist Surface Humidity:* The most energetic molecules are continually jumping away from the water. Many of them bump into molecules already in the air, and the first molecules are bounced right back into the moist surface. Very close above any moist surface, there is a thick mass of water molecules. The thick mass thins out as it extends farther and farther away from the surface.

9. *General Humidity:* The water molecules which, by chance, avoid being bounced right back into the moist surface, finally jostle their way out into the oven space. It is still more difficult for water molecules to get into the oven space when the space itself has become crowded with water molecules. If the oven space is kept as free of water molecules as possible, there will also be less crowding of molecules close to the moist surfaces of the sand grains.

10. *Saturated Air at a Constant Temperature:* A cup placed upside down, with the rim sealed against the water surface, is soon entirely filled with the same thick mass of water molecules that is found very close to any moist surface. When the cup space is filled, or saturated, no more molecules can be evaporated.

11. *Saturated Air at Increasing Temperatures:* More molecules can be driven into the cup space if the temperature is raised but these added molecules will get in very slowly because the space already is almost full.

12. *Drying Speed and Relative Humidity:* The drying speed of the general oven air, at any particular temperature, depends upon its relative dryness. The relative dryness is the amount of moisture actually in the air in relation to the moisture that the air can hold if it is saturated at that temperature.

13. *Dew or Condensation:* When moist air is cooled down, there is less heat energy in the water molecules so that one molecule can stick to another. Beads of moisture on cold surfaces are gatherings of molecules that have lost so much heat energy they can no longer bounce around in the air.

FACTORS AFFECTING PROPERTIES

14. *Green Strength from Surface Tension:* Water has much surface strength, or surface tension. Water coatings on the sand grains are drawn together stronger at the sand grain contacts than these coatings are drawn to their own grain surfaces. This makes a fillet of water around each sand grain contact to give green strength.

15. *Permeability:* It should be kept in mind that core permeability is an accurate measure of the total of clear paths for air to pass all the way through the core and not a measure of the unoccupied, or air space.

16. *Green Strength and Permeability:* The sizes of the air passageways through the green core are actually smaller than in the dry core because of the green strength water fillets drawn up at the sand grain contacts. The permeability of a green core is usually 80 per cent less than when it is baked. Air-drying an oil sand core outside of the oven usually gives less strength because the water fillets are evaporated away from the sand grain contacts. Permeability is greatly increased in relation to the total air space, when all of the water is evaporated out of a core.

17. *Irregular Sand Grains and Permeability:* Normal permeability is cut down greatly in relation to the total air space inside the core by very irregular sand grain shapes. Occasional openings are blocked off entirely by these irregularities, which leaves fewer paths for the passage of air or steam clear through

the core. The size of the clear paths is cut down still more by the
water fillets which are drawn up more readily between irregular
sand grain shapes.

18. *Very Small Sand Grains and Permeability:* The total
amount of sand grain surface is increased by choked-in grains
which are smaller than the normal size. Additional water and
core oil are needed to cover the small grains. The green strength
of such cores is greater, but the permeability is much less because
of the increased number of water fillets.

REMOVING MOISTURE AND BAKING

19. *Air Space Inside the Core:* The free air space inside a
core, in the green state, *is about five times as much as the space
taken up by the water and the oil binder.* This shows how rela-
tively thin the coating of water and binder is on each grain.

20. *Getting Water Out of the Core:* All of the air spaces
among the sand grains will constantly be filled with evaporated
water molecules. The capacity of the air spaces near the core
surface for holding water molecules is greatly increased by a high
oven temperature.

21. Molecules from air spaces near the surface of the core
go out into the drier oven air. Then molecules can leave the sand
grain coatings in order to keep these air spaces filled. Molecules
farther inside the core are constantly moving toward the warmer
air spaces nearer the surface. These spaces are thirstier because
they can hold more at their higher temperature.

22. *Drying Speed:* The relative dryness, or thirstiness, of
the general oven space controls the speed at which water mole-
cules go out into the oven air. A constant supply of dry air evap-
orates moisture fast at low temperatures. Drying slows down and
stops as soon as the oven is filled with moisture if the air is not
replaced by dry air.

23. *Fast Baking:* Green strength is almost entirely the sur-
face strength of the water fillets which draw the grain together.
Core baking, after removing all water, is actually a natural change-
over to the cementing effect of the oxidized binder. The change-
over may just as well be done as fast as possible, if it can be done
without disturbing the original green strength, or disturbing the
final strength of the hardened binder.

OVEN TEMPERATURE CONTROL

24. *The Two Things Affecting the Core Most:* There are two big effects on the core while baking — one is from the dryness, or relative humidity of the oven air near the core; and the other is from the oven temperature.

25. *The Oven Temperature:* The temperature of the oven is usually held just below the temperature at which the binder is burned. 440°F. is considered a safe temperature for most oil finders. We must also think of the effect 440°F. would have on the green strength water of the core.

Boiling

26. *The Oven Temperature and the Boiling Point:* Four hundred and forty degrees F. is twice the temperature that boils water, so we must see what boiling is really like, and what it does to a core.

27. *What Boiling Is:* Boiling is caused by having more than enough heat put into water than is needed for holding its temperature at 212 degrees. The excess heat input energizes the water into steam at 212°F. This excess energy is shown by the bubbling disturbance.

28. The total energy of all the molecules being vaporized at 212°F. is an outward force equivalent to the pressure of the atmosphere. Bubbling disturbs all of the water because the energetic molecules inside the water are no longer held down by the weight of the atmosphere.

29. *The Core Water and Bubbling Disturbance:* Bubbling disturbance moves water away from the source of heat. Bubbling is harmful to the green strength of the core by breaking up, and moving farther into the core, the green strength water fillet bindings which had been drawn up around the sand grain contacts.

30. *The Core and Bubbling:* There is considerable amount of water crowded together when the water has been moved toward the inside of the core away from the very hot oven air. This makes an inner layer of very wet sand, and then the outside of the core, from which that water has been driven, is a dry crust. The greater wetness softens this inner layer, and the pressure of vapor being made there loosens up a scab.

31. *Scabbing and Bubbling Disturbance:* The scabbing effect usually is not sufficient to show an actual scab on the core. The weakness is there, however, in the form of a scab. It will show up in the mold during pouring when the hot metal causes gas pressure, or even steam pressure, inside the core.

32. *Bubbling Disturbance Unnecessary:* Evaporation goes on, from the very beginning of the baking, at any temperature below 212°F. The evaporation speed depends upon the dryness, or thirstiness, of the oven air into which the core moisture must finally go. Most of the water in any part of the core should have been passed out of the core by evaporation by the time that part has been heated to 212°F.

33. *Boiling Temperature in the Core is Very Much Delayed:* Sand takes only one-fifth as much heat as the same amount of water. There is a much larger amount of sand in the core so it takes almost eight times as much heat as the small amount of water. The water coatings farther and farther inside the core are thereby prevented from reaching 212°F. too quickly. This gives a big margin of safety to delay any bubbling in the core, thereby giving time for safe evaporation.

Baking

34. *The Baking Temperature:* The time needed to heat up the core depends on the temperature in the oven, and the thickness of the core from its plate. A high oven temperature, safe for the binder, in some cases cannot be used because certain parts of the actual oven in question are apt to be at a higher temperature than the binder can stand.

35. *An Even Temperature by Controlling the Baking:* The amount of heat used to energize the water molecules into vapor is so great that it causes large temperature drops around every core. In addition there is a relatively small amount of heat per minute being used up for heating the cores. The big problem for fast drying is to get rid of this wet, cooled-down air. If this air is kept separate and taken out of the oven immediately, the temperature of the general oven air is held evenly. More heated air is simply pushed into the oven.

Drying

36. *Stalling the Baking with Wet Oven Air:* If the oven air is allowed to become filled with moisture, the drying becomes

slower and slower, and then stops, regardless of the oven temperature. The oven is then like a boiler and the cores are actually soaked in steam.

37. *Steam Cooking and Burning at the Same Time:* Bubbling action makes a dry crust on the outside of a core, even in a steam-soaked oven as described above. The dry crust is burned if the oven temperature is too high for the core binder, even while the inside is being steam cooked. Steam pressure is created inside the core from the heat of the crust which acts on the moisture just inside the crust. This would cause an explosion if the permeability of the core crust did not relieve the pressure.

38. *Drying Large Cores:* Water molecules from the inside of large cores must be coaxed out through the long, crooked permeability passageways. The oven air must be kept especially dry to get drying speed for large cores. The oven temperature must be held low enough to avoid burning the dry crust.

HEAT NEEDED FOR DRYING AND BAKING

39. *Heat Needed for Drying:* The *small* amount of water in the core needs as much heat *to get into the oven air by evaporation* at 212°F. as the heat required to get the *large* amount of core sand up to the same temperature.

40. *Heat Needed by the Core While Actually Baking:* Heating up the core sand only, from 212°F. to 440°F., *after* the water is gone, is the largest single amount of heat used. This takes 80 per cent as much heat *as the total of all the heat for heating the core sand and the water to 212°F., plus the heat for evaporating that water into the oven air.*

41. *When Heat is Used Most During Baking:* The large amount of sand takes a very much larger total amount of heat, but the sand takes up heat slowly, and always at the same speed for a particular temperature. Core moisture goes into the oven air freely if plenty of dry air is put through the oven. As a result, the greatest amount of heat per minute is used when the oven doors have just been closed, because of the heat furnished for evaporation. As time goes on, the amount of heat used per minute will steadily decrease because the moisture being evaporated must be coaxed from farther inside the core.

DRY OVEN AIR

42. *Plenty of Dry Air Preserves Core Strength:* An im-

portant point is that much drying can be done without having to heat up the core moisture. This avoids damage to the green strength, or the baked strength of the core. Any disturbance to green strength water fillets cuts down the normal baked strength of the core. Bubbling at the sand grain contacts breaks up the final hardened bond of the oil binder. By evaporating the moisture into dry oven air at low temperatures, the binder cannot be harmed.

43. *Dry Oven Air Cannot Harm the Core Mixture:* There is twice as much water as there is oil binder in the mixture which coats each sand grain. Drying water out of this mixture, which is spread out thinly on grains of sand, is entirely different from drying the natural water out of wood, for instance. In a lumber dry kiln, the air must not be too dry or it will shrink up the pores on the surface of the wood, and actually prevent getting any more moisture from inside the wood. Very dry air is not harmful in any way for getting the water away from the core oil.

44. *Dry Oven Air Helps the Core Mixture:* Very dry air actually is best because all of the water is removed away from the oil easily and quickly. Then, the hardening or oxidation of the binder starts sooner, and goes ahead faster. Water which lags behind with the oil because the oven air is moist, is harder to get out because the oil, as it hardens, becomes thicker and thicker from oxidation at higher temperatures.

45. *The Heat Needed for Evaporation:* The total amount of heat needed to energize water into vapor molecules, is almost *seven times,* the amount of heat needed for raising the water temperature from 70°F. to 212°F. This big demand for heat to vaporize core moisture means that air will give up its heat and take up water molecules very rapidly if there is room for more water molecules in the air. The oven air close to a green core is rapidly cooled down because that air is filling up with moisture.

EVEN EXCHANGE TEMPERATURE

46. *An Éven Exchange of Heat for Moisture:* The most efficient use of oven air is to give up enough heat for the evaporation and have that air completely filled with moisture at the temperature to which it has been cooled. This is the "even exchange temperature." The heat of the air has been exchanged for the equivalent of energized water molecules. The air is then saturated at this even exchange temperature.

47. *The Even Exchange Temperature:* Assume that ordinary room air, with its very small average amount of moisture, is in the core oven at 440 degrees. The even exchange temperature at which the heat from that air is given up for all the moisture that the air can hold, is a little higher than 220°F.

IDEAL FOR FAST DRYING

48. *Maximum Drying Speed:* An oven which handles air according to the even exchange temperature is ideal for fast drying. There is 117 times as much actual moisture in air saturated at 220°F. as there is in average air. A constant supply of dry air in the entire oven is fastest for coaxing water molecules out of every core.

49. *The Condition of Even Exchange Air:* The oven air that has this even exchange temperature of about 220°F. has two important advantages: First, it is much heavier than the general oven air; second, it is filled up with moisture at that temperature.

50. *First, The Heaviness of Even Exchange Air:* While the air is very close to the core, it is getting heavier because its temperature is getting lower. This air actually is falling down from the sides of the core during the exchange of heat for moisture. Heaviness is most desirable because it automatically separates the air which is 100 per cent used up from the remaining oven air which has not been used at all.

51. *Second, The Moisture of Even Exchange Air Should be Under Pressure:* The excess pressure of saturated vapor at 220°F. over atmospheric pressure causes these energetic molecules to expand as one thick mass from the used air into the unused, hot, dry, general oven air. This, accordingly, cuts down the usefulness of the entire oven. Therefore, a core oven under a low pressure holds the evaporated moisture within that air which first took it up and prevents evaporated molecules from spreading throughout the oven.

52. *Continuous Even Exchange Drying:* While the even exchange air is falling down from the core, the next nearest general oven air immediately closes in around the core.

53. *All Heat is Used:* A certain amount of water vapor molecules will pass out from the even exchange air into the next general oven air, simply because of the relative dryness. Also,

heat is always being conducted from the general oven air into the air close against the core that is actually being cooled down for evaporation. The general oven air is constantly giving up heat and taking up moisture while on its way toward the core. Therefore, dry, hot oven air is able to fill itself with moisture during the time it moves to the core and then falls from it.

54. *Raising the Core Temperature During Fast Drying:* The core draws a comparatively small amount of heat to the core itself from the general oven space. This small amount of heat is being slowly conducted from one sand grain to another in order to raise the core temperature. All but this comparatively small amount of heat is used actually to fill up the air with moisture during the first part of the baking.

IDEAL FOR ECONOMICAL DRYING

55. Handling the oven air at the even exchange temperature is best for heat economy. After being saturated at 220°F., the air which has been close to a core is pushed out of the oven at once. Then, the only heat leaving the oven is actually being used to hold all of the moisture in the air being pushed out. Dry, hot air is blown in the oven, whenever needed, until the oven is brought up to the desired temperature. Then the thermostat cuts down the flow of hot air to the small amount needed to make up for the radiation loss of the oven. The incoming air is heated to about the desired oven temperature. Then the volume of air being blown into the oven is practically the same as the volume of used, wet, cooled air, that ought to be forced out of the oven. This keeps up the fastest evaporation at all times. The incoming air forces out of the oven all used, undesirable air and prevents any delay which is the most costly thing in core baking.

VOLATILE MATTER

56. *Volatile Matter in the Core Oil:* The real usefulness of core oil is only from that part which becomes hard by taking up oxygen. There is also a considerable part of the oil which does not harden at the highest oven temperatures, and where there is plenty of oxygen.

57. *Volatile Matter Out of the Core Oil:* There are two ways by which volatile matter goes out into the oven from the core oil. One is by the simple distillation which goes on throughout the entire range of oven temperatures. A second is that at higher

temperatures, the distillation increases, and most of the volatile matter is oxidized. This oxidation is like actual burning because it makes unbearable, smoky fumes.

58. *Volatile Matter in the Hardened Binder:* Volatile matter comes from inside the binder itself so it should be removed as early in the baking as possible. Here again, any disturbance from bubbling at higher temperatures, cuts down the final baked strength. The volatile matter must not bubble because some of it is in the binder which is being hardened at the very points of contact between sand grains.

. 59. *Volatile Matter in the Oven:* At any particular temperature, and pressure, the core oven can hold only a certain total number of molecules regardless of how many different kinds of molecules there may be. If the oven air is relatively full of water vapor molecules, the distillation of volatile molecules is held back as well as is the evaporation of moisture.

60. *Delayed Volatile Matter Distillation:* It is most undesirable to have the distillation of volatile matter molecules delayed by an oven crowded with water molecules. It becomes more and more difficult for volatile molecules to get out of the binder as it becomes more viscous at higher temperatures.

61. *Delayed Distillation and Core Distortion:* The outside surface of the oil coating on each sand grain gradually becomes a tough film at the beginning of the actual baking. On the other hand, the higher temperatures energize any volatile matter molecules still inside the coatings so that they will try to bubble through the tough film. If these outer films are not tough enough, the core actually can be distorted. The matter which can be volatilized is more fluid at higher temperatures so that bubbling in the binder fillets at the sand grain contacts will cause the core to slump.

62. *The Strongest Good Core Uses Least Binder:* Any fluid matter left inside the hardened binder dilutes the baked strength. The strongest core has only hardened binder left in it, then; only enough core oil has been used to give the required amount of hardening material.

63. *Big Loss of Permeability and Too Much Gas Unless Volatile is Out:* Baked strength is lost when fluid is left in the hardened binder, and then bubbling action makes the hardened binder spongy. To get the desired strength in spite of these losses,

thicker coatings of binder must be used. Such cores have lower permeability and cause very much more gas in the mold, when poured.

The Thermo-Siphon Effect

64. The cold, heavy, moisture-filled air at about 220°F. should be allowed to fall naturally onto the oven floor. The oven floor should be the coolest part of the oven in order to prevent that wet air from getting any heat which would cause it to rise again into the oven space.

65. The incoming, dry, hot air floats to those places where heat is being used.

66. The pressure of the incoming air pushes the used air out of the oven immediately and helps to hold the moisture within the outgoing air; thereby, giving the oven a thermo-siphon effect.

Carbonizing the Core Binder

67. When the core oven is kept free of volatile matter and water molecules, the supply of dry air also furnishes plenty of oxygen for oxidation. If the oven temperature is high, and held evenly to avoid actual burning, the oxidation goes farther than simply to put oxygen into the binder. A large part of the hard binder itself will then be made volatile in the form of smoke. The binder is then carbonized, which makes it hard, and most of its strength is retained. This is the quickest and most economical way to bake the dark colored, strong, gas-free cores that all foundrymen like.

The Choice of Fuel

68. The importance of dry air for a core oven has already been shown. The fuel should be chosen accordingly. The largest amount of burnable material in any fuel is carbon, which, as all foundrymen know, becomes a carbon dioxide after burning. In most fuels, the next largest amount of burnable material is hydrogen. When hydrogen is burned, it becomes hydrogen oxide, and that is water.

69. A fuel which contains carbon only is ideal because it gives an absolutely dry heat. Coke, therefore, is the ideal fuel.

DISCUSSION

Presiding: R. F. HARRINGTON, Hunt-Spiller Mfg. Co., Boston, Mass.

A. H. DIERKER[1]: A few years ago we made some tests on core drying in which we actually measured the temperature inside the core while it was being dried and baked inside a large commercial core oven. Since our findings were somewhat at variance with the conclusions reached by the author of this paper, we felt it might be well to present some of the data at this time even though they have been previously published.

Fig. 1 shows the method used in making the various sized cores for test purposes. The lead wires from the thermocouples inside the cores ran to a potentiometer outside the oven. Two couples, placed within a few inches of each side of the group of cores were used to record the oven temperatures.
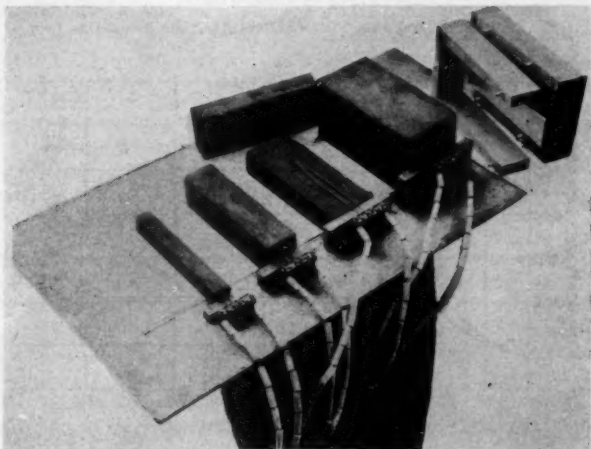
FIG. 1—THERMOCOUPLE POSITIONS INSIDE CORES.

Fig. 2 shows the readings of three thermocouples placed inside one 5 in. x 5 in. x 8 in. core which rested on a solid cast steel plate. It will be noticed that the temperature of the core rises rapidly to approximately 212°F. and holds constant until the water is evaporated then rises rapidly toward the oven temperature. It will be further noticed that the drying proceeds from the outside toward the center and that the core dries just about as fast from the bottom (where it is against a solid plate) as from the top. From this we would suspect very strongly that the nature of the atmosphere surrounding the core had little to do with the rate of drying. It would appear rather that the drying rate would depend largely on the rate of heat transfer to the core and, of course, heat can be transferred as easily through a solid plate as through the atmosphere.

[1] Research Engineer, Ohio State University, Columbus, O.
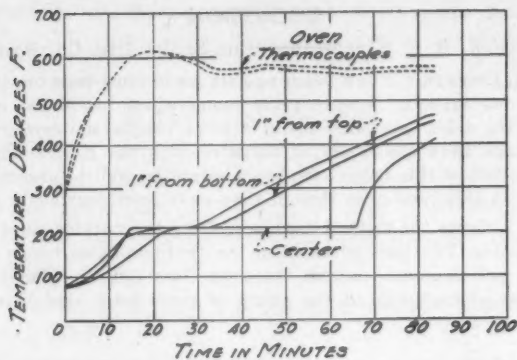
FIG. 2—RESULTS OF THREE THERMOCOUPLE READINGS INSIDE ONE CORE.
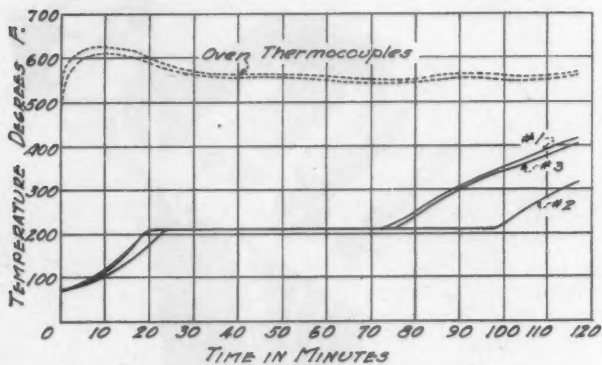
FIG. 3—EFFECT OF DRYING RATE ON CORE POSITION.
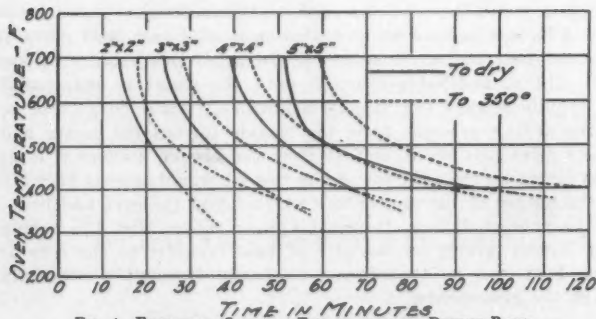
FIG. 4—EFFECTS OF SIZE AND TEMPERATURE ON DRYING RATE.

Fig. 3 is interesting in that it shows how the placing of cores on a plate can effect the drying rate. Three cores 5 in. x 5 in. x 16 in. were placed as close together as possible. Nos. 1 and 3 were on the outside. No. 2 between them. It can be seen readily how the drying of the center core was effected. When the cores were spaced one inch apart, the drying rate was approximately the same in each.

Fig. 4 is given to show how size and temperature effects drying rate. These cores were all 8 in. long. The solid line indicates the time to completely dry the core. The dotted line represents the time to bring the center of the core to 350°F.

F. E. REINERS[2]: What effect would the mechanical circulation of air have upon the conditions described in the author's theoretical determinations?

MR. WILLSEA: Any movement of air is beneficial because it wipes the molecules of water vapor off the surface of the core. The drier the air that is being moved over the surface of the core, the more water will be actually taken into the space air and held in the general oven space.

MR. REINERS: If you have a rapid circulation of air, is that not going to destroy the stratification that you are looking for? In other words, will it not remove your super-saturated air from the vicinity of the core? Is not that super-saturation going to be rapidly dispersed through the air of the oven?

MR. WILLSEA: You have to think of the oven in terms of how much moisture per minute, per cubic foot of air that actually goes out. Simply moving the air fast over the core might physically push more molecules of water vapor off the surface of the core, but the problem is still left of what is to be done with them. If you create too much disturbance, you would tend to have the entire oven space full of moisture. If the movement of air is done in relation to the air that goes out of the oven, you will tend to keep the general space free of moisture, so that as the cold, wet and heavier air falls to the floor, the hot, dry oven air will close right in on the core itself.

MR. REINERS: I can visualize a super-saturated condition or a saturated condition at 212° or 220°F., but I cannot visualize a saturated condition around 400° to 450°F. What are the relative quantities of moisture in air at the saturated condition of 220° and at 450°F.?

MR. WILLSEA: I do not think saturation at 440°F. degrees ever occurs. I mentioned that as being the ridiculous extreme in order to get the complete picture. If an oven is simply ventilated and has a source of heat, there would be a number of temperature differences within the oven space and the temperature differences would control the amount of moisture within those particular spaces because of that relative temperature difference. To get to the other part of your question, I believe there are 376 grains of moisture per cubic foot in air saturated at 220°F.

MR. REINERS: What would moisture quantities be, approximately, at 400° or 450°F? In other words, there is a very, very large spread between the possible amount of water that you can get in air at 450° and what you can get in at 220°F.

MR. WILLSEA: It is way up in the thousands, but that hotter air is not economical to handle. Perhaps I had better illustrate it in this way. When you have taken all the heat out of the air that you can from 440° down to 220°F., you have the heat for filling that air with moisture. Now, when you have done that, you have done your work for that one phase. If that air can be handled as a unit, fine, and its very heaviness unifies it, because it will fall to the floor. On the other hand, if you say, "Well, that air could hold a lot more moisture at a higher temperature; let us warm it up," you still have to cool the air down in order to get the heat for putting in more moisture. You will be coming back to where you were the first time, if you are going to handle that air and heat it efficiently; in other words, have as much moisture in the air as possible and have only the heat in the air that is necessary to hold the moisture.

MEMBER: In air conditioning your ovens that way, do you not sacrifice a lot of heat?

MR. WILLSEA: When the cold, wet air falls to the floor, it has only the heat that is needed to hold the moisture in it. If that air is pushed out of the oven, that heat in the air is that heat which enables the air to hold the moisture. It is a difference in weight. It is just as though there was water dripping off the core when the cold wet air drops from it. The heat that has given its energy to put those moisture particles or molecules into the air has disappeared from our consideration. However, the moisture is in that air and unless we leave that heat in the air, the moisture cannot stay in it. So that heat is being used as the moist air goes out of the oven and is just as useful as the air itself.

MEMBER: I take exception to the remark made by Mr. Dierker in regard to the cores drying at the bottom just as quickly at 212°F. as they do at the top. Our experience has shown that when taking a good sized flat core, it will not bake down as readily at the bottom as it will on the outside at the top.

MR. WILLSEA: I think you are both right. Mr. Dierker was referring to a local phenomenon of the heat of the plate driving the moisture away from the plate. At the surface, if there is any ventilation at all, or air movement, it would tend to wipe moisture off the surface. However, you will notice from his curves that in the center of the core it still held the moisture at 212°F.

MR. BINNS[3]: One or two important things have been somewhat overlooked. At what temperatures do the oils actually polymerize or oxidize and what effect does the moisture or a dry oven have on that?

MR. WILLSEA: One of the things that is done to the raw oil in making core oil is to polymerize it by heating it without oxygen. The big

problem in the core oil is the part which cannot be oxidized and must be volatilized. If you do not get that off early, it will tend to be trapped in the oil which is being hardened. There may be some more polymerization going on but it is only oxidation that hardens the oil. The point I tried to make is that the oil is already polymerized. The purpose of the oxygen is to oxidize it. Polymerization is a gathering together of molecules into a bigger molecule. It is merely making a bigger molecule. The oxidation is taking on oxygen by the molecules of the binder.

MEMBER: Does the presence of water in the binder prevent oxidation, even at a high oven temperature?

MR. WILLSEA: Those effects that you are speaking of are so small and those questions are in the realm of speculative chemistry. It is a question of whether it is only a physical problem of having a molecule of water in the way of a molecule of linseed oil taking on oxygen. This gentleman is referring to a practice of baking small cores in an oven at a temperature that would be enough to burn if there were not water in the core.

If you paint a house, you are using a binder similar to that in a core and it takes several days at normal room temperatures to oxidize, and polymerize that binder. In a core oven, it is only as the binder begins to get hotter and hotter that it begins to oxidize. As the core begins to oxidize, it is an oxidation similar to any other kind of oxidation even to the extent of burning, it makes heat and the core actually heats itself up. In these curves showing the heating up inside the core, it was not so much the heat coming in from the outside at that time which caused the difference, but the heat of what is called an exothermic reaction. It is really cumulative; that is, the heat from the binder increases while it is oxidizing. As soon as the temperature starts to rise, the oxidation simply goes faster. The oxidation will go ahead at any temperature. Even if you just set it down on a table at room temperature, it will start to oxidize. The reason for a core oven is first, to get the water out (which was added only for green strength), and next, to make the oxidation proceed as fast as possible.

A. M. ONDREYCO[4]: In answer to Mr. Binn's question, some years ago when we first started to use oil binders, I made some tests in a laboratory muffle furnace under very close controlled temperatures, 450° to 500°F. The core did not get hard neither did it develop strength, because there was no oxygen in the furnace to oxidize the oil and regardless of how long I left the core in the oven, it did not get hard. I think that is where a lot of foundrymen made a mistake in changing over from the cereal binders to the oil binders. The core ovens were so operated that there was not enough oxygen in the air in the oven to oxidize the oil and they produced weak cores and had a lot of trouble.

MR. WILLSEA (Written closure to discussion): We believe Mr. Dierker misspoke in calling it evaporation when referring to the fact that the water had been moved from the core near the solid metal core plate

[4] Chief Metallurgist, Meehanite Metal Corp., Pittsburgh, Pa.

at about the same time the water had actually, by evaporation, left the top of the core which was exposed to the oven air. In our paper, we did not call attention to the fact that an iron core plate picks up heat and actually puts more heat into the core than the oven air can at the same temperature. Therefore, the moisture in the core close to the plate is apt to be bubbled away from the plate toward the middle of the core. A member pointed out that this bottom effect is soon negligible in comparison to the actual removal of water by evaporation from the exposed top. The effect from the plate can go into the core only as far as the heat of the plate can be felt by the core.

On the other hand, we did mention the similar effect from an oven with very moist air which prevents evaporation. In that case, the oven air heat will bubble the moisture toward the middle, from the outside of the core. (Paragraphs 36 and 37.) The ordinary oven temperature is high in relation to liquid water and when the oven air is moist, the total of heat being conducted into the core from the air will bubble the outer core moisture toward the middle of the core. If the oven air was kept dry instead of moist, the oven air would take up that core moisture before it became hot enough to bubble.

Mr. Ondreyco brought out the important point that a core oven fired with gas or oil usually does not receive enough air even to oxidize the oil binder. On the other hand, a coke fired heater can give complete combustion with an excess supply of air so that not only is there enough extra air for oxygen but plenty of added air flow volume for carrying the core moisture out of the oven as fast as it can be evaporated, and, thereby, start oxidation sooner. Mr. Ondreyco also brought out another important point. He maintained an oven temperature between 450° and 500°F. and kept on trying to harden his cores without any effect regardless of how long in the oven. 500°F. ordinarily would burn these cores, but there was no moisture at all being carried out of the oven so that the cores remained very moist, thereby preventing their getting as hot as the harmful oven temperature. In fact, the cores could not even get hot enough to start the rapid oxidation which hardens the core oil. If oxidation can get started, it goes faster and faster because of the heat produced right there in the core by the oxidation itself. Polymerization, which is the thickening of the core oil by having the molecules of the binder join together into bigger molecules, would only take place under conditions where there is a high heat and no oxygen present, which were the actual conditions in the Ondreyco experiment. Therefore, the answer to Mr. Binn's question as to what temperature actually oxidizes the binder, is a temperature above 212°F. inside the core (because it is free of moisture) and plenty of oxygen is all that is needed to start oxidation. The oxidation will then take care of itself if plenty of oxygen continues to be available. Where there is no oxygen, there could only be more polymerization and that would be only in addition to the polymerization which had been done to the core oil during the original preparation by the core oil makers.

Mr. Reiner's questions brought out the fact that mechanical circulation of oven air is much less desirable than the natural, rapid circulation

caused by the cooled off air being so heavy that it falls down to the floor quickly, carrying all evaporated moisture with it. The hot, dry oven air will immediately close in around the core. The important point is to have a big volume of air, at about oven temperature, being forced into the oven by a pressure which forces the cooled, wet air out of the oven. This will go on continually as fast as water can be evaporated from the core so that the circulation of air is as fast as the air could be used. One would find no heat loss from this method because the heat in the air being forced out of the oven in this way is all needed to hold that much moisture in the very air being forced out of the oven.

As Mr. Reiner points out, a mechanical circulation (other than by pressure as described above) would push the cooled-off, wet air (or "supersaturated" as he calls it) into the general oven space. When this wet air is dispersed into the general air of the oven, the entire oven is correspondingly less effective for taking up more moisture from the cores.

Handling Molds and Castings
By Overhead Materials Handling Equipment
In the Small Foundry

By A. F. ANJESKEY*, WICKLIFFE, O.

The statement has been made that in order to produce a ton of finished castings, the handling operations involved are equivalent to handling from 25 to 100 tons of materials, such as sand, flasks, molds, cores, pig iron, scrap, molten iron and the rough castings through the various cleaning, grinding and inspection operations, before they are ready for shipment to a customer. It is obvious that the wide variation in the amount of materials handled, is, of course, due to the size of the finished castings.

The necessary handling and re-handling operations all represent part of the total cost of a casting, and in many foundries this represents labor cost of workmen, who could do a minimum of such laborious work if the handling between operations were performed by labor-aiding equipment. While there are large, mechanized foundries operating today, there are also many small and not-so-small foundries where the human element is necessary to perform handling operations, despite the fact that such foundries are using up-to-date molding equipment.

A chart published by *Factory Management & Maintenance*, May 1938, which was prepared by the National Industrial Conference Board, shows that of 50 leading industries the foundry industry was second highest in wages paid, representing 39.3 per cent of the value of foundry products. The reduction of this percentage is not to be construed as an effort to reduce or eliminate wages, but we believe that wasteful handling where the human element is involved should be eliminated, and it is thought that a description of how this was done in one foundry will be of interest to foundrymen.

PLANT LAYOUT

Fig. 1 shows a general plan of this foundry. The molding and pouring floor is approximately 75 x 260 ft. In this area are 52

* Sales Mgr., Cleveland Tramrail Division, Cleveland Crane & Engineering Co.
NOTE: Presented before Material Handling Session 42nd Annual Convention, Cleveland, Ohio, May 19, 1938.
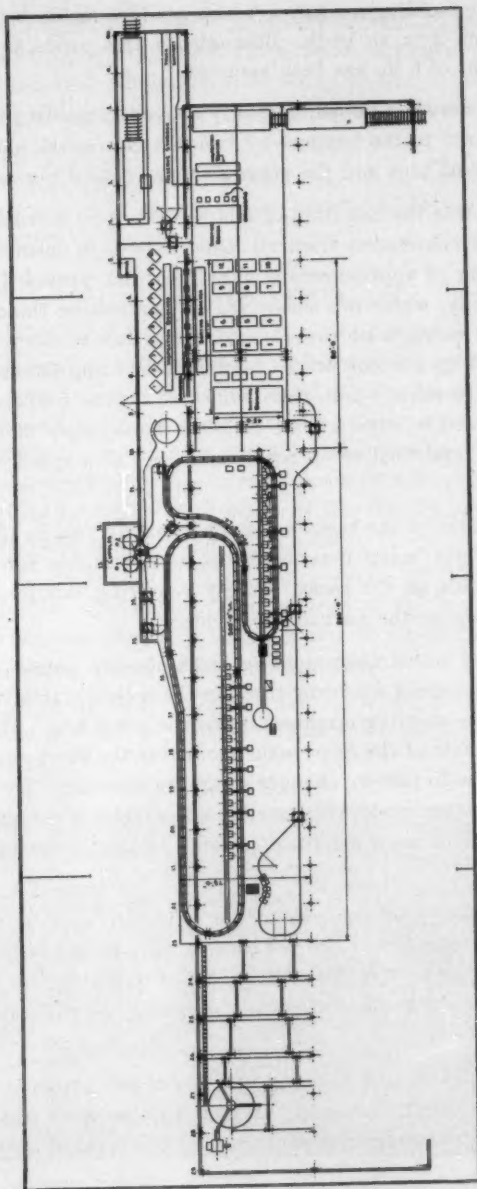
FIG. 1.—PLANT LAYOUT—SHOWING MATERIAL HANDLING SYSTEM FROM CUPOLA TO FREIGHT CAR.

squeezer type molding machines, which produce molds for castings weighing from 2 oz. to 18 lb., although for this paper an average casting weight of 6 lb. has been assumed.

Sand is stored in hoppers directly above the molding machines and is delivered to the hoppers by two belt conveyors, one having a capacity of 55 tons and the other 25 tons of sand per hour.

To facilitate the handling of molds, flasks and bottom boards, two overhead continuous tramrail loops have been installed. One loop consisting of approximately 500 ft. of track provided with 90 racks 5 ft. long, which are suspended on carriers on the overhead rail, with 30 carriers or racks coupled together to form a train. This is pulled by a motor driven tractor unit of approximately one horsepower. In other words, there are three trains of 30 racks each and three tractors with a total of three horsepower moving the racks with a total load of 50 tons on the rail at a speed of 24 ft. per min.

The bottoms of the racks are approximately two ft. above the floor, this height being determined as most suitable for placing completed molds on the racks, thereby requiring minimum bending or stooping on the part of the molder.

It will be noted that molding machines are placed on both sides of the overhead conveyor track in the molding area, with the majority of the molding machines inside the track loop. The molders on the outside of the loops make molds for the short run orders and as many as 15 pattern changes a day are necessary. By making the molds on the outside for small pattern runs it is possible to take advantage of mold handling, pouring and shake-out operation economies.

Approximately 60 per cent of the molds are made in 11 x 22 in. flasks weighing 150 lb., 20 per cent in 12 x 22 in. flasks weighing 80 lb., 5 per cent in 28 x 24 in. flasks weighing 500 lb., and 15 per cent are miscellaneous molds weighing approximately 225 lb. each.

The molders in this loop produce about 675 molds per hour, or 5400 molds in 8 hours, and the pouring operation requires 38 tons of iron, producing approximately 11,000 finished castings in an 8-hr. day.

The smaller loop consists of an overhead tramrail track approximately 300 ft. long provided with 56 racks and carriers arranged 28 to a train, each being pulled by one tractor. These racks are of the same dimensions as on the large loop and travel at a speed of 15 ft. per min.

This loop serves 20 molders, who produce 420 molds per hour, or 3360 molds in 8 hr., requiring an average of 18 tons of molten iron to produce 5000 castings.

Conveying System

Empty flasks and bottom boards are delivered to the molder on the overhead conveyor; the drag and cope molds are made by the same molder; the mold is closed, lifted off the molding machine a few inches by the molder and placed on the moving rack of the conveyor with a minimum of handling as shown in Fig. 2. This operation is repeated by the various molders around the loop, so that the racks are loaded with molds as they finally pass the molding zone on their way to the pouring zone.

Since 80 per cent of the total molds weigh 150 lb. or less, they are handled by one molder. The balance of the molds must be placed on the conveyor with the assistance of the molder adjacent to the molder who has just completed a mold.

The length of the racks has been determined to allow for placing not less than two molds on them, although some molds are multiple stacked.

Fig. 2—Placing finished molds on the conveyor with a minimum amount of handling.

As the molds reach the pouring zone, they are poured by a gang of three men with a 60 lb. ladle suspended on another tram-rail system, which is so laid out that there is no interference between any of the men in the pouring gang.

Iron is melted continuously during the molding and pouring operations and is run into an insulated mixing ladle, which repours into the 60 lb. ladle.

The pouring of molds on the smaller loop conveyor, as illustrated in Fig. 3, is performed by a gang of two men, who also obtain their iron from the same mixing ladle serving the large loop.

After the molds have been poured, they travel through a cooling zone, then reach the shake-out area. The molds are dumped on a vibrator type shake-out; flasks and bottom boards are replaced on the racks of the conveyor; castings are hooked and dumped into a roll-over type bucket; sand passes through the shake-out screen onto a conveyor belt in the floor for delivery to the sand conditioner and back onto the conveyor belt serving the storage hoppers. Fig. 4 shows the location of conveyor, shakeout and roll over bucket.

FIG. 3—POURING MOLDS OF FIG. 2 WHEN THEY REACH THE POURING ZONE.

Fig. 4—Shake-out of molds. Showing vibrator shake-out, dust collector, and roll-over bucket.

The castings at the shake-out point are accumulated in the bucket in loads of approximately 3500 lb., then are picked up by an overhead cab controlled tramrail carrier and delivered to a sorting bench in the cleaning area for sorting, inspection and cleaning operations. The tramrail carrier travels back and forth between the shake-out and the sorting bench and easily takes care of all the handling that is necessary in the day's output of castings. The dumping of castings at the sorting table is especially interesting, as shown as A of Fig. 5, due to the fact that the operator of the cab controlled carrier is traveling on an overhead rail at the shake-out point, and as he approaches the cleaning area he runs his carrier onto a 3-runway, motor driven, transfer bridge which permits him to dump the load of castings at any point on the sorting bench. This dumping is done by the operator in the cab, who releases the latch on the roll-over type bucket container without any assistance from a ground man.

The castings on this table are separated by shape; the gates are knocked off, and similar castings are dumped into the roll-over
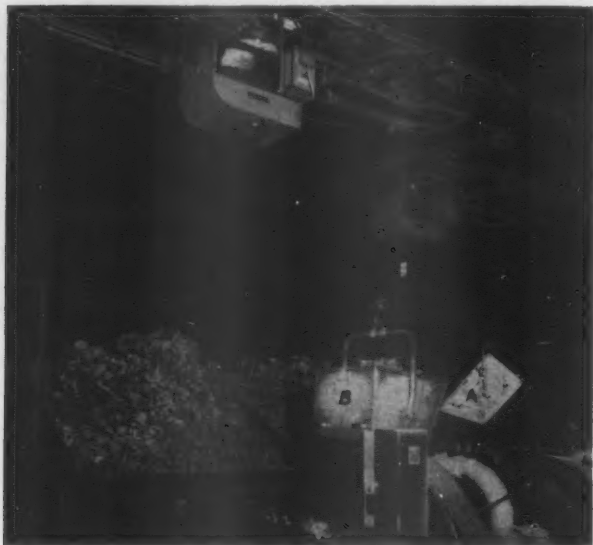
Fig. 5—Dumping castings from shake-out on sorting table (A) and inspected castings (B) being taken to tumbling barrels.

type bucket container, and when a load has been accumulated, they are picked up by a carrier operating on a second 3-runway transfer bridge, as illustrated at B of Fig. 5, and delivered to a point over the tumbling barrel and discharged into the barrel. It is necessary for a ground man to spot the container over the opening in the tumbling barrel and release the latch. The castings are discharged into the barrel, one container load being equivalent to the capacity of the mill.

Defective castings and gates are also placed in a roll-over type bucket and delivered to the scrap pile outside of the foundry and adjacent to the cupola charging floor.

After the castings have been tumbled they are dumped into a skid type container, lifted by the tramrail carrier and dumped on the sorting bench in the grinding department, shown on the plan and in Fig. 6. The castings are ground, receive final inspection and are dumped in a similar container, picked up with the tramrail carrier and delivered to a storage area to await shipping instructions, or they can be delivered to the truck shipping area and dumped into the truck.
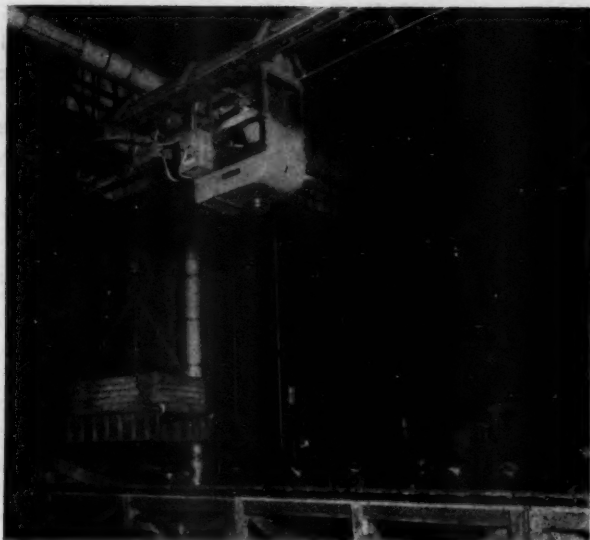
FIG. 6—TUMBLED CASTINGS BEING DUMPED ON INSPECTION TABLE PRIOR TO GRINDING.

I wish to emphasize that the tramrail system with two operators takes care of all the handling of castings and scrap from shake-out to grinder, to shipping at a cost of 18 man minutes per ton.

An electric lift truck is also used to handle the skid type containers from one section of the grinding department direct to the car, if box cars are available. Unfortunately present day box car design does not permit loading from the top, but recent developments indicate that the day is not far distant when box cars will have sliding tops, permitting loading and unloading by existing overhead equipment in most plants.

APPLICABLE TO SMALL FOUNDRIES

The same layout and handling principles involved in the making of castings in the plant just described can be duplicated in even the smallest foundry by the use of manually propelled carrier racks traveling on overhead equipment. In a small production foundry, it is possible to provide racks that will enable one man to move from 2 to 6 racks, or 5 to 10 tons, on overhead tramrail

equipment with very little effort. These racks can be moved from the molding floor to the pouring floor to the shake-out, and it is my opinion that the operation could be so scheduled that a reserve of completed molds, ready for pouring, could be moved to the shake-out zone and returned to the molding floor in time for another load, thereby eliminating costly floor space, which would be necessary if the conventional hand molding floor layout were used. When one considers that a single story foundry building for light work cost $2.50 to $3.00 per sq. ft., the savings in a smaller plant frequently more than justifies serious consideration of handling, as well as modern molding equipment.

DISCUSSION

Presiding: E. W. BEACH, Campbell, Wyant and Cannon Foundry Company, Muskegon, Mich.

MEMBER: I would like to hear some comments on mold handling for large jobbing shops, where you have non-ferrous metals, a large number of alloys and different pouring temperatures. Most of these papers have been along iron foundry practice, or where you can run things to a central shakeout from a central pouring point. This would be rather difficult to do if you had molds which required a number of different alloys and also different pouring temperatures. As an example, for 70 or 80 lb. those which would include mold and equipment.

MR. ANJESKEY: If you make a mold in one spot and are going to pour it in another spot, you can either do it as I have described in the paper or you can do it with some other form of conveyor. The molding, pouring and shakeout are all done at different points. Something is used to bring the material or take it from one point to another, and the thing to do as I see it is to workout the best scheme of materials handling. The selection of the particular conveyor will be determined entirely by the conditions in your plant. The thing I want to emphasize is — wholesale handling. Do not handle just one mold if you can avoid it, handle a quantity of them.

Stop and consider that in the foundry described we use only three horsepower of energy to move fifty ton. Three horsepower is roughly three kw, and at two cents a kw-hr., that is six cents an hour for power. How many men would you need to do this work? I would say probably another 25 men to help the 52 molders do the handling done by the conveyor.

MEMBER: I wanted to bring out that after you get the molds to a central pouring point, if you have a large group of them, you cannot tell one from the other, or what alloy is to go in, or what temperature the molds are to be poured at.

MR. ANJESKEY: How do you pour mold at present?

MEMBER: We carry the metal to the molder which is a costly proposition instead of running all the molds on a conveyor to a central pouring point.

MR. ANJESKEY: With overhead conveyor equipment you do not have to be confined to one pouring zone. You are able to buy equipment today, either hand or power propelled, and by the use of switches divert the particular molds that require a certain alloy to a particular pouring zone and pour them. You perhaps have a central shakeout zone, and as long as you are keeping the molds moving without rehandling you are doing work with the least effort. You may have to use a few switches, but a layout of overhead trackage will be a lot simpler for your operation.

W. A. DIEHL[1]: There are three carriers on one unit. If a man moves those individual carriers and there are some oscillations you will have shifts, etc. How do you handle that difficulty?

MR. ANJESKEY: On the layout shown the racks were coupled at the top and bottom of same, with the coupling so designed that there would be no binding. Considering the speed of the conveyor at 24 ft. per min. there was little tendency for the racks to swing.

MR. DIEHL: Are the three units indexed so that a man cannot start a unit and bump into the next one?

MR. ANJESKEY: Each tractor pulls a train of racks, but if one tractor is pulling a greater load than the other it slows down a bit and the other tractor behind it, if it is traveling faster, "nudges" the last rack on the slower train, causing the particular rack to speed up. This speed is also transferred to the racks ahead, thereby taking some of the load off the slower tractor and causing it to speed up. Since three separate trains are operating on the loop, the load on the tractor motors is equalized and you have the advantage of applying power at three different points on the loop instead of one point, thereby relieving stresses on the overhead track.

MR. DIEHL: Is there a lot of core work being carried on them?

MR. ANJESKEY: There is some core work, but this has not been a problem.

MR. DIEHL: What do you do about lubrication, particularly of the carrier wheels?

MR. ANJESKEY: Carrier bearings are anti-friction ball bearing type which are prepacked with a lubricant that will last as long as the wheel. Five years ago this was considered impossible but we have found that the type of bearing used is very satisfactory when provided with suitable seals for retaining the lubricant. This reduces maintenance. I might also mention that the improvement in bearing design also enabled us to

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[1] Master Mechanic, National Malleable & Steel Casting Co., Cleveland, O.

use the same type of bearing without a lubricant on core oven applications for temperatures up to 450° F.

MEMBER: Why is not an overhead monorail conveyor used to distribute the metal to the pouring zone? It seemed to be the one point that was out of step with the layout of the ladles being handled by man power. Why was that not handled by a motor driven carrier?

MR. ANJESKEY: It can be done if the purchaser is willing to pay the price. We have worked on several projects where it was desired to deliver metal a distance of 150 ft. by dispatching the ladle from the cupola to the pouring zone. This can be done by providing an electrically propelled carrier traveling at a speed of 50, 75 or 100 ft. per min., which can be dispatched from the cupola with proper control equipment, and such unit will travel to the pouring zone without the assistance of a man, as it reaches the pouring zone the unit stops. At this point the operator in the pouring gang can energize the unit and begin his pouring, having full control of the travel speed of the carrier, and when he has emptied the ladle the carrier with the empty ladle is shunted to a return track where it can be dispatched automatically to a point near the cupola where it stops, and the control of it is again in the hands of the men at the cupola who can refill the ladle and dispatch the carrier automatically to the pouring zone. Such carrier can be provided with an electric hoist, permitting the pouring man to raise or lower the ladle as necessary to properly pour molds. Automatic dispatching equipment of this type is practical, and such carriers have been used to dispatch loads between operations in the automotive industry. It must be remembered, however, that in dispatching ladles of molten metal automatically speeds cannot be as high as they would be for handling other types of loads.

MEMBER: I realize it is being done. My point was — why was it not incorporated in the layout illustrated here.

MR. ANJESKEY: With 60 lb. of iron per ladle and with a pouring zone that is not over 35 ft. long and three men doing the pouring, with so little travel one can hardly justify delivery of metal by a power unit. It could be done and probably would justify itself if the distance between the cupola and the pouring zone was greater. In some plants metal is delivered in two ton bull ladles by an electrically operated tramrail system and repoured into smaller pouring ladles at various pouring points in the foundry. This is practical when loads are greater and distances traveled are too fatiguing for an operator.

Hints on Molding and Preparation of Castings for Enameling.

By R. B. Schaal*, Cleveland, O.

Abstract

Enamel is one of the most beautiful, useful and practical finishes for cast iron parts. Its success is influenced by design, sand, molding and casting preparation. Accuracy of dimension with proper clearances and tolerances is a fundamental requirement. Aluminum matchplates are advisable for accuracy and interchangeability. Large flat surfaces are difficult to enamel and should be avoided. All factors conducive to warpage and buckling in enameling furnace should be considered and worked out, to prevent hairlines, poor fit, and blistering. Foundry sand should be suitable, and carefully controlled. Graphite facing should be avoided. Castings should be removed from sand before cooling, stored in a warm place, carefully ground or buffed to remove surface defects, and blasted with care before enameling.

1. When properly applied, enamel affords a beautiful and exceedingly durable finish for cast iron parts. Its widespread use on the cast iron parts of ranges, and on all types of sanitary ware is eloquent testimony to this fact.

2. Basically, enamel is a form of glass, possessing certain special properties such as fit to the iron, great covering power, and infinite color possibilities. It is resistant to the ordinary corrosive and abrasive conditions encountered in service. It is heat resistant, alkali resistant, and in some forms acid resistant to a remarkable degree.

3. With proper precautions in design of parts, and their preparation, it is neither difficult nor expensive to apply and, in cases where the appearance of the finished article is of vital importance, it is probably without equal as a finish.

4. The success or failure of enamel as a finish are vitally influenced by the design of the parts, the accuracy of their fit when assembled, by factors in the foundry such as sand condition, types of pattern, and treatment after molding in preparation for the enameling itself.

DESIGNING THE CASTINGS AND PATTERNS

5. In the manufacture of cast iron parts for enameling, the foundryman and molder are helpless without appreciation of certain enameling requirements on the part of the designer of the articles to be enameled.

6. Where iron parts are to be assembled, rather low standards of accuracy sometimes will suffice if enameling is not in question, and the parts are to be merely japanned, plated, parkerized, or finished in some such manner. But in the case of parts to be enameled, exact fit sometimes will be the success or failure of the entire project. Utmost accuracy of dimension is required.

7. Enamel has a very appreciable thickness, and clearances and tolerances need to be worked out carefully to insure accurate fit, especially where two enameled surfaces or edges are to be fitted together. In fact wherever possible, direct enamel to enamel contact should be avoided, or these joints shimmed with soft metal strip. Stresses introduced by assembling imperfectly fitting enameled parts may cause immediate damage in mounting or parts may fail prematurely in service. Enamel, while it will take a surprising amount of abuse or rough treatment, still is by nature similar to glass, and needs to be handled and considered with this fact in mind.

Importance of Accuracy

8. Accurately made enameled parts will fit with a minimum of grinding, and hammering or forcing into position. Where an enameled edge has to be ground, there is always the hazard of damage in the grinding operation itself, and increased liability to future premature failure in service due to weakness at such points.

9. Accuracy in manufacture is promoted by the use of accurately made aluminum match plates, with the gates and runners cast as an integral part of the pattern. This practice usually affords the best results from the production point of view, and is a valuable aid in obtaining accuracy and interchangeability of parts.

10. In addition to uniformity and accuracy of dimension, carefully worked out, several other desirable precautions may be considered with profit.

Flat Surfaces

11. Probably there will be agreement that one of the very hardest types of light casting to mold successfully is a large perfectly flat surface. Enamel defects and blemishes become very prominent on such surfaces and, therefore, they are most difficult to enamel successfully. In view of the decided design trend toward flat, or modernistic lines in range design, it is desirable that these surfaces be relieved or broken up wherever possible. The clever designer can do this without detriment to the general effect, and such forethought will be repaid many times in production gains, freedom from defects and reduced costs.

Warping and Buckling

12. In this connection, the designer familiar with enameling requirements will realize that enameling itself is to the castings a form of heat treatment, and he will use particular care to see that the castings are so turned out that they will not warp or buckle in the enameling furnace. In this, he will receive much aid by close cooperation and consultation with his foundry superintendent and pattern maker. In addition to the mounting and assembling difficulties introduced by warping or buckling, these two phenomena are the cause of much direct rejection of castings in the enamel shop.

13. Enamel is sprayed on the castings somewhat in the form of thin clay slip and is dried before burning. When the casting buckles from the sudden heat of introduction to the enameling furnace, the dry enamel film may crack open. Usually, such cracks partially fuse together but are visible as depressed lines after firing. These depressed lines are called hair lines by the enameler. If all the buckling tendency is removed in the first fire, a second coat may cover the lines, but if the buckling is repeated (it usually is) the fired first coat, as well as the dry film of the second coat, may crack and deep lines, possibly blistered, result, with consequent rejection of the enameled part.

13. When areas of the casting are grained in a contrasting color, even faint hairlines will show through the grain. Slow heating, such as that normal to the operation of the continuous

enameling furnace, usually helps to keep down this hairlining tendency.

14. Warping and buckling, or their absence, are dependent largely upon the casting section and shape. It seems to be well demonstrated that castings which heat uniformly buckle the least. Here again, the designer has a problem, the solution of which calls for both considerable knowledge and much thought. Even heating above all calls for uniform casting section. A heavier section always will arrive at a given temperature more slowly than a light one adjacent to it, consequently much effort is directed to producing a pattern in which all sections are equally thick. Edges seem to heat faster than flat surfaces, so they may be designed to compensate for this tendency. Heavy lugs and ribs are avoided when possible.

Location of Parting Line

15. The location of a parting line can influence enameling results materially. For example, in Fig. 1, castings A and B are flanged. The difference between them seems to be only a matter of a parting line, yet casting A will probably hairline where B will not.
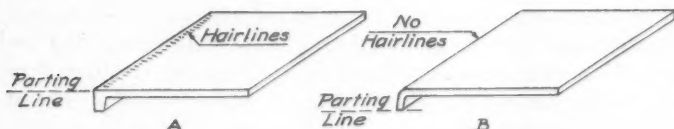
FIG. 1—HAIRLINES ARE INFLUENCED BY TYPE OF FLANGE.

16. Another important enameling consideration is illustrated by an examination of Fig. 1, if we consider that the outsides of the flanges are to be enameled. Where the parting line occurs in casting A, or where the top of the flange meets the face of the casting, there is a sharp edge. When fused, enamel tends toward a spherical or cylindrical shape, consequently this edge tends to come through the enamel coating as shown in Fig. 2. Either a dark line, or bare iron probably results. Casting A of Fig. 2 shows the behavior of enamel over the corresponding point of casting A, Fig. 1.

Uniform Casting Section

17. A further important consideration from the enameler's and the foundryman's point of view may be illustrated by Fig.

3. It also will help again to illustrate the advantage of uniform casting section. If we consider for a moment what happens to the metal when it solidifies in the mold at point X of casting E, Fig. 3, we find that this point, representing a heavier section than the corresponding point Y of casting F, will be slower to solidify than the remainder of the section. This means that the iron in the point of surface above X will be at least looser grained than in the surface over the remainder of the section. Loose grain or "soft" areas in the casting surface are more reactive with enamel than the adjacent closer grained surface, and tend to boil and blister more, given the same enameling treatment.
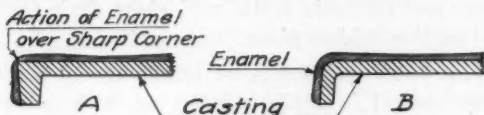
FIG. 2—SHARP EDGES ALLOW IRON TO COME THROUGH ENAMEL.
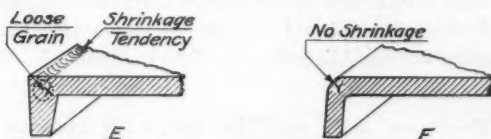
FIG. 3—BLISTERING IS PROMOTED OVER LOOSE GRAINED AREAS AND SHRINKAGE IS A PROBABILITY OVER HEAVIER SECTIONS.

18. In addition to the increased boiling and blistering hazard at this point, we may even find shrinkage. If this heavier section is fed (as is usual) through the lighter section adjacent, the feeding may be cut off before the flange solidifies completely. This may result in a shrinkage line, the depth of a "rat tail," let us say, along the entire top of the casting over the flange, which will certainly cause rejection after enameling, even if no boiling or blistering has occurred.

Desirable Features in Design

19. The desirable features in design of enameling castings then may be stated as follows:

(1) Careful attention to clearance and tolerances.

(2) Accuracy of the pattern to insure proper fit upon assembling.

(3) Reproduction of pattern accuracy in each casting.

(4) Freedom from unusual warping and buckling on heating.

(5)   Uniform section.

(6)   Care in selection of parting lines.

(7)   Breaking up of large flat surfaces.

(8)   Elimination of sharp edges.

SAND

20.   Successful enameling is largely dependent upon soundness and regularity of the casting surface. For example, "rat tails," which may not be prominent enough to show plainly on the raw casting, will be thrown into high relief when enamel is applied. Extremely rough areas on a surface, caused by sand falling from vertical mold walls, will show dark spots through the enamel on the finished piece.

21.   Sand holes and blow holes usually are filled with ceramic filler before enamel is applied, but even with great care this filling sometimes blows out, shrinks a little or shows a depressed or raised area which may cause rejection of the enameled part. A honeycombed surface, due to excessive steam generated in the mold, usually causes blistering or black specks in the enamel coating.

22.   Evidently sand condition plays an important role in enameling.

Rat Tails

23.   Apparently, there are two usual explanations for "rat tails," the first being buckling of the mold surface from contact with the hot iron, the second the formation of a water globule, similar to a drop of water on a hot iron plate, which is pushed along by the molten metal to an extremity of the mold, forming the "rat tail" in the metal with which it is in contact.

24.   Whichever of these explanations we choose to adopt, the conditions causing both buckling of the mold, and the formation of water globules may be prevented with profit. Lack of permeability, due to fineness of the sand, can cause the mold surface to buckle, while excessive water can cause the formation of water globules.

25.   Washing and sand falling from vertical mold surfaces usually are caused by insufficient green strength in the sand. Blow holes are caused by lack of permeability due either to fineness or to high water content of the sand. Mold gases cannot

escape properly and blow back through the sprue. Judicious venting of molds helps out well in such cases.

26. Reviewing these conditions which cause enameling difficulties, we conclude that the sand should have good green strength, combined with reasonably high permeability and that the sand should not be too fine nor the water content too high.

Selecting Sand

27. To set arbitrary limits to these specifications would be quite impracticable, and perhaps impossible. We can, however, learn much as to our own individual requirements by reviewing briefly the principles involved in selecting a suitable type of sand.

28. *Grain Size*: Probably the first consideration in selecting a sand for any given type of work is suitability of grain size. In general, this property is determined largely by the finish required on the castings. For enameling surfaces, extreme smoothness is not desirable. Although the metal itself should be as dense and close grained as possible to resist attack by the enamel, a certain mechanical "texture" is a helpful means of promoting adherence. Manufacturers of sheet metal for enameling purposes achieve this type of surface by cold rolling to produce *metal density,* and then running the sheets through etched rolls to *roughen* the *surface.* An inspection of a casting surface prepared for enameling by heavy blasting with steel grit will indicate at once the surface "grain" desirable. The coarsest possible sand which will produce this texture uniformly in the mold usually will be found very satisfactory.

29. *Uniformity:* Those who handled synthetic sands early, learned to distinguish between "average" and "uniform" grain size. From the theoretical point of view, sand of a given grain size would have each grain alike, a condition which is practically impossible. However, synthetic sands may be prepared by washing, drying and screening which will produce considerable uniformity, all coarse material, and all fines being rejected.

30. Natural sands may be riddled to remove coarse grains, but the disposition of the "fines" is more difficult. The sand cannot be washed and dried, otherwise all bonding material would be lost. However, sand may be purchased which has only a proper proportion of fine grains, and such sands should be specified carefully. A general principle to follow, after the proper grain size

is once determined, is to obtain a sand whose grains are as uniformly the desired size as possible.

31. *Refractories:* If proper grain size and uniformity of grain size are important properties of molding sand, probably refractoriness is equally significant. Synthetic sands are prepared from material which yields grain as free as possible from iron and very difficult to fuse. All bonding material and fines are removed carefully by washing, ridding the sand of all salts and fusible ingredients. The bond then is restored by the addition of plastic material of the type of fire clay or bentonite.

32. Natural sands inevitably contain fusible material. It only can be emphasized that material such as alkalies or alkaline earths should be present in as small amount as possible, consistent with fair bonding properties. Refractory sand resists the washing action and heat of the metal in a way to permit the use of a minimum of graphite painting opposite the gates, and consequently makes for good enameling properties.

33. *Permeability:* Given a refractory sand of proper and uniform grain size, its content of water and bonding materials should be so balanced as to permit gases to escape from the mold with the greatest possible ease. This sand quality is called permeability and should be kept at the maximum commensurate with sufficient "green strength."

34. *Strength:* Green strength is the measure of cohesion, or of the ability of the sand to resist deformation or disintegration after molding. It increases with the amount of bonding material in the sand and, up to a certain point, with water content. Both bonding material and water reduce permeability, and it is therefore necessary to effect a suitable compromise in these two properties. Since grain size is dealt with and established, it remains to develop suitable criteria for the two variables, bonding material, and water content.

35. The bonding material in molding sand will vary in bonding ability, depending upon its source. If we were to construct a synthetic sand using fire clay as bonding material, we would need a certain minimum quantity to produce good green strength with a reasonable amount of water. If the bonding material, instead of fire clay, were bentonite, a much smaller quantity would be needed for the same water content.

36.   In general, it may be stated that the stronger the bonding properties or plasticity of the bonding material, the less of this material needs to be present for satisfactory green strength. Given two sands of equal grain size, which, with equal water content (say 7 per cent) will produce equal green strength, the one which carries the least bonding material will have the greatest permeability, and usually will be the better molding sand.

37.   From the other point of view, given two sands of equal grain size and equal amounts of bonding material, the one which produces satisfactory green strength with the least water will have the greatest permeability and likely will be the more satisfactory of the two.

38.   The clay content of a good natural sand, then, should have high plasticity, or good bonding quality, and be not too great in amount. Good green strength should be obtained with 5 to 7 per cent water. Permeability should be as high as possible.

39.   *Permeability and Moisture:* Lack of permeability in a mold hinders the escape of the relatively large volumes of steam and other mold gases. These gases sometimes penetrate back through the first thin skin of the iron to solidify, forming fine sub-surface holes. Such holes may or may not be opened by blasting, and are a prolific source of blisters in enameling. Formation of large quantities of steam, or hindrance to its escape, sometimes causes fine holes and chilled surfaces on thin portions of the casting. Chilled surfaces usually blister and boil on enameling. From the enameling point of view, then, sand of low water content and high permeability offers the best chance of good finished ware.

40.   *Deterioration:* Molding sand deteriorates with use. The terrific local heating at mold surfaces causes the grains to split and shatter, and destroys the bonding property of the adjacent clay. The tendency is for the proportion of fines to increase, and green strength to decrease (with constant water content). A certain amount of ferrous material also drops from the castings during shaking out.

41.   Unless the deterioration is compensated at frequent intervals, the sand eventually becomes unworkable. The accumulation of fines, by choking the spaces between sand grains, lessens permeability. To make up for the destruction (but not removal) of bonding material, more water must be added to produce suffi-

cient green strength. The extra water needed also chokes the spaces between grains and lessens permeability.

42. Permeability and grain size are kept within limits, usually by the daily addition of new sand to each heap. The fresh sand is thoroughly mixed or "cut in" with the old before wetting down. Eventually, even the addition of moderate amounts of new sand no longer will restore bond and grain size to an old sand heap. The entire heap then either is discarded, or part of it is mixed with a large proportion of new sand.

43. *Use of Coal or Coke Breeze:* Molding sand sometimes is mixed with sea coal or, preferably, with coke breeze, with a view to increasing its refractoriness and workability, but this practice has been viewed with suspicion by the enameling industry, and cannot be recommended. Black sand, as it is termed, is in use in very few foundries making light castings in the United States. It is only fair to state that castings which have enameled with satisfaction have been made in such sand heaps, and a thorough test might develop the possibility of using some such sand without harm resulting in the enamel shop. Certainly, such a practice is far less harmful than the use of graphite facing.

FACING MATERIALS

44. Much debate has centered around the use of facing materials in connection with castings to be enameled. Probably the only type of facing against which any real case has been established is graphite. It has been shown definitely that graphite facing is detrimental to enameling qualities, chiefly, perhaps, because it cannot be removed entirely from the casting surface by blasting. Any graphite remaining on the casting surface will react with enamel and cause blisters, or will destroy the adherence of the enamel.

45. Probably any refractory facing material which does not close the pores of the mold might be used with success if it does not react with enamel or can be removed completely by blasting. Facing material should not give off much gas when in contact with hot iron.

46. The prejudice against facing materials in general, in enameling circles, possibly is not warranted entirely, and probably has been brought about by the proved detrimental effect of graphite. The need for facings is lessened materially by the use of very fluid iron poured at only moderate temperatures.

47. To restate then, molding sand for making enameling castings should have uniform, not too fine grain size. It should be fairly refractory. The bonding materials should be highly plastic, and comparatively small in amount. Satisfactory green strength should be obtained with a minimum percentage of water. Finally, the sand should be brought daily to a definite, uniform condition. Graphite facing preferably is not used in connection with castings to be enameled. Venting of molds should not be neglected.

MOLDING CONSIDERATIONS

48. Given the proper kind of pattern equipment, as discussed under design, and a well selected and conditioned sand, the molding operation itself is reduced to its simplest terms. From the enameling point of view, little is desired in making a mold beyond the requirements for the manufacture of any good quality casting.

49. Special effort may be necessary to keep sand from sticking to patterns, thus roughening the casting surface. A little lycopodium powder, judiciously dusted at critical points, is an aid in remedying this condition.

50. Strengthening exposed edges by wetting should be avoided, and should be unnecessary, if suitable casting design is provided and sand conditions are correct. Organic compounds such as dextrin are avoided.

51. The necessity of protecting gates and molds from hot metal wash by graphite painting, often is the result of a combination of gating and iron unsuitable for the job. For enameling purposes, iron should be particularly fluid, so it may be poured at the lowest temperatures possible. This is a metallurgical problem which will not be discussed here. Gates preferably should be distributed and sized so all parts of the casting may be filled without a large proportion of the iron having to pass over any one or two spots in the mold, and so that all parts shall be filled as nearly simultaneously as possible. This is a point to be settled between foundry superintendent and pattern maker. It is remembered that a uniform, close dense surface makes for successful enameling, and gating systems that cause swirling and washing will probably produce alternately dense and porous areas on the surface of the casting.

52. Vertical gates are not used on flat surfaces. An endeavor is made to gate no vital spots where enamel must be absolutely

perfect, due to the usual looseness of grain and tendency for holes to occur at gated points. Consequently, gating a flat surface vertically is to invite trouble at a most vital point.

53. A combination of low temperature, very fluid iron, and a judicious gating system are vital factors in the production of an enameling surface.

SHAKING OUT, STORING AND CLEANING

54. Castings to be enameled preferably are shaken out before cooling much below red heat, and are removed from contact with wet sand. Absorption of much moisture by the casting usually is detrimental and sometimes fatal to enameling results. Therefore, castings are preferably enameled as soon as possible after molding, and preferably always are stored while in process in a warm dry atmosphere.

55. Formerly, sand was removed from castings in the tumbling barrel. This practice probably was preferable to the modern grit blasting or sandblasting prior to trimming. The peining action of the tumbling barrel helped seal the pores of the casting against moisture. In consequence, dry warm storage during processing has become more than ever desirable.

56. After the removal of the sand, burrs are ground from edges and parting lines by hard wheels. Rat tails and surface defects must be removed by soft buffing wheels, since the mark left by the hard wheel usually is as visible through the enamel as the original defect itself. Little knobs formed on the surface where the pattern has picked up sand from the mold surface also are removed by the soft wheel.

57. Just prior to enameling, the surface of the casting is cleaned and prepared by blasting, either with compressed air and sand, air and steel grit, air and a mixture of sand and grit, or by mechanically slinging steel grit against the casting. Grit without air is finding increasing application, and in many installations is working with fair satisfaction.

58. If compressed air is used, every effort should be made to insure its freedom from an abnormal amount of moisture. Sand, or sand and grit mixtures keep drier than steel grit by itself, and also keep cleaner.

59. After blasting the castings are ready to be enameled.

DISCUSSION

Presiding: DR. JAS. T. MACKENZIE, American Cast Iron Pipe Co., Birmingham, Ala.

MEMBER: The author stated that it has been shown definitely that graphite is detrimental because it cannot be removed from the casting surface. Do I understand that if the graphite is on the surface, there is not any penetration? If graphite cannot be removed, is it visible after the blasting operation?

MR. SCHAAL: In most cases it will be visible. Whether there is much penetration into the casting by the graphite is open to argument. Whether it could be called a penetration of the graphite, or whether the iron will take some of the graphite and incorporate it into the iron structure itself is open for discussion. Nevertheless, heavy graphite painting is impossible to remove properly by any amount of sand-blasting within reason. Another bad feature is the cumulative effect. When using steel grit, graphite will get all over the surface of the grit and stop the grit from cleaning properly.

An illustration of the effect of graphite paint is found in England and France where all the enameling iron is made in black sand and the molds are all heavily coated with graphite before pouring. When these castings come out of the mold, there is no sand adhering to them; the surface is absolutely bright; there is no necessity for doing any cleaning before putting them in storage; they are just as bright as can be.

Those castings are absolutely impossible to enamel without annealing. It is impossible to blast them. The iron surface is as good as that of any castings made in this country. The composition of the iron is good and after the castings are annealed, they enamel very well. But the blacking on the surface has to be burned out; it can not be blasted out.

There are other kinds of facings that are not in any way detrimental to enameling. Graphite, we think, is a hazard in this work and we can not recommend it.

MEMBER: Is it not difficult to make stove plate without enameling and stove plate for enameling with the same analysis?

MR. SCHAAL: I do not see any particular difficulty. The thing we are after is soundness of surface and that would be equally desirable in the case of castings which are not to be enameled as well as castings which are to be enameled.

MEMBER: I could not reconcile myself to a finished stove plate casting made in sand with the permeability and strength figures that were quoted. I can not see where it would give good enameling results.

MR. SCHAAL: There are no figures given. We know that we have to have sufficient green strength in the mold but, too much water in the mold is also bad. The idea, then, is to have as much green strength with as little water as possible. That is the important point. If we have

sufficient green strength, with little water, we may have to have recourse to a lot of clay in the sand, unless that clay is of an extremely plastic nature. The ideal combination of the sand is a clay content which is just moderate in amount but which has very high plasticity. Therefore, only a moderate amount of water will give a great deal of green strength, and maximum permeability. I do not mean to suggest that we should have the grain size too large.

E. H. KING:[1] Referring to stove plate practice as found in this country, the matter of moisture content and permeability has been entirely over-stressed. The fineness of the sand and the clay content will necessarily dictate the amount of water which the sand will properly absorb. Many foundries will hit toward a high permeability in their sand and deliberately work the sand with what we are prone to call "free" water in the sand. There is actually more temper water added to the sand than the clay is able to absorb. The grain of a molding sand will absorb no water. If free water is on the surface of a mold for a very small casting or a light section casting, that moisture will go into the metal and will give a chilling effect. Excess temper water will invariably lead to enameling trouble.

The stove plate industry does try to have more bond in its heaps than perhaps any other branch of foundry practice, the reason being that stove plate foundries encounter trouble with the expansion and contraction of their sand due to the very light metal section. Pyrometers have been buried in stove plate molds within ⅛ in. of the surface, and have yet to record a temperature at the mold metal interface of higher than around 1800°F. When hot metal is thrown into sand, the sand will naturally expand. The only way the stove plate man or the man making light sections has to counteract that expansion is through the contraction of his clay, which will contract as its temper water is driven off through a natural process. If he will hold his bond high, higher than that which would be considered good foundry practice for slightly heavier work, he can balance out his sand and get castings that are free from surface defects. I have seen some of the finest stove plate castings which are later enameled and enameled perfectly, made in moisture contents of 8½ per cent, 9 per cent, 9½ per cent and even as high as 10 per cent, providing, of course, the sand can absorb that moisture and it does not have a distinctly wet feel. Trouble in cases which require higher permeability can be done away with as long as an excess of water is not present in the sand.

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[1] Hougland and Hardy, Inc., Evansville, Ind.

The Relation of Microstructure to the Enamelability of Cast Iron

By G. H. Spencer-Strong*, Baltimore, Md.

Abstract

The relation of the micro-structure of enameling iron to the enamelability, as concerned with the four major enameling defects caused by the iron, that is, blistering, hairlining, poor adherence, and warping, are discussed. Blistering is the most prevalent of the defects. The effect of iron oxide, chill structures, inclusions and blow holes and of the effects of graphite, free carbon and the decarburization of pearlite is discussed. It is indicated that a considerable amount of investigation will be required before it will be safe to generalize on the relation of microstructure to enamelability.

Introduction

1. The problem of the enamelability of cast iron has for many years been of great interest and importance to enamelers and foundrymen alike. Thirty years ago, most stoves were constructed largely of cast iron. Today, gas and electric stoves contain little cast iron and manufacturers of wood and coal ranges are searching for means to eliminate the iron from their products as well.

2. Continual tightening of inspection standards coupled with the difficulties encountered in producing first-class ware in cast iron and the expensive sieges of defects so often encountered, all have helped to reduce the use of the material. However, enameled cast iron does have its field of usefulness so that its use might well be increasing rather than decreasing were it not for the growing disgust on the part of many enamelers at the difficulties encountered in its use.

ENAMEL DEFECTS

3. When one considers the fact that the enameler demands almost chemically pure iron for his sheet iron requirements, the difficulties encountered in cast iron enameling do not seem strange. Four major defects encountered in the enameling of cast iron may be caused either in part or entirely by the iron. These are:

(1) Blistering—the disruption of the enamel by the formation of gas bubbles at the enamel iron interface and the passage of these bubbles through the enamel to the surface.

(2) Failure of the enamel to adhere properly to the iron.

(3) Warping of the casting either before or during the enameling process.

(4) Hairlining—the formation of fine cracks in the enamel due to strains set up either during the enameling operation or in service.

INVESTIGATION OF ENAMELABILITY

4. The study of the enamelability of cast iron and of the causes for failure of cast iron to enamel, have been carried on chiefly at the instigation of the enamelers rather than by foundrymen and metallurgists. Thus, in 1924. the research committee of the Enamel Division of The American Ceramic Society, in answer to appeals from cast iron enamelers, undertook an extensive study of the causes of blisters and the results of the work having been recorded in a paper entitled "Blistering in the Enameling of Cast Iron," by Krynitsky and Harrison, published* in 1930. The work was carried on at the NATIONAL BUREAU OF STANDARDS and the authors reported that the majority of the defects were the results of a phenomena which they called "micro-chill," in contrast to the well-known macro-chill. The following year, the research committee decided to continue its work in a private laboratory. The results of the investigation were published in 1933 in two reports by E. P. Poste, entitled "The Blistering of Cast Iron Enamel." More recently, the technical research section of the Porcelain Enamel Institute has published a bulletin on the "Production of Castings for Enameling," with the hope that the work may give the enameler and foundrymen alike a better conception of some of their mutual problems.

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* A bibliography is appended to this paper, giving reference to the publications mentioned in the body of the text.

5. During the past few years, the laboratories of the several enamel frit manufacturers have taken an active part in the study of the enamelability of cast iron in an effort to aid their customers. This was somewhat impeded by the lack of data concerning the metallurgical history of most of the specimens received. This lack of information makes generalization concerning the results of a great deal of the work difficult. However, such studies do have one advantage over strictly laboratory investigations in that they usually represent the worst problem of the enameler.

6. Considerable work on the enamelability of cast iron also has been carried out in Europe and more especially in England. Of these, the most complete summary was presented in the article entitled "Cast Iron for Vitreous Enameling" presented at the Scottish Conference of the Institute of Vitreous Enamelers in 1936 by Dr. H. P. Angus.

CONSIDERATION OF THE STUDY OF MICROSTRUCTURE

7. Many persons may prefer to interpret the meaning of the term "microstructure" as the consideration of only the actual basic structural components of cast iron; that is, the graphite, temper-carbon, pearlite, cementite, ferrite, and steadite. In the consideration of enameling iron metallographically, often it is found that foreign materials and structures in the iron are as important as the basic components. Thus, the present paper will include as part of the microstructure not only the ferrite, pearlite, steadite, and graphite, but also such components as slag inclusions, iron oxide inclusions, and blowholes.

8. In a great many of the past considerations of the enamelability of cast iron, interest has been centered chiefly upon the composition of the iron. A great deal of attention has been given to what goes into the cupola but not so much attention has been paid to what comes out, from the metallographic standpoint. Thus, the temperature of the iron, the size and design of the casting, the molding practice, length of time in the sand, and finally, since the enameling operation is an annealing or normalizing treatment, the temperature and time of enameling fire all will affect the microstructure of the casting and the enamelability. Therefore, the study of the relation of microstructure to the enamelability of cast iron should prove of considerable value in solving the joint problems of the cast iron foundrymen and the enamelers. Such studies

also may prove of interest to metallurgists working on the mechanism of the graphitization of cast iron.

EFFECT OF MICROSTRUCTURE UPON BLISTERING

9.   Blistering is by far the most serious defect encountered in the enameling of cast iron. The defect arises from the generation of gas in or at the surface of the iron after the enamel has begun to fuse over and results in the destruction of the continuity of the surface of the enamel. Sizes of blisters and their advancement will vary rather widely. Thus, one may encounter blisters ranging from the size of a pinhead — so-called "pinholes" — through eruptions the size of a pea to great craters about as large as a dime.

10.   If the blister has broken and partially healed, pits in the enamel are evident, so that pitting also may be considered as a type of blistering. The gases giving rise to blistering defects have been identified by Krynitsky and Harrison as the oxides of carbon; i. e., carbon monoxide and carbon dioxide, indicating their source as either the graphite and temper-carbon in the iron or the combined carbon therein.

(A) Iron Oxide

11.   Obviously, a source of oxygen is necessary for reaction with the carbon. Krynitsky and Harrison state that the oxygen is derived almost entirely from the air. Poste offers the oxide of iron either at the surface, or, in some cases, within the metal as an added source. Angus feels that blistering is due very largely to the ability of an iron surface to absorb oxygen and iron oxide, and that this absorption may take place during the cooling of a mold, during annealing before shot blasting, or during the early stages of enameling. He states the belief that the ability of irons to absorb oxygen and iron oxide differs and causes the variation in blistering tendencies noticed. Gaseous evolution, due to the presence of iron oxide, has been described by Moldenke. Wm. Bogenschutz, a foundryman of great experience in the production of enameling iron, also has maintained that iron oxide is one of the chief sources of blistering defects.

12.   Evidence at hand would indicate that iron oxide may or may not cause blistering. Fig. 1 shows a section through a blistering casting in which the various areas on the surface were spongy
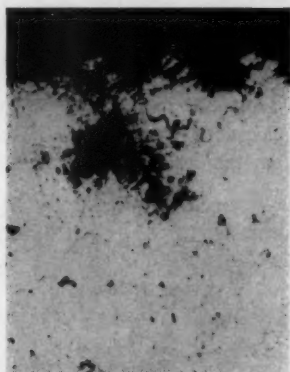
Fig. 1—Section showing iron oxide concentration in blistering cast iron—Unetched— 100x.

with rather large iron oxide formations. In each case, blistering occurred over these areas.

13. Fig. 2 shows a somewhat similar formation of oxide from which no blistering resulted. This photomicrograph also shows a rather heavy formation of iron oxide between the enamel and the iron. Poste shows photomicrographs which are cited as examples
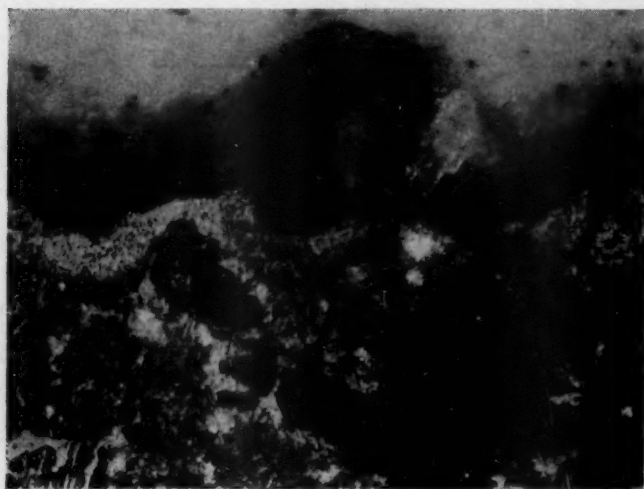
Fig. 2—Section showing iron oxide formation in blowhole (top center) and between iron and enamel (left center) in non-blistering iron.—Etched 10% Nital— 150x.

of blistering due to iron oxide. In some of these cases, the iron
appears to have been burned; that is, there is an oxidation along
the boundries of the ferrite grains. Other evidence has appeared
in which iron oxide, scattered through the body of the iron in the
form of crystals of magnetite, appear to have had a connection
with blistering phenomena.

14.   This rather contradictory evidence would appear to in-
dicate that other factors must operate in conjunction with the iron
oxide if blistering is to arise from this source. A comparison of
the oxide formation in Figs. 1 and 2 indicate that the oxide of
Fig. 1 is much finer and more spongy than the oxide shown in
Fig. 2. Thus, it is quite possible and probably true that the degree
of oxidation of the iron oxide and its form in the iron may be
factors in the blistering problem.

(B) Blowholes and Slag or Dirt Inclusions

15.   Most authorities in the United States agree that blow-
holes and slag or dirt play an important part in causing defects,
if present, in enameling iron. Their presence is objectionable and
indicate improper foundry practice. Angus states that he has
conducted numerous experiments in the porosity of enameling
iron by simulating blowholes artificially. However, it does not
seem reasonable to believe that a drill hole, although tightly
capped, will show the same characteristics as a natural blowhole
in the iron, since some blowholes contain gases, while others appear
to contain slag and iron oxide. Certainly the pressure conditions
inside of a blowhole would be difficult to simulate. The fact
remains that the presence of blowholes in blistering iron is more
common than in non-blistering iron.

16.   At times, it would appear that small blowholes may
cause more trouble than large ones. Thus, Fig. 3 shows a series
of inter-connected blowholes partially lined with iron oxide, which
caused blistering, while Fig. 4 shows a much larger blowhole,
which caused no blistering whatsoever. Although this blowhole
was almost 2 mm. in depth, the opening to the surface enameled
over perfectly and the defect was discovered by accident.

(C) Chill

17.   Most authorities agree that micro-chill structure in cast
iron for enameling is highly detrimental, since it may be the cause
of blistering, poor adherence, warping and excessive shrinkage
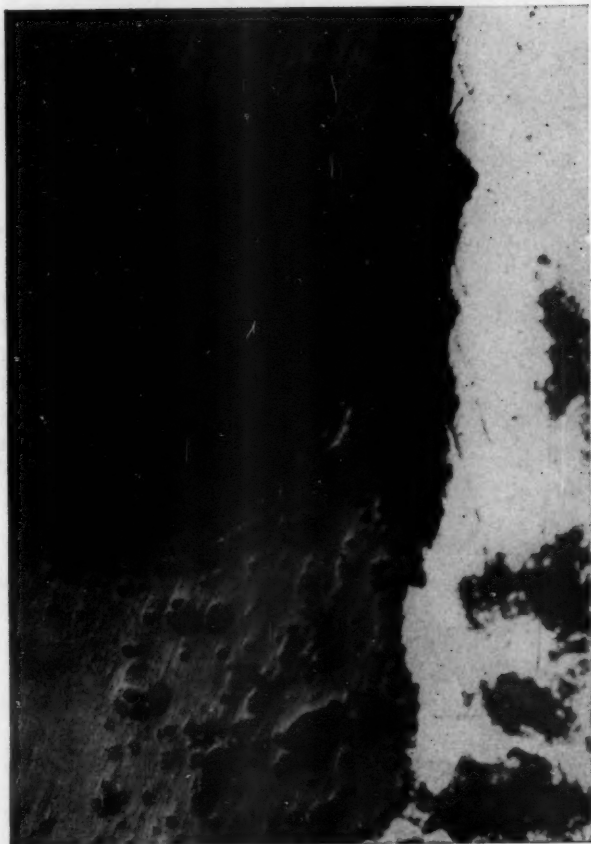
Fig. 8—Section through blistering area of enameled iron showing inter-connected, iron oxide lined blowholes. Note blister bubble above iron—Unetched—150x
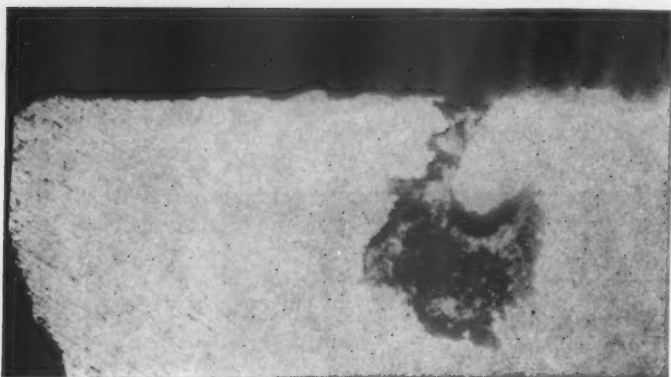
FIG. 4—SECTION THROUGH VERY LARGE BLOWHOLE WHICH CAUSED NO DEFECT ON ENAMELING.
BLOWHOLE DEPTH 1.9 MM.—UNETCHED—20X.

in the castings. The defect often is encountered in small castings
in very thin sections, such as stove grates. Fig. 5 shows an interest-
ing example of the effect of chill in such a stove grate. It will
be noted that about one-fourth of the area of the grate has blistered
but that the remainder enameled perfectly. Figs. 6A and 6B and
Fig. 7 show the microstructure of the blistering and non-blistering
areas of this grate. These sections would indicate a pronounced
chill structure in the blistering area.
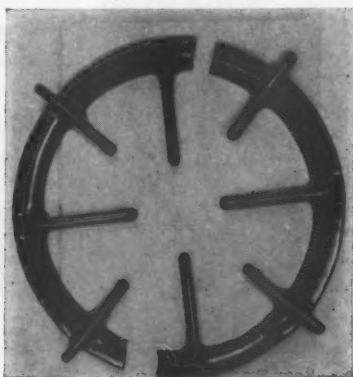
FIG. 5—BLISTERING GRATE DUE TO CHILL STRUCTURE. NOTE BLISTERING AREA IN TOP
QUADRANT. SECTION TAKEN FROM TOP AND BOTTOM OF GRATE IN AREAS INDICATED BY
SPLIT IN RING.
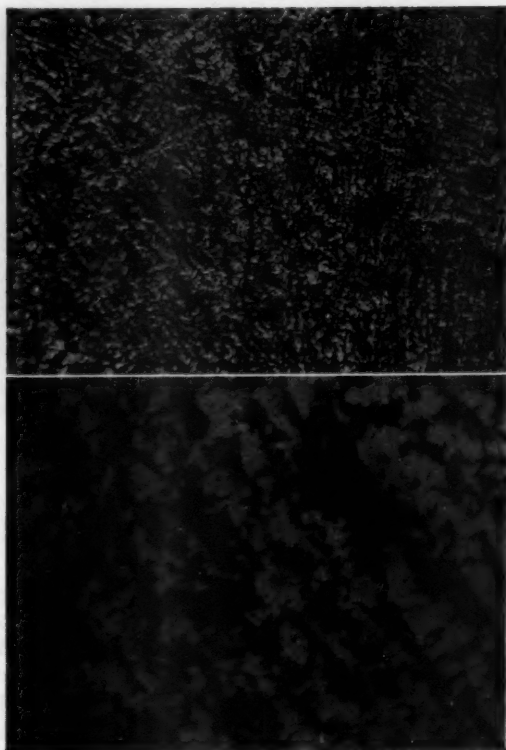
FIG. 6—(A, ABOVE)—CHILL STRUCTURE UNDER BLISTERING AREA OF GRATE SHOWN IN FIG. 5.
ETCHED ELECTROLITICALLY IN 10% CHROMIC ACID FOLLOWED BY 5% NITAL ETCH. 30X.—
(B, BELOW)—SAME STRUCTURE AS A EXCEPT AT 350X.

18. While there is complete accord over the undesirability
of macro-chill structures in enameling iron, there is one form of
chill, the importance of which is somewhat controversial. Krynit-
sky and Harrison state that the chief cause of blistering is micro-
chill — a thin chilled layer or skin at the surface of the castings
which, they report, is extremely difficult to remove by ordinary
sandblasting methods. Poste agrees in part with this contention
that some factor at the surface of the iron may be the cause of
blisters, in view of the fact that he has been able to improve many
blistering irons by removal of the surface layer. He further states
that apparently there are other factors which enter into the
problem.

FIG. 7—NON-BLISTERING AREA OF GRATE SHOWN IN FIG. 5. ELECTROLITIC CHROME ACID ETCH FOLLOWED BY 5% NITAL.—350X.

19.   Angus feels that the chief evidence of micro-chill appears to be circumstantial only and maintains that microscopic examination does not appear to uphold the presence of any special form of cementite in the surface layers. In place of the micro-chill theory, he substitutes his absorbed oxygen and iron oxide theory. While this might be true in the case of oxygen and iron oxide absorbed in the mold during cooling, it does not explain the experiments of Krynitsky and Harrison and Poste who claim to have eliminated blistering in some irons by mechanically removing a few thousandths of an inch of the surface of the iron.

20.   While the writer occasionally has encountered micro-chill in commercial castings, it has not been noted in anywhere near the majority of the blistering iron examined. He has, however, noted it quite often in the smaller test castings.

(D) Free Carbon and Graphite

21.   Most investigators maintain that a coarse graphite structure is more apt to be present in a blistering iron than a finer type of graphite. Some writers also hold specular carbon to be blister producing. Results of the study of a number of blistering and non-blistering commercial irons would indicate that coarser graphite does appear more often in blistering iron than finer graphite, that the presence of larger amounts of amorphous carbon in con-

junction with coarse graphite is more often found in blistering than non-blistering iron, and that the combination of very coarse and very fine graphite, especially when the fine graphite is located at the surface of the casting, often is found in blistering iron.

22.   Non-blistering irons usually show a finer and more uniform graphite structure than blistering irons. Very fine graphite structure, due to chill may be expected where blistering is due to such chills. Angus feels that graphite formation is a major factor in the blistering problem and reports that a fine graphite structure is more desirable than a coarse graphite structure. Fig. 8 is a good example of the combination of coarse graphite with fine graphite at the surface of a blistering casting. In this instance, the graphite has broken down in a rosette structure and the surface of the iron is spongy with fine graphite and amorphous carbon.

23.   The literature makes no mention of the relation of the pattern of the graphite to enamelability.

(E) Combined Carbon

24.   The presence of combined carbon as cementite in enameling iron admittedly is undesirable. Such iron practically always develops blisters. The combined carbon as pearlite, present in enameling iron, presents one of the most interesting phases of microstructure study of enameling irons. Since the enamel operation represents an annealing or decarburizing process, the pearlite usually will be either partially or entirely broken down during the enameling treatment.

25.   The rate of break down of the pearlite constitutes a vital
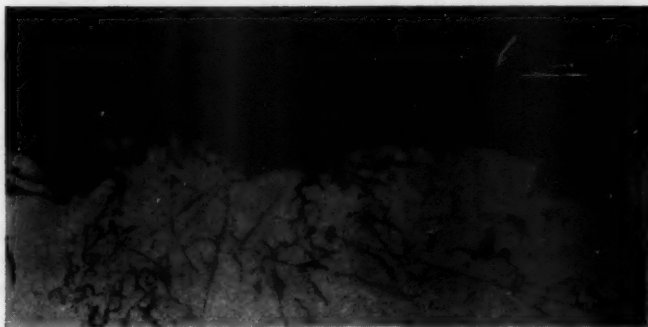
FIG. 8—BLISTERING IRON AFTER ENAMELING. NOTE SLAG POCKETS AND SPONGY FINE CARBON STRUCTURE AT INTERFACE. UNETCHED.—350X.

part in the question of enamelability. The problem is a complex
one, since the whole thermal and chemical history of the iron is
involved. Most investigators of the enamelability of iron have
agreed that an iron in which all the combined carbon breaks down
during enameling operations is superior to one in which decarburi-
zation only is partial. For this reason, a high silicon iron is speci-
fied. The reasoning is correct so far as it goes but does not entirely
solve the problem, since thermal history and section weight or heat
capacity of the iron are not considered. Thus, it is quite possible
for iron to decarburize completely but if the major portion of the
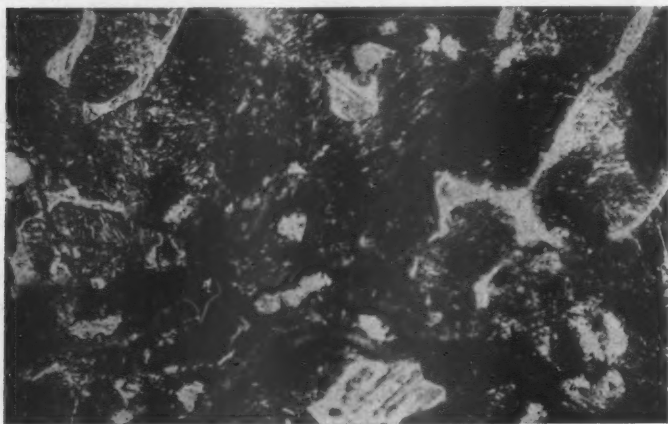
FIG. 9—NON-BLISTERING IRON AFTER ENAMELING. NO BREAK DOWN OF COMBINED CAR-
BON. ETCHED IN 10% NITAL.—850X.

process takes place after the enamel has begun to fuse, there may
be trouble.

26. Poste has shown, in a rather extensive investigation, that
the carbon dioxide gas given off in the annealing of blistering iron
usually is considerably less than that given off in the annealing
of non-blistering iron, indicating the delayed carburization may be
a factor. Some irons decarburize so slowly at enameling tempera-
tures that no difficulty is encountered in enameling them. Fig. 9
shows a photo-micrograph of such an iron. The chemical analysis
showed no decarburization and the enameling was perfect.

27. Some irons will decarburize at the surface only, while
others will decarburize partially more or less uniformly through

the casting. Occasionally, an iron is encountered in which the center of the casting will decarburize while the surface does not appear to do so. All of these cases present interesting possibilities of study to those who are interested in the mechanism of the formation of graphite and other phenomena in cast iron.

28. Cast iron enamels are fused at temperatures ranging from 1250°F. (677°C.) to 1400°F. (760°C.). The firing time usually is approximately 20 min., although in extreme cases, the time may be as short as 15 min. or as long as 30 to 40 min. Since these temperatures bracket the eutectoid temperature and since some metallurgists claim decarburization progresses very slowly below 1472°F. (800°C.), the mechanism of the decarburization of enameling iron should be of interest, and it is believed that careful and comprehensive studies of the subject should prove of value to both founders and enamelers of cast iron.

29. The relation between enamelability, decarburization rate and enameling temperatures is illustrated clearly in the case of an enameling iron recently studied by the author. As is the case with many similar problems, the thermal and foundry history of the iron is not available. The casting, a moderately sized stove top, was broken so that the sections were relative in size and proximity to the gates. Sections were made at various locations to give the

FIG. 10—UNENAMELED SECTION OF STOVE TOP CASTING WHICH BLISTERED WHEN ENAMELED AT 1250° F. BUT NOT AT 1350° F. AND 1400° F. NOTE ROSETTE PATTERN OF FERRITE AND GRAPHITE. WHITE LINES IN CELL-LIKE PATTERN IN PEARLITE MATRIX ARE STEADITE. ETCHED IN 10% CHROMIC ACID FOLLOWED BY 5% NITAL.—150X.

structure at the gate, fartherest from the gate and at the thinnest and heaviest sections, prior to enameling. The casting was then enameled with three enamels maturing respectively at 1290, 1350, and 1400°F. The 1290°F. enamel blistered over both heavy and thin sections — the blistering being more pronounced over the heavy sections. The pieces enameled at the higher temperatures did not blister.

30. Sections of the unenameled iron show a rosette graphite pattern, the rosettes of graphite and ferrite (See Fig. 10) being surrounded with pearlite, ranging in structure from coarse lamellar pearlite over the thin sections to both coarse pearlite and fine sorbito-pearlite over the heavy sections (See Figs. 11 and 12). Blistering sections showed residual pearlite in a cell-like structure over the heavy sections and as lamellar pearlite in the thinner section. Fig. 13 shows the structure of the blistering iron in the heavy section. After enameling, the non-blistering sections showed complete breakdown and loss of combined carbon.

31. The effect of the structure of the pearlite upon blistering has not been commented upon widely in the literature. Data at hand show a tendency for greater blistering with very heavy lamellar pearlite than with finer sorbito-pearlite. This is probably to be expected since the finer pearlite would present the greater surface area.

32. Available data would indicate that the pearlite structure of iron of a given composition depends chiefly upon the thermal history of the casting. Thus, when three sets of identical castings were poured together and one set was removed from the sand while very hot, in fact, as soon as the castings could be handled without distortion — a second was shaken out as soon as the core in the riser was black and the third set allowed to cool in the sand, it was found that, after enameling, the first set blistered while the other two did not. Table 1 shows the results of metallographic study of these cast irons.

33. Chemical analysis shows that the combined carbon in set *I* was only half removed in the enameling operation, while in sets *II* and *III*, the iron was practically completely decarburized. Figs. 14 and 15 show sections of Set *I* prior to enameling, while Figs. 16 and 17 show sections of set *III* prior to enameling. These data indicate that coarse graphite and coarse lamellar pearlite were obtained by rapid cooling, while fine graphite and pearlite

Fig. 11—Pearlite structure in thin section of stove top casting. Note coarse pearlite. Etched in 10% chromic acid followed by 5% Nital.—350x.
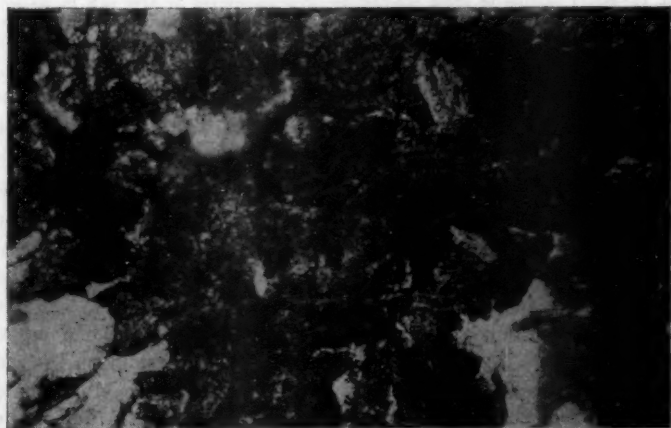
Fig. 12—Pearlite structure in heavy section of stove top casting. Note very fine structure. Etched in 10% chromic acid followed by 5% Nital.—350x.

FIG. 13—HEAVY SECTION OF STOVE TOP CASTING AFTER ENAMELING AT 1280 DEG. F. ENAMEL BLISTERED. THE HAZY INTERCONNECTED MATERIAL IS RESIDUAL PEARLITE SEEMINGLY ROSETTES OF FERRITE AND GRAPHITE. ETCHED IN 10% CHROMIC ACID FOLLOWED BY 5% NITAL.—250X.

structures resulted from slower cooling. The results indicate the possibility that slower cooling possibly might eliminate blistering in some types of iron.

34. The effect of pouring temperatures upon the microstructure of metals is well known and has been widely commented on in the literature. As Bolton so aptly states "Proper pouring temperature ranges have been worked out for a number of metals and it is quite common, in non-ferrous foundries especially, to see pouring temperatures controlled pyrometrically. In some cases, this is done also in steel foundries and perhaps least often in iron foundries." It might be added that practically no stove plate foundries have much of an idea as to what the temperature of the iron they are working really is.

35. Most foundries do use chills in the ladle when certain

Table 1

STUDY OF EFFECT OF COOLING RATE ON STRUCTURE OF CAST IRON FOR ENAMELING

Treatment	Set 1	Set II	Set III
Treatment	Shaken out as soon as possible	Shaken out when black	Cooled in Sand
Enameling	Blistered	O. K.	O. K.
Graphite	Very Coarse	Moderate	Very Fine
Pearlite—Unenameled	Coarse to moderate lamellar	Moderate lamellar	Very fine sorbito-pearlite
Pearlite—Enameled	Moderate Sorbito-Pearlite	None	None

Fig. 14—Unenameled section of iron of Set I, casting shaken out of mold as quickly after pouring as possible. Note coarse graphite.—Unetched 350x.
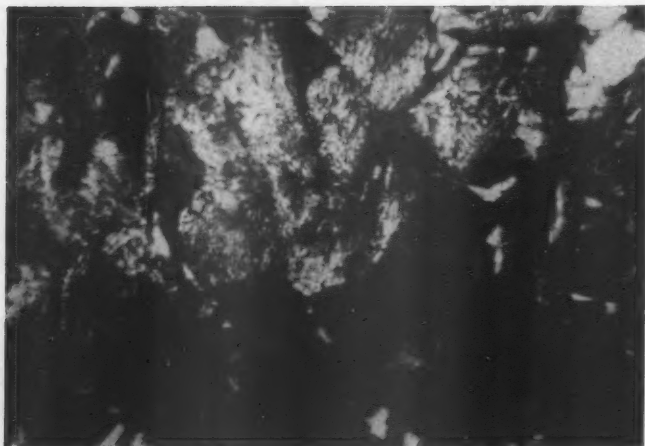
Fig. 15—Same as Fig. 14 etched in 10% Nital. Note coarse lamellar pearlite.—350x.

FIG. 16—Unenameled section of iron cooled in mold—Set III—unetched. Note fine graphite structure.—350x.

FIG. 17—Same as Fig. 16. Etched in 10% Nital. Note very fine pearlite and increased amount of ferrite present.—350x.

types of castings are being poured. Since very hot irons cool more slowly than colder irons, it is to be expected that graphitization will progress farther in the irons poured at the higher temperatures. This, of course, means that the combined carbon content of the iron will be lower and that the pearlite structure also might be expected to be different than is the case with the colder irons, since the mold will be hotter and cooling through the transition from austenite to pearlite will be slower.

36. No data are available as to the effect of pouring temperatures on the enamelability of cast iron with the exception of the fact that excessively hot irons are to be avoided because of burned in sand and the opinion expressed by several founders known to produce high grade enameling iron to the effect that hot iron will produce better enameling iron than colder iron.

37. It is a well-known fact that certain types of castings will give more trouble in enameling than others and it would seem quite possible that casting temperatures might have some connections with the matter. At any rate, the study of the relation of casting temperatures to the microstructure and enamelability of cast irons would appear to offer a field of investigation which may be expected to yield valuable results.

38. A final factor in the relation of pearlite to blistering, is found in the variation of section weight over the casting. Marked variations in structure often give rise to blistering due to the fact that the mass of metal in the heavy sections retards decarburization and also because of the gases formed in the heavy sections after the enamel has fused. Thus, if a piece showing wide variation in section be enameled to give the best job on the thin section if the iron has blistering tendencies, blisters will show up over the heavy sections; while if the piece is fired to get good results over the heavy sections, the thin sections will be overfired. The problem, therefore, resolves itself into a matter of so designing the casting to secure as uniform a section as possible.

39. The microstructure of residual pearlite, found in enameled cast irons showing blistering, varies considerably. Thus, one iron, showing initially a coarse lamellar pearlite, will show, after enameling, a fine sorbito-pearlite structure while another may retain its lamellar structure and merely show a dimunition of amount of pearlite remaining. As has been stated above, some irons show surface decarburization only, while others show uni-

form decarburization over the entire section. Data at hand are not adequate for forming any conclusions as to just what the effect of the method of decarburization in these cases is upon enamelability nor is it possible to predict from examination of the unenameled casting how it will react to the enameling operation as regards the type of decarburization. Study of the problem will require a considerably greater amount of data concerning the foundry operations than is at present usually available.

(F) General Comments on Blistering

40. While the effects of microstructure upon blistering defects have been considered from the standpoint of the individual components which go to make up the microstructure, it would appear reasonable to believe that many, if not most, of the troubles encountered are due to a combination of improper factors rather than to any single one. For example, some irons containing considerable iron oxide or slag will enamel satisfactorily, while others which appear to contain less will, due to the conditions of the other components, blister badly.

41. It is to be noted that the above considerations have not included steadite. Steadite always is present in enameling irons. The size and shape of the particles varying from fine, widely disseminated particles to large concentrations, depending upon composition and thermal history. No mention is made in the literature connecting steadite with the blistering phenomena. Boyles failed to find any connection between phosphorus and graphitization and most other authorities agree that the chief effect is the depression of the freezing point of the iron. The examination of many blistering enameling irons fails to indicate any connection between the steadite and the enameling defect as regards blistering.

Relation of Microstructure to Adherence

42. Adherence defects in enameled cast iron are not as prevalent as blistering defects, nor are they as closely connected with the composition and microstructure of the iron. Usually, improper design is a major cause and often improper furnace atmosphere during the enameling will be a factor. In one case, it was found that, with a slow decarburizing iron which blistered under the application of cover coat enamels direct on the iron, a coating of temper carbon was formed at the interface when a ground coat

enamel was used with the resulting poor adherence. While the
defect was due primarily to a leaky furnace muffle, the defects
were found to be more prevalent on certain types of castings, in-
dicating that the microstructure of the iron could have been an
accessory factor.

43. In one other case, a very heavy steadite structure at the
surface caused poor adherence. A number of other cases of poor
adherence have been shown to be the result of a chill condition, the
chill being in some cases micro but more often macro-chill.

RELATION OF MICROSTRUCTURE TO WARPING

44. The only cause of warping, of interest from the metall-
ographer's standpoint, is chill. Since molding practice and design
play a much more prominent part in the control of this defect, the
metallographer is seldom, if ever, called upon to deal with it. The
fourth major cast iron defect due to iron, hairlining, is the result of
strain set up due to the improper design and is not to be blamed
upon microstructural factors.

SUMMARY

45. In the study of the relations of microstructure to the
enamelability of cast iron, it would appear that blistering, the
major defect encountered, has received practically all of the atten-
tion of investigators studying the problem. While chemical com-
position and chill phenomena have received a great deal of atten-
tion and the effect of iron oxide, and to some extent graphite,
have been commented upon, the highly interesting and important
problem of the factors influencing the formation and decarburiza-
tion of the pearlite have been largely ignored.

46. Data at hand would indicate that:

(1) Macro-chill may cause blistering, warping, and poor
adherence.

(2) Micro-chill will cause blisters and occasionally poor
adherence. It is not believed to be quite as important a factor
as some investigators have suggested.

(3) Iron oxide may or may not cause blistering. The
vagaries of this component might make an interesting study.

(4) Data would appear to indicate that iron containing

very coarse graphite, especially if associated with temper carbon or very fine graphite at the surface of the casting, is more prone to blister than iron containing moderate or rather finely sized graphite in uniform proportion.

(5) Pearlite is believed to be a more important factor in blistering than has been believed in the past.

(6) Decarburization rate of pearlite as influenced by structure, variation in section of casting and temperature of enameling has an important effect upon blistering.

(7) Fine pearlite appears less apt to give rise to blistering than coarse, lamellar pearlite.

(8) Chill appears to be the major microstructural factor influencing poor adherence of the so-called "pop off" type. Decarburization rate and steadite content in some instances have been connected with this defect.

Bibliography

1—Krynitsky and Harrison *"Blistering Phenomena in the Enameling of Cast Iron,"* BUREAU OF STANDARDS, JOURNAL OF RESEARCH, vol. 4, June, 1930, pp. 757-808.

2A—Poste, Emerson P., *"Blistering of Cast Iron—Part I,"* BUL. AMER. CER. SOC., vol. II, pp. 201-213, August, 1932.

2B—Poste, Emerson P., *"Blistering of Cast Iron—Part II,"* JOUR. AMER. CER. SOC., vol. 16 (6), pp. 277-292, 1933.

3—Angus, H. D., *"Cast Iron for Vitreous Enameling,"* FOUNDRY TRADE JOURNAL, vol. 55 (1037), p. 12, 1936.

4—Malinovszky, A., *"Enameling Defects Due to Cast Iron,"* JOUR. AMER. CER. SOC., vol. 8 (1), pp. 72-78, 1925.

5—Krynitsky, A. I., *"Enameling Defects Due to Cast Iron,"* JOUR. AMER. CER. SOC., vol. 8 (10), pp. 618-622, 1925.

6—Spencer-Strong, G. H., *"Examination of an Unusual Cast Iron,"* JOUR. AMER. CER. SOC., vol. 21 (2), pp. 60-65, 1938.

7—*"Analysis of Cast Iron Suitable for Enameling,"* Abstract of Lecture by H. Cowan, FOUNDRY TRADE JOURNAL, vol. 58 (1120), p. 126, 1938.

8—Schall, R. B., *"Study of Metallographic Characteristics and Enameling Qualities of Certain Cast Irons,"* JOUR. AMER. CER. SOC., vol. 21 (1), pp. 24-27, 1938.

9—*"Production of Castings for Porcelain Enameling,"* Porcelain ENAMEL INSTITUTE, July, 1937.

10—Schall, R. B., *"Foundry Practice and Methods of Control of Cast Iron for Porcelain Enameling,"* ENAMELIST, October, 1936.

11—Moldenke, R., *"Principles of Iron Founding,"* McGraw-Hill (New York), Second Edition, 1930.

12—Boyles, Alfred, *"Freezing of Cast Iron,"* A.I.M.E. Tech. Publication 809.

13—Bolton, J. W., *"Gray Cast Iron,"* Penton Publishing Co. (Cleveland), 1937.

14—Parke, Crosby and Herzig, *"Graphitization of Gray Cast Iron,"* METALS AND ALLOYS, vol. 9 (1), p. 9, 1938.

15—Hurst, J. E., *"Metallurgy of Cast Iron,"* Isaac Pittman & Sons, Ltd., 1926 (London).

16—Sauveur and Boylston, *"Metallography and Heat Treatment of Iron and Steel,"* Cambridge, 1936.

DISCUSSION

Presiding: DR. JAS. T. MACKENZIE, American Cast Iron Pipe Co., Birmingham, Ala.

N. A. ZIEGLER[1] *(written discussion):* The problem of blistering of cast iron enamels is indeed an important one, and because of a very large number of variables involved, a difficult one to solve. The author apparently concentrates his discussion on the enamels fabricated by the wet process. Our experience has been almost exclusively with dry enamels, but nevertheless in view of many points of similarity, it is felt that with a few remarks about our observations may be of interest to the listeners.

In addition to the state of the metal, the nature of the enamel, itself, is at least equally as important, and it is very difficult to differentiate between the two in studying blistering phenomena. If an average enamel coating were examined under a microscope, one can readily see that it is impregnated with tiny gas bubbles. In the writer's mind it is not clear whether these bubbles are a necessary constituent of an enamel — i. e., without them an enamel would not be an enamel — or whether they are a necessary evil, which we do not know how to eliminate. It is equally uncertain whether these bubbles are generated in the enamel itself, or are evolved by the metal, or are a result of a reaction between the metal and the enamel. Perhaps all three of these phenomena enter into the picture. One should remember that any enamel is composed mostly of oxides and may well expect that their reaction with the carbon of cast iron is capable of generating considerable quantities of gas.

Leaving the subject of the nature of these gas bubbles, it is easy to visualize that, other factors being equal, there is a possibility for them in certain cases to flow together or coalesce, thus forming one large

[1] Research Metallurgist, Crane Co., Chicago, Ill.

bubble instead of several tiny ones. If this progresses far enough, a blister may result. In other words, the problem of eliminating blistering reduces itself not so much to elimination of gases (average enamel is full of them anyhow), but to maintaining them sufficiently finely and uniformly dispersed.

In the present writer's opinion, the nature of the enamel is a far more important factor in creating blisters than the nature of the metal. A short time ago in connection with one of our research problems we wanted to have some blistered samples made of our regular enamel by our standard process. It has been decided to do this by off-setting the manganese-to-sulphur ratio. Forty pound samples of molten cast iron were taken from the cupola at our Chattanooga foundry and treated by adding to them predetermined amounts of either ferro-manganese or ferro-sulphur. These samples were then cast into standard 4 in. x 8 in. test pieces, all of which have been enameled in the ordinary way, using standard Crane enamel. Although the manganese-to-sulphur ratio in these samples varied from 0.5 to 22.0, none of them showed any signs of blistering. The same was true about two standard lavatory bowls and two flat rim wash sinks, made of cast iron artificially enriched in manganese up to 1.5 per cent. It may be mentioned that the higher manganese test plates, as well as sinks and bowls were rather hard.

The main object of these remarks is to emphasize that, in studying blistering phenomena, the nature of the enamel seems to be far more important than the nature of the metal. Nevertheless, the nature of the iron should not be neglected and, therefore, any paper on this subject, particularly when presented as ably as that by Mr. Spencer-Strong, should be greatly welcomed.

In conclusion, I would like to ask one question. It is not clear in my mind what the author means by the word "decarburization" as used in pages 793, 794, and 795. Does he mean that carbon is actually removed from the metal through a chemical reaction with the oxygen in the atmosphere or in the enamel, or does he mean merely "disintegration," "breaking up," or "graphitization" of the free carbide and the carbide part of pearlite? If the latter is the case, the word "graphitization" seems to be more proper.

A. I. KRYNITSKY[2] *(written discussion):* The writer has reviewed this paper with great interest since he considers that the examination of structure is very important in the study of phenomena observed during enameling of cast iron.

In general, the paper confirms the working hypothesis advanced in the paper of the National Bureau of Standards, to which references are kindly made by the author, namely, that the carbon in some one or more of the possible forms (ordinary graphite, submicroscopic graphite, temper carbon or combined carbon) plays a major role in the formation of blisters. It is understood, of course, that carbon present in pearlite is but one of the forms of combined carbon. The decomposition of cemen-

tite to form "nascent" temper carbon or graphite will depend on: (a) the original content and form of combined carbon, (b) the composition of the cast iron as to elements that promote or hinder graphitization and (c) the duration and the temperature of the enameling treatment.

It is the personal opinion of the writer that from the fundamental standpoint there is no such thing as "non-blistering" iron. There is always a possibility of blistering with any cast iron, unless carbon in the iron is inactivated by some means.

The author's experiments with three sets of identical castings, poured together but cooled at different rates, are very interesting. The writer is somewhat handicapped in discussing these experiments because the chemical composition of these castings and location of the spots represented by the micrographs are not given. As far as it can be judged from Figs. 15 and 17 in the preprint, casting No. 1 shows the presence of free cementite and should contain a rather large amount of combined carbon, whereas casting No. 3 shows, as is indicated by the author, the presence of ferrite, and should contain less combined carbon than casting No. 1. If this is true, the evidence upon which the conclusion that "fine pearlite appears less apt to give rise to blistering than coarse, lamellar pearlite" is based, may as well be subject to a different interpretation.

It is rather surprising that casting No. 1 (Fig. 14), which was cooled more rapidly, possessed a much coarser graphite than casting No. 3 (Fig. 16) which was cooled at a slower rate.

The writer believes with the author that the pouring temperature is very important but also considers that the maximum heating temperature is probably of equal importance.

MEMBER: Has any consideration been given to the use of a small amount of carbide stabilizing alloy, the alloy being inexpensively used along with sufficient amounts of silicon or graphitizing elements, thereby giving a more uniform graphite flake size and distribution, and in stabilizing the matrix so that it would not break down.

DR. SPENCER-STRONG: There are a number of foundries which are at present experimenting with that very idea, which is a very good one, in view of the scrap situation and other factors which enter into the problem. A number of authorities feel that the pearlite should break down entirely, but it has been my experience that if we could prevent the pearlite from breaking down at all, the chances of blistering would be much lessened.

MEMBER: In the slow cooling of the casting in the mold, a fine type of pearlite formed, while in the more rapid cooling due to shaking the casting out immediately after solidification, a coarser grade of pearlite was formed. Is it not true that just the reverse happens on that type of cooling?

DR. SPENCER-STRONG: I cannot say as to that. I can merely report my findings. The behavior of pearlite in enameling irons is very peculiar. Sometimes you start with a heavy lamellar structure and end with

a fine sorbito structure. This, to my mind, is exactly the opposite to what should happen, but, nevertheless, we find that to be the case. It does seem that there is a tendency in these irons which have had the longer heat treatment for the pearlite to be in a finer structure. It is because the pearlite is starting to break down. In these irons described the pearlite had half broken down in the final specimen that had the very fine structure and that may be the cause of the appearance.

MEMBER: I have had a problem in the blistering of castings and in the process of trial and elimination I have found that a high sulphur iron seems to react better in the enameling process than other irons do. I would like to have Dr. Spencer-Strong tell me if they have had any experience with that kind of iron.

DR. SPENCER-STRONG: There has been considerable discussion among manufacturers of enameling irons concerning the problem of the sulphur content of the iron. The literature for a long time has cursed sulphur and a number of authors have agreed that it should be eliminated entirely. However, more recently the practice has been to increase the sulphur content so that it is between about 0.08 and 0.09 per cent. If you go too high on it, you may get into trouble, but I believe it does have some desirable properties if present in, say, 0.08 or 0.09 per cent.

G. H. SPENCER-STRONG (authors written closure): The comments of Mr. Ziegler were appreciated, and as Mr. Ziegler has surmised, the larger part of the data contained in the present paper was the result of work on wet process enamels.

Due to the fundamental differences in the two processes, it appears to me that the wet process enameling of cast iron may be somewhat more susceptible to defects than the dry process. Mr. Ziegler has stated that he feels that the enamel may be more likely to give rise to defects than the iron. While the author has insufficient data to either prove or disprove this assertion in regard to the dry process field, in the wet process field, where commercially smelted frits are largely used, the reverse is found to be true. It is true that some types of wet process enamel are more susceptible to defects than others, however, the improvement of inspection standards in the wet process field, which are largely due to the improvements in the sheet iron enameling field, have given rise to complaints about defects which, fifteen years ago, would have been considered inconsequential, and it has been found that a considerable part of these defects may be attributed to the iron.

As Mr. Ziegler states, all enamel coatings are impregnated with microscopic gas bubbles, and unfortunately, these appear to be a necessary evil, since they are caused either by gases, such as fluorides, emanating from the enamel process itself, or are due to air or water vapor trapped during the enameling process.

The author has never seen an example wherein these bubbles have coalesced of their own volition to form blistering of the type encountered with iron defects. There is, no doubt, a reaction between certain types of enamel, and the graphite on the surface of the casting, which

forms due to the breakdown of the combined carbon, during the enameling operation. Under certain conditions, this reaction may take place during a time in the enameling operation when it does not give the enamel coating time enough to heal over the blisters in its surface which are formed by the resulting gases.

The author was interested to note the experiment carried on by Mr. Ziegler, in which he attempted to cause blistering by offsetting the Manganese-Sulphur ratio. In regard to Mr. Ziegler's comment concerning the meaning of the word "decarburization," there is probably some question as to the actual mechanism of the reaction which takes place on the surface of the iron during the enameling operation. However, the author agrees that the major part, if not all, of the breakdown of the combined carbon, is actually the graphitization of the free carbide, rather than a true decarburization process. The word was used more in contradistinction between graphitization, as it is often thought of; that is, as the breakdown of the combined carbon during the cooling of a casting after pouring, rather than in the present case which, of course, transpires during the enameling operation.

The author was especially pleased to read the comments of Mr. Krynitsky, since he has done a great deal of work along similar lines.

The three castings upon which these experiments were carried out show the following composition:

	No. 1	No. 2	No. 3
Total Carbon	3.64	3.64	3.56
Combined Carbons	0.48	0.37	0.24
Graphitic Carbon	3.16	3.26	3.32
Silicon	3.06	3.06	3.14
Phosphorus	0.35	0.35	0.37
Manganese	0.56	0.57	0.62
Sulphur	0.57	0.59	0.52

It might be stated that the samples from which the sections were made, were taken from the some position on each casting.

The author is in accord with Mr. Krynitsky's statement that he feels that there is no such thing as a non-blistering iron, since experience has shown that often, where difficulties are encountered using an enamel of one fusibility, if an enamel of a different fusibility is substituted therefor, the blistering defect may be eliminated.

The field covering the relation between the pouring temperature and the enamelability of the cast iron, and the maximum heating temperature of the iron and its enamelability, which are mentioned in the last paragraph of Mr. Krynitsky's comments, is almost untouched so far as research goes, and offers wide and interesting possibilities for future work and development.

It is believed that if the various phases of the manufacture of enameling iron were to receive as much attention as has been given to the engineering phases of cast iron, that the use of the product as a basis for enameling would be materially increased.

Pattern Development and Molding Methods

By E. T. Hynan[1], Saginaw, Michigan

Introduction

1. This paper is a story of some of the developments that have taken place at the Saginaw Malleable Iron pattern shop during the past 20 years.

2. Most noteworthy is the development of small hand plates to machine equipments. Improvement of molding machines and improved pattern designs have enabled us to run nearly all heavy production jobs as machine equipments. Machine patterns were developed to a point where most jobs can be operated as drop plates, doing away with the rollover and stripper frame types.

3. A checking department carefully checks all new pattern equipments prior to release for production, and no production is run until the castings are approved by the customer.

4. Much research work has been done to develop steel and bronze metals to replace aluminum in constructing patterns.

5. The demand for malleable iron castings in other industries has given our wood pattern section much training and experience. Diesel pistons and engine parts are the most prominent in this respect.

6. Corebox maintenance has been reduced through extensive use of vents and narrow brass strips on the outline of the joint faces.

7. Despite the yearly model changes, we have been able to standardize most of our equipment, especially flask sizes and molding machine frames.

8.   We constantly are adding and developing attachments and devices to our present equipment. This is done mostly to our molding equipment where small conveniences aid greatly in keeping up production.

THE BEGINNING

9.   At the time Barney Oldfield established a record of 90 miles an hour at Daytona Beach, there was one car for about every 150 persons in the United States. These 600,000 horseless carriages, with few improvements in 1911, began an era of development that made the product appealing to the buying public. Bumpers, doors, and battery ignition made their appearance.

10.   To C. F. Kettering's invention, the electric starter, should go a major share of credit for opening a new field of industries which enabled the automobile to progress so fast that people expressed the opinion that too many cars would be made and overproduction would result.

11.   Instead, new inventions and improvements followed, thus opening a new period of economic and social advancement. Thousands of new jobs were created and a higher standard of living established. The automobile, which was considered a luxury, soon became a necessity. Years later, other forms of transportation progressed, mainly through the influence exerted by the automobile. The Diesel engines used on locomotives and ships are outstanding examples.

12.   The development and use of various steels and alloy metals caused new foundries to spring up, and those who were actively engaged in producing various castings on a small scale, began to concentrate on automobile parts. Advanced and new production methods taxed skilled labor. Patterns had to be made to handle increased production. These required much time and created many problems. Pattern shops in foundries were merely maintenance and repair departments. Most patterns were made in job shops that were not thoroughly familiar with specifications demanded by automobile manufacturers.

13.   The Saginaw Malleable Iron plant, at the time of its organization in 1917, was no different than other plants of its kind. The small inadequate pattern shop made only a few match plates and loose patterns. There main products at this time were heads and bases for 3-in. trench mortar shells. At the conclusion of

the world war, the foundry began to produce automobile castings.
Some stove plate still was made, but auto parts were given first
consideration.

PAST PRACTICES

Patterns

14.   At this time, castings were made from small aluminum
match plates. The plate and pattern were cast separately and
mounted together by bolts. Complicated and intricate core jobs
were avoided. Car manufacturers had to design castings to suit
foundry conditions. Patterns were not developed beyond the point
of straight draft angles. This condition existed for a number of
years until these problems were overcome.

15.   The manufacturers of molding machines were making
progress and introduced the jolt-squeeze stripper molding machine.
Patterns were being built and developed for use on these machines.
The extensive use of the stripper type equipment, the addition of
the roll-over molding machine, and the introduction of conveyors,
gave the foundry pattern shop many new problems. The cope and
drag patterns were cast separately, hand worked and mounted on
stools which were bolted to the molding machine jolt table. The
stools were machined to the exact contour of the pattern. The strip-
ping frames were machined to the shape of the stools and pattern,
and the gating mounted on the frame.
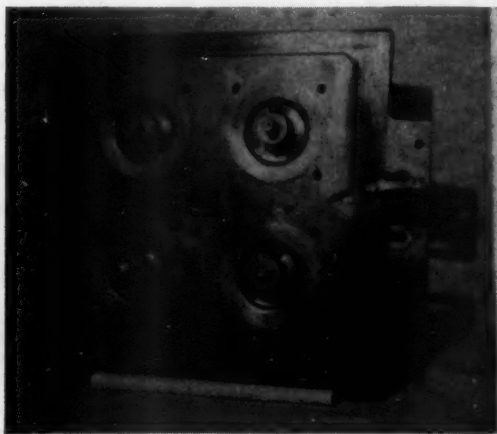
FIG. 1—OLD STYLE MATCH PLATE PATTERNS CAST IN SEPARATE HALVES AND BOLTED TO-
GETHER ON PLATE.
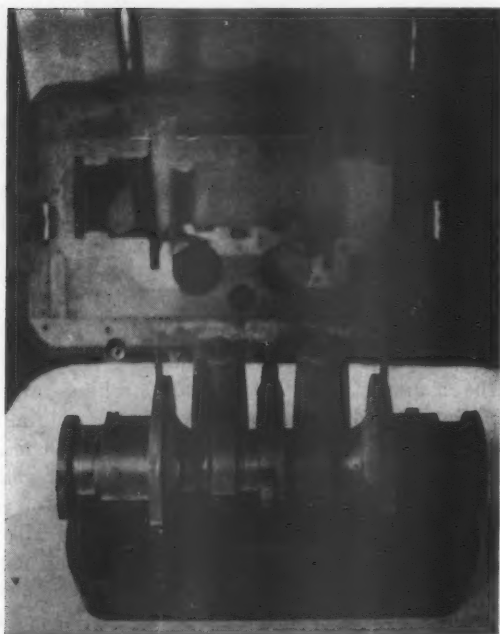
FIG. 2—STRIPPER TYPE EQUIPMENT. COPE AND DRAG CAST SEPARATELY MOUNTED ON STOOLS. STRIPPER EQUIPPED WITH SHIFT BLOCKS TO OVERCOME SHIFTS. PATTERN OUTLINE IN FRAME LINED WITH BRASS STRIPS TO COMPENSATE FOR WEAR OF STRIPPING FRAME. THIS TYPE OF EQUIPMENT WAS MOSTLY USED FOR HUBS AND CARRIERS.

Difficulties Encountered

16. The fact that the stripper was located off the pattern and the flask in turn located off the stripper, provided a good chance for shifts. Excessive wear on the stripping frame and patterns caused no end of trouble. Shifts, castings undersize at parting, or dirty and deformed castings were caused from stripping frame being set too high or too low.

17. The cope stripping frame was provided with shift blocks which carried the flask pins. These were shifted in any lateral direction, removing the fault rather than the actual cause. They later were discarded because of the inability to keep them in a fixed position.

Repairs

18. To compensate for the wear on the stripping frame, as

soon as the opening became large enough for fine sand to fall through or a fin of sand to show on the parting of the mold, the stripping frame was milled out ¼-in. deep around the contour of the pattern and lined with brass. These brass strips, which had to be worked in by hand each time, were replaced periodically as excessive wear developed.

19. The patterns also had to be chipped out at the parting and patches inserted to bring them up to size due to wear.

20. Since nearly all pattern equipments were built in job shops, very little thought was given to engineering improvements. The shop that received the order constructed the patterns on a standard scale from past experience. All they were given was a customer's print marked with a few instructions for gating. They used their own judgment on fillets, radii, finish, and assembly.

ESTABLISH PATTERN-ENGINEERING DEPARTMENT

21. One of the greatest improvements in pattern advancement at the Saginaw Malleable Iron Plant was the organization of the pattern-engineering department. This department made assembly drawings showing detailed pattern layouts known as the foundry sheets. The job shop still received the customer's print showing detailed dimensions. We expected them to conform to this print and incorporate in the construction everything that was given them on the foundry sheet.

Pattern Manufacturers Routine

22. Upon receiving a pattern equipment from the job shop, samples were made and sent to the customer, who checked them. A written report on the check was sent to us. It was either an acceptance or criticism of dimensions not up to print. The dimensions not according to specifications were corrected in our own shop at our expense. The cost for this work ranged from $10.00 to $200.00 or more, depending upon the type of equipment and number of corrections made.

23. Due to this condition, there was little or no chance to duplicate an equipment when making identical patterns. Corrections and delays became costly, since it required at that time from ten days to two months to put a pattern equipment into production.

NEW ACCEPTANCE STANDARDS

24. To eliminate this costly and uncertain procedure, a new method of accepting pattern work from job shops was originated. During the summer of 1933, when patterns were being received for 1934 models, a match plate of eight on, picked at random, was laid out with dimensional lines. Both sample castings and plate were checked according to print, and if not up to specifications, a written copy of the criticisms was sent to the production manager and the purchasing department stating that the plate was rejected. The same procedure was followed on other pattern equipments and criticisms recorded.

25. A representative of the job shop, whose plate was rejected, was called in and the new policy explained to him. He was told that we expected him to take back the plate and rectify all errors. This procedure was such a radical change from our past attitude that a mild objection was raised. The jobber stated that it could not be done as we expected. He was then informed by the management that in the future any pattern equipment made by his or any other job shop, would have to come to our plant ready to run in production. Any work that was required for corrections would be done by us and charged to him, or he could take the pattern back to his shop and make the changes at his expense.

26. This decision on the part of the management was a revolutionary one, and probably the most decisive, as far as our pattern department was concerned. All jobbers making patterns for us were invited to visit our plant and the new policy explained to them. They asked various questions as to how they would be able to do this with the information they received from us up to this time. This matter was taken care of whereby all detailed information needed to properly construct a finished product would be supplied by us. This would include the usual customer's print, and a complete set of detailed drawings showing every dimension necessary for the finished pattern. No small details were left to the judgment of the patternmaker.

27. The draftsmen of our pattern-engineering department had a big task before them. They worked long hours for three or four months to complete layouts for new model equipments. When this was completed, they started another big job by collecting data and breaking down the design of all the equipments they had just detailed, so that a group of standard sheets could be

set up which would simplify their work in the future and provide everything, except actual pattern dimensions for the new models.

ENGINEERING STANDARDS ADOPTED

28. Standards set up by the engineering department easily were adopted by us. With the job shops, the arrangement was somewhat different. We were asking them to change completely their way of doing business with us at a short notice. This may have been a drastic measure, but a very important one from our point of view. We possibly did go to extremes on some dimensions or minor details in order to impress the patternmaker with our standards. At the same time, we made it plain that the jobbers could send their men to our plant to acquaint themselves with our methods and practice. In this manner, they could get a definite idea of what we wanted.

CHECKING DEPARTMENT

29. A checking department was established to check patterns and coreboxes before sample castings were shipped to the customer. By doing so, we were able to put into production equipments on new models as soon as they were received. Previously, we had to wait for a customer's report on the samples and act according to their instructions.

30. If the customer rejected the samples, we would check the equipment and notify the jobber to make the changes. If the error was a small one, we would make the correction and submit new samples. Sometimes new patterns needed minor rearrangement of gating and bobs to suit our foundry conditions. Again, samples would have to be made. Each change would take time, and often a new pattern equipment would be a patched up job, even though never run in production. Maintenance of a pattern in this condition became very high.

DEVELOPS PATTERN MATERIALS

31. Because of the excessive abrasive and corrosive action on the pattern and stripping plate equipment, considerable research work was done to find a suitable material for machine equipment patterns. Aluminum, brass, and gray cast iron had been used extensively to date, with little thought given to the specific analysis of the material. Molding machine operators in the foundry

were constantly hitting, nicking and breaking patterns through careless handling of steel flasks. We endeavored to find a hard material or a metal that could be hardened.

Alloy Gray Iron

32. A trip was made to Detroit to a foundry that specialized in a standard analysis alloy iron that they claimed could be machined to size, heat-treated, and quenched with a minimum expansion and distortion. The analysis of this iron was approximately 24 per cent chrome, 2 per cent carbon, 1-2 per cent silicon, 0.75 per cent manganese and 0.07 per cent sulphur.

33. Test bars and a sample hub pattern casting were obtained for experimental purposes. The test bars were turned for machinability and finish. The size was carefully checked and recorded. They were then heat-treated and quenched in oil. When checked, after heat-treatment, it was found that the expansion was about 0.001 in. per in. The sample hub casting of this high chrome iron alloy was treated accordingly and found satisfactory.

Tool Steel

34. A tool steel casting was machined and heat-treated in the same manner, but the results were somewhat different. Distortion and growth were very noticeable. Finish grinding was necessary, and a large shrink crack developed from hardening.

FIG. 3—DIFFERENTIAL CASE PATTERN OF HIGH CHROME IRON, MOUNTED AS DROP PLATE EQUIPMENT. HEAT-TREATED AND QUENCHED IN OIL. THIS WAS A ROLL-OVER EQUIPMENT.
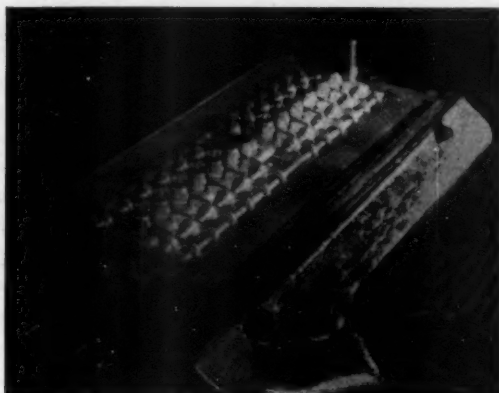
FIG. 4—TWENTY-FOUR ON ROCKER ARM—BRASS PATTERN. OPEN END DROP PLATE. COINING DIMENSION AND 4° RADIAL DRAFT ON BOTH SIDES OF BARREL HELD TO MICROMETER DIMENSION.

High Manganese Tool Steel

35. A piece of high manganese tool steel bar stock was purchased and a hub pattern turned from this. The object of the high manganese was to reduce expansion after hardening. This point proved successful, but there was too much distortion, which made finish grinding necessary.

36. Another pattern of ordinary gray iron was finished completely and the four experimental patterns mounted on a stool and put into production for a test period of four months.

RESULTS OF TESTS

37. The high chrome gray iron alloy showed the best result. The tool steels showed a reaction to acid and rust. The ordinary gray iron showed excessive wear. We purchased seven additional chrome iron castings and assembled two complete machine equipments.

38. The hub being a job of simple straight lines, made it comparatively easy to machine finish, but to apply this metal to more intricate and irregular patterns where hand finish was necessary, much more thought had to be given to wood patterns, allowing only about 0.010 in. for finish.

39. A differential case pattern, allowing this small finish, was tried and proved successful. Other patterns were made, and after much hard usage over a long period of time in our foundry,

are in good condition. Very little care or maintenance was given
them, proving that this alloy iron, though expensive, is very
practical.

MATERIALS FOR DROP PLATE EQUIPMENT

40. Although alloy iron patterns served satisfactorily, we
were doing much research work to develop drop plate equipments,
standardize steel flask sizes, eliminate roll-over molding machines,
and standardize the general assembly of all processing equipment.
Our first step was to develop a bronze metal that would withstand
the hard usage of foundry production, and replace aluminum
which was used extensively up to this time.

41. Our engineering department specified for brass patterns
80 per cent copper, 10 per cent tin and 10 per cent zinc. As
equipments were received, we began taking analyses of the metal.
Of the six tests taken, none came up to our specifications.

42. We prepared six different formulas and made a small
pattern that would give us about 15 different conditions that
would normally be encountered in molding, machining, hand scrap-
ing and finishing of a production pattern. These included different
metal thicknesses, radii, angles, flat surfaces, square corners, and
straight sides. The pattern dimensions were accurately recorded
and a template made so we could check the shrinkage.

43. All the data from this work was recorded so it could be
referred to from time to time. The record included molding pro-
cedure, composition of each alloy, and the results of each test to
determine its physical qualities, these being machinability and ease
with which the metal could be hand worked. Various methods of
gating and pouring were tried. A complete analysis of each alloy
was made to check melting loss. A uniform heat control was main-
tained at all times. To determine abrasive wear, test lugs were
made and mounted on production patterns in the foundry. The
formula that suited our needs best was found to contain 80 per cent
copper, 12 per cent tin, 4 per cent lead and 4 per cent zinc. All
job shops making brass patterns for us are asked to adhere to this
formula.

DROP PLATE EQUIPMENT DEVELOPMENT

44. The use of the roll-over molding machine was avoided in
building new pattern equipments. It was slow, awkward, and taxed
the strength of the operator. The stripper type of equipment was

more satisfactory, but needed too much maintenance. The consideration of drop plate equipments brought much discussion on sand conditions, and a more accurate rebuilding and repair of molding machines. The management decided to build a drop plate pattern equipment, and completely rebuild two molding machines to close limits.

Pattern Equipment

45. This drop plate equipment involved a slightly different principle in pattern construction. The pattern is cast integral with the base and the parting lines machined on the pattern where possible. The flask pins and bushings are located on the pattern as well as the runners and bobs which all strip out with the pattern. The ragged parting line with its costly repair was eliminated; standardization of equipment was made simple; match plate jobs could be converted easily into multiple pattern machine equipments; and the time for mounting and dismounting patterns was cut down considerably.

46. The first job was put into production and results found were so satisfactory that plans were made to adopt this method for further use. This move did away with the roll-over molding machines and limited the stripper type to a few jobs. To-day, we have only one stripper pattern equipment in production.

REBUILD MOLDING EQUIPMENT

47. The development or improvement in methods of any shop practice, as was the case of the two rebuilt molding machines, brought a demand from the foundry superintendent and the management that we completely revise our methods in rebuilding and repairing our molding machines. Previous to this time all of the reconditioning was done in job machine shops. Although the work was supposedly done according to our instructions, the results proved unsatisfactory. Molding machine operation in the foundry was very irregular. The stops for repairs resulted in a big loss of production.

REPAIR MACHINERY PURCHASED

48. We started in a small way to equip our shop for rebuilding these machines. We had to construct fixtures, jigs, plug and snap gauges, and make patterns for necessary repair parts. Various machine tools were purchased. A very rigid inspection on all parts was maintained. Later we purchased a 72 in. boring
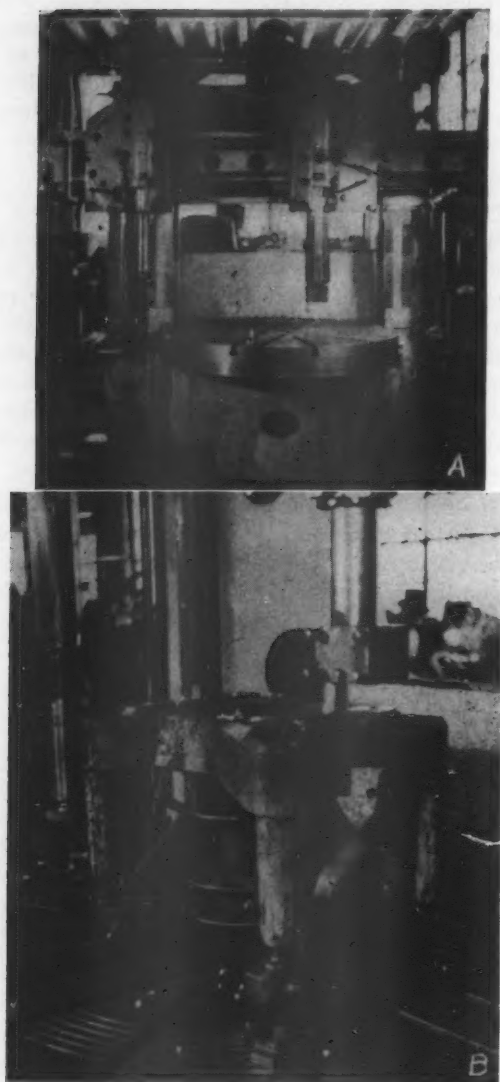
FIG. 5—A-72-IN. BORING MILL FOR GRINDING AND BORING MOLDING MACHINE PARTS. B—
EQUIPPED WITH VARIABLE SPEED GRINDING ATTACHMENT.

mill with a variable speed grinding attachment. Also a 16 in. x 72 in. external cylindrical grinder. These machines were used in turning and grinding molding machine squeeze cylinders and squeeze pistons. For grinding bushings, we installed a Heald internal grinder. Where necessary, all parts are hardened and finish ground to snap or plug gauges and Rockwell hardness tested.

DEVELOPS JOLTING AND SQUEEZING TIMING DEVICES

49. After establishing a standard for molding machine repair, we improved and added special features to suit our foundry conditions. An automatic timing device was perfected for jolting and squeezing molds. Instead of the operator determining the number of jolts required for each mold, this mechanism gives the mold the same number of jolts and a uniform squeeze each time. It can be adjusted to give any number of jolts desired.

50. Another improvement was the use of an automatic air draw stop. The draw rods previously came to an abrupt stop. Now they are supported by air, giving them a cushion effect.

51. A further improvement was the use of an automatic slow-up valve, a small device which has helped considerably in reducing the breakage of molds due to fast lowering of the machine table after the squeeze operation. When the air was exhausted from the squeeze cylinder the mold dropped fast until it stopped at the draw position with a bump. The new device drops the mold fast for a definite distance, then the exhaust air from the cylinder is automatically slowed so that the mold comes down gently into the draw position.

TEST MOLDING MACHINES

52. Much thought was given to the purchase of machines for future use. We asked representatives of various companies to visit us. They were given much information about our needs, practices and methods. Some showed an interest in our ideas, others did not. Machines from five manufacturers were purchased and put into operation. An actual check was kept on all lost time, maintenance, and molds produced per hour. After studying the records, we selected one which we thought would serve our needs. A number of these have been purchased and are now being used.

53. In our study of different molding machines, we found that there is a difference in opinion of various limits and tolerances used in the construction of machines by various manufacturers.

STEEL MOLDING FLASKS

54. Included in the standardization set-up are steel molding machine flasks and hand plate, snap-flasks. These are now limited to a few sizes, which take care of all jobs. For machine equipment, aluminum or wood squeeze heads are used. Permanent sprue cutters are attached to all cope heads. Each is so constructed that it can be adjusted to perfect alignment over the pattern and flask.

COREBOX CONSTRUCTION

55. Two noteworthy developments have taken place in corebox construction, especially those used in blow machines. First, it always had been a problem to get the proper degree of hardness on cores of irregular shapes. All corners would be soft or not properly filled. By placing screen covered vents in these places to allow the air to escape, the cores can be blown uniform throughout.

56. The other, due to high production which reaches as high as two hundred insertions and blows making 3300 pin cores per hour, caused the box to wear quickly on the parting face. After constant repairs, the boxes became undersize or out of round and had to be replaced. To overcome this, the face of half is milled
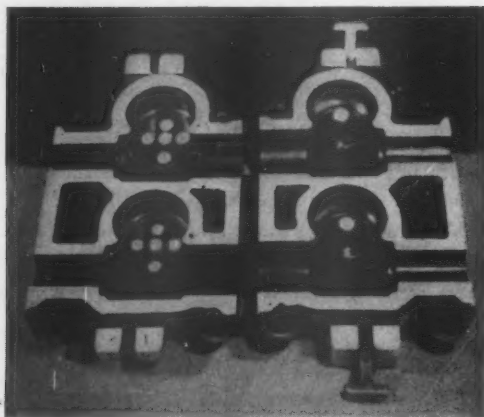
FIG. 6—ALUMINUM TWO-GANG HOUSING COREBOX. WHITE OUTLINED PARTS REPRESENT BRASS STRIPS 0.0625 x ½-IN. FOR BLOW MACHINE WORK, THIS FEATURE INSURES A TIGHTER COREBOX SQUEEZE AND MAKES REPAIRS EASIER. ROUND WHITE PARTS ARE SCREEN VENTS WHICH ALLOW AIR TO EXHAUST FROM THE BOX MORE READILY. THIS RELIEVES THE PRESSURE ON THE BOX, ESPECIALLY THE JOINT FACE.

FIG. 7—8½ IN. DIAMETER DIESEL PISTON AND CORE.

off 0.0625 in. A strip of brass 0.0625 in. thick and ½-in. wide is cut to the exact contour of each core outline and tacked to the face of the box. The brass is repaired easily and when worn from repeated spotting can be replaced.

WOOD PATTERNS

57. The wood pattern section of the shop has developed rapidly, due to our close association with the General Motors Research Division, in building patterns for research and the develop-
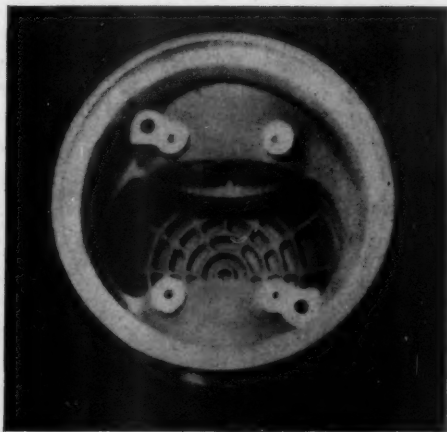
FIG. 8—TOP VIEW OF PISTON SHOWING INTERIOR CONSTRUCTION.

FIG. 9—FOUR SECTION, THREE-GANG PIN COREBOX. COPE HALF AND TOP OF 4-IN. PISTON COREBOX SHOWING RIBBING CONSTRUCTION, ESPECIALLY IN TOP SECTION.

ment of Diesel engine parts. Pattern equipments have been constructed for about fifteen different sizes and designs of Diesel pistons, ranging from 3 to 8.5-in. diameter. Piston patterns also were constructed for some car divisions, all being cast in "pearlitic malleable."

58. The patterns had to be held to very close dimensions on all inside diameters, because all machining operations were indicated from the inside of the casting. The castings were carefully checked in machining and held to a plus or minus 0.005-in. on the rough casting. The wall thickness of the finished product was 0.032-in. plus or minus 0.005-in. In piston work, several types of core designs were developed, which previous to this time were unknown, or never tried in our plant.

58. Our department incorporates wood and metal patterns, coreboxes, molding machine repair, flask maintenance and processing dies, everything that has to do with close dimensional castings. The production and maintenance of these things demand skilled workmen of a versatile type.

TRAIN MEN

59. Feeling that the primary requirement of an efficient department is competent and satisfied workmen, our method is to

FIG. 10—A—A BRASS DROP PLATE EQUIPMENT OF IRREGULAR LINES AND DEEP POCKETS.
B—NOTE THE CONSTRUCTION ON THE BACK OF THE PATTERN ELIMINATING EXCESSIVE
WEIGHT AND METAL.
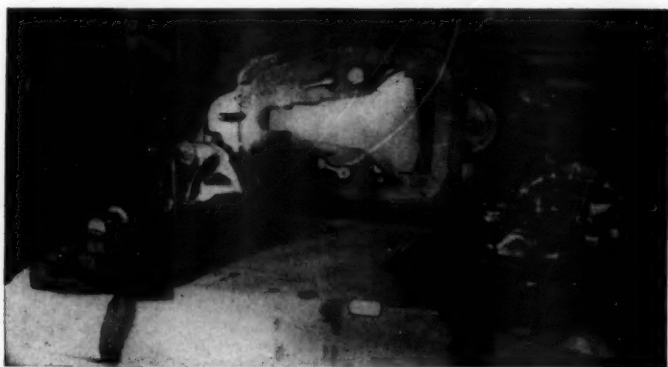
FIG. 11—CARRIER-STRIPPING FRAME EQUIPMENT. IRREGULAR PARTING LINES; NUMEROUS
PATCHES ON PATTERN AND FRAME; GATING MOUNTED ON STRIPPING FRAME.

FIG. 12—CARRIER—DROP PLATE EQUIPMENT. PATTERN AND GATING MOUNTED ON DROP PLATE.

select young men who have had some technical or trade school training and plenty of initiative who want to become expert mechanics.

60.   At present, we have about 40 young men who were hired for the shop or transferred from some other department because of their ambition to better themselves.   We feel that with this type of workmen, who are not classed as pattern makers or toolmakers, we have a better chance to use them where we see fit to
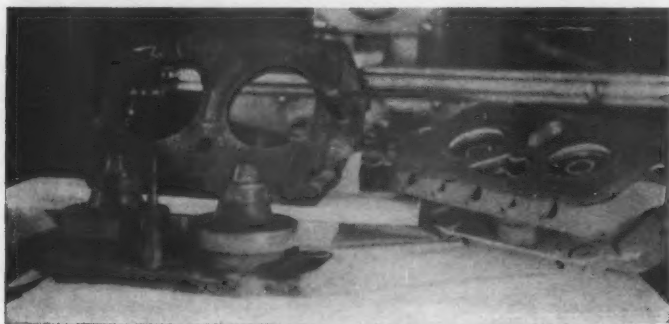
FIG. 13—HUB—OLD STYLE STRIPPER EQUIPMENT. NOTE HOLES IN PATTERN STOOL FOR STRIPPER GUIDE PINS.

FIG. 14—HUB—NEW TYPE DROP PLATE EQUIPMENT. PATTERNS PILOTED IN PLATE. MOLDING
MACHINE FRAME DESIGNED TO TAKE CARE OF SAND SHED.

meet the fluctuation of our work. The precision work brought
about by the demand for close limits on castings and high pro-
duction schedules means that we must have men trained expressly
for our work. Approximately 80 per cent of our men are S.M.I.
trained and have been with us for a considerable time.

CONCLUSION

61. The pattern shop has been able to cope with the demand

FIG. 15—FAN PULLEY. OPEN END DROP PLATE IN DRAW POSITION. RAILS USED
ON FRAME OF OPEN END DROP PLATES VARY IN HEIGHT TO CORRESPOND WITH THE SIZE OF
PATTERNS AND STOOLS. PATTERNS MADE OF HIGH CHROME IRON.
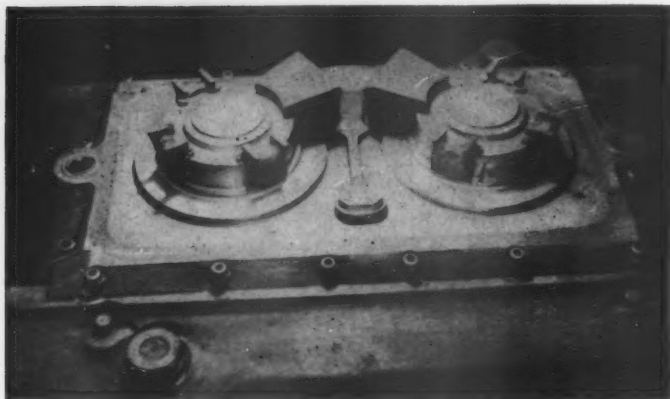
FIG. 16—DIFFERENTIAL TRUCK CASE. RAM-UP GATE CORES AND HEAVY CHILLS USED. PATTERN MADE OF BRASS. CHILL SEATS AND RAM-UP GATE CORE STOPS MADE OF STEEL. CORES NOT GROUND TO SIZE FOR SNUG FIT. A FLAT STEEL SPRING IS ATTACHED TO BOB AND GATE CORE STOP. BOTH OF THE JOBS WERE FORMERLY MOLDED ON ROLL-OVER TYPE MACHINES.

for a constantly improved product. Research developments of new ideas and new ways of doing things have made the possibilities of future accomplishments infinite. The amazing progress of the automobile during the past 20 years is just a small beginning compared with what is being planned for the future. Today's achievements are only a foundation for the higher standards of living of tomorrow.

DISCUSSION

Presiding: V. REID, City Pattern Works, Detroit, Mich.
Vice-Chairman: H. M. HARROLD, Caterpillar Tractor Co., Peoria, Ill.

MR. HYNAN: I wish to say that the paper as presented has not attempted to lay down any definite program of the efficient operation of a pattern shop. There is no question in my mind but what pattern making and molding methods have advanced with the present day methods of industry. I believe that a pattern shop can and should be used as a laboratory for foundry research as well as the everyday problems that are presented. You can make a pattern shop just another expense department or you can make it a real asset to your plant.

CHAIRMAN REID: In using brass for facing your core boxes, do you not find that it peels out at the joint of the box and tends to rip up your core and requires frequent dressing?

MR. HYNAN: That is quite a problem. You have to be very, very careful in hand-working the contour of your core down after you have applied the brass, but we have not had any trouble in the core tearing from the brass. Of course, we use a hard brass, a hard-tempered brass, and it will have to be worked down pretty close after the application of the brass to the contour of the box.

MEMBER: What is the advantage of brass over steel for such applications as core box facing?

MR. HYNAN: I do not know that there is any advantage of brass over steel. I imagine it is a little easier to work than steel. We do use a lot of steel in facing the ends and sides of the boxes. On the joint face of the box, for this particular application we use brass. I do not know of any reason why we could not use steel.

MEMBER: Do you use metalizing for facing your core boxes, particularly the application of the metal with the gun?

MR. HYNAN: I have had a demonstration in our shop of the application of both bronze and aluminum. I also used it on the application of plates, alumatizing plates for our new kilns that we have installed. It is just a mechanical bond and under the abuse it took in our plant, it did not seem to hold up. We applied it to a box and it did not hold up. We have also used chrome plated boxes, but the chrome plate peels.

MEMBER: I have been working with it for the last six months or so. It holds out pretty good on the face of a pattern, but at the joint it peels. I wonder if anyone here has had that experience.

MR. HYNAN: You have to be very careful in cleaning the particular part to which you aim to apply it. Just the oil from your finger, from rubbing over the part, will break it. It is purely a mechanical bond. It is not fused in with the metal at all.

MEMBER: In making three million castings, how many a day do you produce?

MR. HYNAN: We are operating two eight-hour shifts and producing around a hundred and twenty molds per hour, full run. We have run as high as a hundred and forty-six molds per hour.

MEMBER: We have had success in facing our core boxes with cyanided steel. Of course, we have a binding machine that takes care of the irregularity to the edges, but we are on a high productive basis on malleable iron and cast iron jobs. The higher the casting, the better the fit. Everything is drilled with dowels and jigs which cuts the cost down to a minimum in core boxes.

MR. HYNAN: Due to the fact that we blow all our cores, there is quite a problem presented in holding a tight joint face. We have spent a lot of money and a lot of time trying to learn how to properly vent a box. A cyanided steel face would be quite a problem to spot because every box we use we have to go over periodically and spot down because it will be blown out on the joint face and give scrap cores.

The Role of Heat Resisting Alloys in Short-Cycle Malleablizing Furnaces

BY H. H. HARRIS*, BOSTON, MASS.

1. Short-cycle malleablizing is merely the modern common-sense way of doing the job. Like most good things, it is simple and obvious. Formerly, castings were packed in a pot to keep the fire and combustion products away from them. Now we "pack" the fire and combustion products in an elongated pot called a tube. It is easier to encase the combustion than the charge of castings. In this, we have the choice of using a muffle large enough to house the entire load in which we can control atmospheres, or smaller "muffles" called "combustion tubes" to house the combustion source of heat.

2. There are a variety of types and designs of combustion tubes operated on different theories and sales arguments. Some are of sheet alloys, still others are a combination of both castings and sheets,—but all are of high nickel-chromium heat resistant alloys. All types are commercially satisfactory now but there will be many improvements in combustion tubes, in form, material and heat application. Patents have been allowed recently on a cast tube of spirally corrugated form which materially increases inner and outer surface, imparts a rotary motion to the combustion creating greater turbulence and liberating more heat. This rotary action creates a vortex suction which increases the capacity of the tube. One objective of this design is to eliminate the "hot spot" at which most tubes now fail.

ALLOY MECHANISM CARRIES LOAD

3. Nickel-chromium heat resistant alloys made the short-cycle furnace possible. The alloy mechanism carries the load, doing all the real work in the furnace. No malleablizing furnace is better than its alloy parts. The controlled atmosphere is made possible by alloy combustion tubes, the load is carried on alloy trays, and

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* President, General Alloys Company.

NOTE: Presented before Malleable Session 42nd Annual Convention, Cleveland, O., May 19, 1938.

it is pushed through the furnace on alloy roller rails. Failure of
any one of hundreds of stressed alloy parts could wreck the furnace
and seriously interrupt production. Rollers breaking or jumping
out, roller rails sagging and weaving, trays warping and cracking,
have been principal sources of troubles and are traceable to poor
alloy design and manufacture, for which there is small excuse in
the present state of the art.

MALLEABLE SHORT-CYCLE FURNACES INFLUENCED BY DESIGNS FOR OTHER INDUSTRIES

4. The continuous or pusher type of short-cycle furnace is
copied largely, as to form and alloy mechanism, from long proved
continuous and counterflow box-carburizing furnaces and similar
furnaces for annealing forgings, stampings, etc. The loads in mal-
leablizing furnaces are much heavier, and the "pushes" larger
and longer; thus the bearing and alloy loads are very much greater
than on carburizing and annealing furnaces—in some cases 300 per
cent greater. Because of the lower malleablizing temperatures, it
has been assumed erroneously that the life of malleablizing trays
will be much longer than trays used in carburizing furnaces. There
is no evidence to support this assumption, and a great deal against
it such as:

(1) The loads carried and the pushing stresses are much
     greater.

(2) Malleable castings grow as well as expand, and stress
     trays and containers.

(3) The very rapid cooling in many furnaces imposes added
     strain on the trays.

(4) Some combustion tubes have been set so close to the rails
     (remember that radiant heat is effective as the square
     of the distance) as to heat them unequally, exerting
     great strain. Trays, if too close to the tubes, may come
     to temperature much ahead of the charge so that the tray
     is expanding under a relatively cold charge with great
     interior temperature differentials.

(5) Carburizing atmosphere hardens rollers and bearings
     adding to their life—(some reducing atmospheres are
     detrimental to alloy).

(6) The sand falling into open bearings adds to friction and

shortens bearing life of rails and *increases compression stresses on trays.*

(7)   There has been a considerable amount of sagging and warping of rails in short-cycle malleablizing furnaces. This creates mis-alignment of trays, causing them to "snake" through the furnace and localizes pushing loads on corners making for early tray failure.

COMPRESSION LOADS UNDER TEMPERATURES

5.   The writer has designed trays, containers, and furnace rails for nearly 20 years and for the past 12 years has made a practice of gauging the compression loads of pusher furnaces with specially designed direct reading hydraulic compression gauges used under regular operating conditions. These gauge readings, made in more than 40 pusher furnaces equipped with rails and trays, show a great variation, depending upon rail and roller design, temperature at which rails operate (this is frequently quite higher than furnace chamber temperature due to proximity of burners), and upon foreign materials, sand, scale, etc., which bearings are asked to grind. Friction is increased and pushing pressures climb on account of sagging rails, settling of brickwork or foundations, etc., resulting in variations from level travel.

6.   Obviously, the greater the total frictional drag of the roller rails, the greater compression loads on the alloy trays. Some furnaces start with relatively low friction which builds up rapidly. As the compression loads on the alloy trays are many times greater than the beam loading, the extent of this compression determines how much alloy one has to buy for a *ram-rod, and how long it will last.*

ROLLER RAILS

7.   Consider roller rails. Obviously, there is no such thing as a truly straight or level rail in a furnace. Heating and cooling stresses, expansion, warpage, settling of brick work, and other factors prevent machine accuracy. It is equally certain that the bottoms of trays are not maintained to fine standards of straightness. Therefore, it is not expected that all rollers will bear. Frequently many do not and excessive loads are carried by the bearing rollers. Where there are three rails under a rigid tray, it is safe to say that the three rails will not bear evenly for ten per cent of their length, unless the tray is hinged.

8.   A new self-leveling roller rail was illustrated in the March
1938 number of *Metal Progress*. This rail is composed of small
"scale-beam" or "rocking-chair" supports carrying the rollers
and inter-related in such manner that the load is distributed and
all rollers brought into bearing. This self-leveling rail (Fig. 1)
weighs no more than non-load-conforming rails and, while a great
advantage in a continuous furnace, is almost a necessity to a well
designed batch type furnace.

9.   In the batch type furnace, the charge does not move under
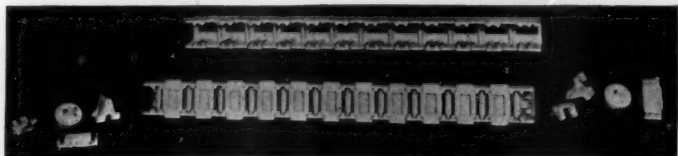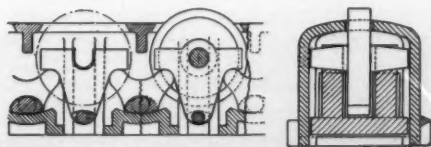
FIG. 1 — SELF LEVELING ROLLER RAIL — THIS RAIL HAS AN INTER-RELATED SCALE BEAM
ACTION AND DISTRIBUTED LOAD ON ALL ROLLERS, PREVENTING BEARING OVERLOADS. THIS
MECHANISM IS ENCLOSED IN INVERTED "U-MUFFLE" RAIL.

temperature but rests on the roller rails in one position for long
periods, which permits the trays or hearth-plates, if not excessively
rigid and heavy, to sag to meet the contour of the rails. The "cam-
pattern" thus developed on the trays subjects them to great strains
in moving them over other rollers of different conformations in
loading and unloading. Reduced to fundamentals, a roller rail
is simply a series of "grind-stone" bearings supporting rollers of
from 3 to 5-in. in diameter on trunnions of a diameter approxi-
mately one-fourth the roll diameter. They run dry. Most of the
alloys commonly used, resist oxidation, in part, through the forma-
tion of a tenacious surface oxide. This surface oxide is ground off
under the severe bearing service. The hardness of the bearings at
high temperature, tenacity of oxides, and rate of oxide replacement
determines the co-efficient of friction, which is also varied by the
secondary ingredients of the alloy, methods of casting the bearing,
and, of course, foreign materials between the bearing surfaces.

USE OF DIS-SIMILAR ALLOYS FOR ROLLERS AND BEARINGS

10. It is advantageous to use dis-similar alloys of different hardness and oxide characteristics for rollers and bearings. Much experience in our work with hot roller and ball bearings has been of great assistance in selecting bearing analyses, and we have proved beyond doubt that the most durable bearing analyses are not sufficiently ductile to make good rails. We are designing all new rails with the structural rail of the best known analysis (38 per cent Ni., 18 per cent Cr.), except where low first cost is primarily important. Bearings are of a harder alloy and rollers of a still different analysis. Such bearings are cast in "twin-bloc" pairs.

FIG. 2 — ROLLER RAILS AND ROLLERS (38 PER CENT NI. AND 18 PER CENT CR.) REMOVED FROM A CARBURIZING FURNACE WHILE IT WAS BEING DISMANTLED AND RE-ERECTED IN ANOTHER PLANT. BEING IN EXCELLENT CONDITION, THEY WERE RE-INSTALLED.

11. The idea of bearings being renewable is good sales talk but *separate* replaceable bearings are impossible of proper alignment under high temperature. Even machining the bearing blocks and their seats on the rails means little, for the small pieces come to temperature quicker, expand, and rock out of alignment. These "individual" loose bearings are an inadequate answer to a heretofore non-existent demand;—so far none of them has been in service nearly half as long as bearings cast integral with rails. Many of the latter are 10 to 18 years old and still in service. (Fig. 2).

12. So far, there has been no need for, and many disadvantages in, "replaceable" bearings in carburizing and annealing furnaces, but the higher loads of malleablizing furnaces call for harder bearings, demanding alloys too brittle for long rail castings. The answer in a replaceable bearing, illustrated in Fig. 3. The only

replaceable bearing that will stay in alignment under heat is one in which *both bearings supporting one roll are cast as one piece*, matched to their roller and set as a unit to a height gauge reading from the face of the roller to the base of the rail.

13.   For years we have all been, rather stupidly, guiding the trays through the furnace by letting them bear on the sides of the rollers. This, of course, crowds the rollers into side-thrust against the rail, and where the trunnions are tapered, makes them ride up the taper to increase friction. Guiding the trays direct on the rail is obviously a better mechanical job.
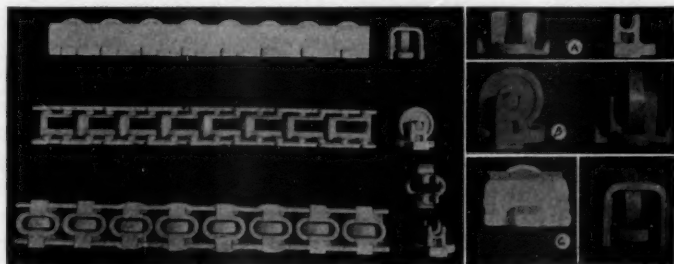
FIG. 8 — (LEFT) MUFFLE RAIL WITH RENEWABLE TWIN BLOCK. THESE BEARINGS ARE CAST IN ONE PIECE TO INSURE ALIGNMENT. MUFFLE INVERTED "U" SECTION PROTECTS BEARINGS FROM DIRT, INSURES UNIFORM HEATING OF CONTENTS AND HAS MAXIMUM WEIGHT —STRENGTH RATIO. — (RIGHT) REPLACEABLE BEARING BLOCK — (A) BEARING CAST IN CHILL, (B) ROLLERS CAST IN GRAPHITE BUSHINGS, JIG GROUND, (C) BEARING "A" INTERLOCKS INTO ROLL SECTION "C."

COVERED BEARINGS

14.   The accepted roller-rail designs all have open bearings into which dirt can fall and which do not prevent the rollers from climbing out, occasionally jambing the procession with considerable damage. All will agree that the bearings should be covered to keep out sand and grit, and the rollers retained. Simple common-sense stuff, — guide on rails, not rollers, — trap the rollers, — cover the bearings. These features are a few of the improvements incorporated in the new "muffle-rail" (Fig. 3). This rail is a simple inverted "U," having great strength, smooth exterior with no projections to heat irregularly. Inside this "U-muffle" section are the bearings and rollers where they can come up to temperature uniformly. Unlike the old rails, largely closed at the bottom to collect dirt, the muffle rail is open at the bottom. The ends of the roller trunnions contact the inner faces of the "U" section which centers the rollers, eliminating the objectionable taper and reducing friction.

15. Roller rails can bridge brick piers, be supported on alloy "stools," or rest entirely on brickwork, — a matter for the furnace engineer, as almost anything in structure or mechanics can be done in alloys.

ACCURACY OF BEARINGS AND TRUNNIONS

16. The greatest accuracy known to foundry practice is employed to make the bearings and trunnions of the new roller rails. Machining of any heat-resisting alloy casting is undesirable, for the fine grained "chill" surface is removed by machining or excessive grinding, as well as the surface oxide. (This oxide is formed at higher temperatures than the casting will ever again reach in service). Every sand casting is a "tri-laminar" structure, — two fine grained surfaces boxing a coarser center strata. Removing the fine grain chilled surface exposes an inferior wearing surface of coarser grain, throws previously equalized cooling strains out of balance, frequently results in warpage. Bearing pressure may actually grind oxides into the grain boundaries of coarse sections creating separation. A process has been developed using graphite cores and polished alloy chills which produces "smooth" bearing surfaces of great accuracy and much better wearing surface than heretofore obtainable.

NUMBERS OF RAILS TO SUPPORT LINE OF TRAYS

17. Assuming that we have now equipped a furnace with modern rails of latest design to minimize friction and reduce the tremendous compression loading of our alloy trays as far as possible, — shall we use two, three or more rails to support each line of trays? Logic permits one of two conclusions, *i. e.*, use two rails and "beef up" the tray to bridge the wide span necessary to a practical size of tray, or use *three or more* rails and use hinged trays which break up the span, weigh much less, cost less, and last longer, with an over-all saving of 40 to 70 per cent.

18. A three rail solid tray is a mechanical misconception as there is no such thing as three point continuous in line support in a mechanism under high temperature, and the tray will sag and flex to conform with the droop of the center rail, or cantilever to follow a low outer rail unless it is heavy and rigid enough to span the outer rails. Even then, depending upon load distribution and a high center rail, it can get some high adverse leverages which reduces its fatigue life.

HINGED TRAYS

19. Hundreds of hinged trays have been in service several years at temperatures up to 1800°F. There are several variations of the simple hinged and articulative principles. A hinged tray is based on the simple and obvious mechanical principle that a beam supported on a series of short centers can be very much lighter in weight than a beam with a wide span.

20. Because it is impossible to maintain uniform support of a number of in-line bearings in furnace, a series of short "beams" hinged together (Fig. 4) are substituted for one long one. In addition to the span advantage in the tray (effected by increasing the number of rails in the furnace and reducing the weight), it is found that as the spans are reduced, large castings in the charge have dimensions equal to or exceeding the rail span, which removes much of the load from the span of the tray and transfers it to direct compression over the rails. Because of this, it is possible to use trays of $\frac{1}{4}$-in. section or less on 7 or 8-in. centers. Fig. 5 is an illustration of a four-rail hinged tray without box sides. Hinged trays will save from 20 to 70 per cent of the weight of solid, grid, corrugated, or "bed-slat" trays, depending upon the number of rails employed. Several sets of trays will be purchased during the life of one set of rails.
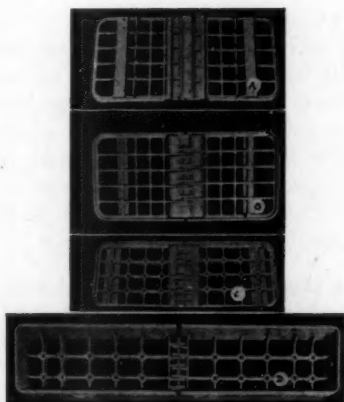
FIG. 4 — HINGED TRAYS, GRID TYPE — LATE DEVELOPMENT IN TRAYS FOR ANY SERVICE, CARBURIZING, NORMALIZING, ANNEALING, MALLEABLIZING, CAST IRON ANNEALING, ETC., (A) AND (B) CONTINUOUS, POSITIVE 3-POINT SUPPORT IRRESPECTIVE OF MINOR RAIL LEVEL CHANGES, LIGHTER WEIGHT FOR SHORT SPAN (C) MAXIMUM STRENGTH-WEIGHT EFFICIENCY BY USING MULTI-HINGED (NOT SHOWN) GRID. BEARING LOADS AND UNSUPPORTED SPAN STRESS REDUCED TO MINIMUM, (D) SPECIAL HIGH SIDES FOR NORMALIZING ANNEALING, ETC.
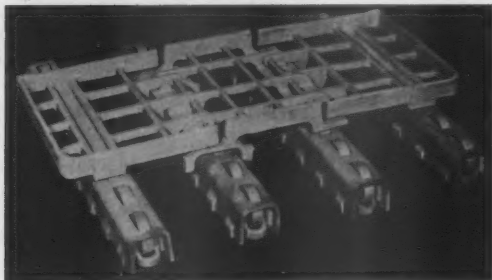
FIG. 5 — HINGED TRAY, FOUR RAIL — THIS TRAY IS ONE-HALF THE WEIGHT OF TWO-RAIL TRAY OF SAME LOAD CAPACITY, RAILS WEIGHING 11 LB. PER FT. PER RAIL AS AGAINST 20 LB. PER FT. PER RAIL OF TWO RAIL INSTALLATION. TRAY GUIDES ON RAIL, NOT ON ROLLER.

21. It is perfectly obvious that putting in extra rails which remain at temperature, and taking weight out of trays which are heated and cooled, is good business. Actually, the first cost of the extra rail, or rails, is usually less than the saving through weight reduction on the *first* set of trays, but this saving is multiplied four or five hundred per cent by the savings on future trays, not including the saving through eliminating the heating and handling of hundreds of tons of excess weight and the reduced investment and depreciation.

PROVISION FOR EXPANSION OF TRAYS

22. If trays have box sides, such sides should be flexible or have provision for expansion. This sounds obvious as we all know that malleablizing increases the size of castings and that the charge in the tray actually "grows" in addition to normal heat expansion, yet until this problem was tackled for a large implement manufacturer, nobody had bothered to provide for expansion. They just let the charge grow and distort the tray and, incidentally, warp the malleable castings. As a tray rigid enough to resist the expansion and growth of iron under heat cannot be built, we conclude that allowance must be made for expansion.

23. Presume that a box-tray is loaded with castings bearing against the sides when cold. The tray will distort slightly under the strain of each heat, becoming progressively wider. Each heat is loaded to the tray's new dimension, "chinked" with small castings, with this procedure being repeated until failure. Of course, if the trays are *never loaded full* solid inflexible sides will work, but at a considerable sacrifice of tonnage. A tray with expansible sides is illustrated in Fig. 6. The sides have approximately a one
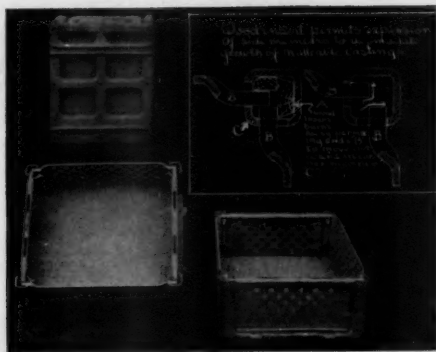
FIG. 6 — TRAY WHICH ALLOWS FOR GROWTH AND EXPANSION OF MALLEABLIZING CASTINGS WITH FLOATING SIDE DESIGN. TUBULAR CHASSIS WITH THIN BOTTOM LINER PROVIDES MAXIMUM STRENGTH AND RIGIDITY. THE SIDES ARE REPLACEABLE.

inch total side movement. Before the tray is loaded the sides are collapsed and a piece of wood dropped behind the prongs supporting the sides. When the wood burns out, the tray can expand before the expansion of the load. Patents are pending on expansion provision by means of shearing pins, elongating wire loops, and catches released by metallic expansion.

24. Another method of providing for expansion of the load is a flexible frame around the tray, in principle like a link bracelet, which is free to float and cannot be distorted into leverages against itself. Such design should provide maximum life. A third and fourth method are being developed and trays made to these new designs will be in service by the time this paper is read.

TUBULAR CHASSIS TRAY

25. On an entirely different line of argument, there has been developed an entirely new type of tubular chassis tray or "hot-buggy," equipped with small wheels, for use in a continuous short-cycle furnace. This job uses less alloy, in proportion to the loads carried, than any short-cycle furnace yet built. All moving parts come out of the furnace and the only alloy in the furnace, other than combustion tubes, is the light weight channel rail fully supported on brick-work.

26. A special replaceable bearing is employed in the design of these trays, making the wheels and bearings a simple, easily serviced unit (Fig. 7). In this type of tubular chassis tray, economy in alloy is achieved through eliminating the roller rail—using
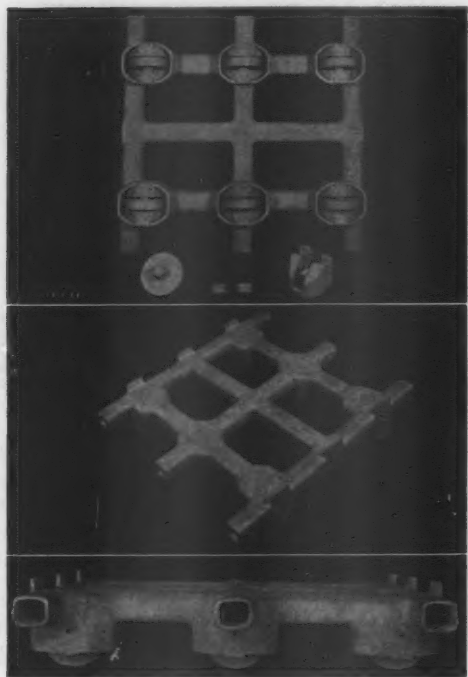
FIG. 7 — A "HOT-BUGGY" CAR AND PARTS, WHICH SUBSTITUTES FOR TRAYS. IS OF MAXIMUM STRENGTH TUBULAR CONSTRUCTION OF ONE-FOURTH INCH WALL SECTION, EQUIPPED WITH TWIN-BLOCK CAST-IN-PAIRS RENEWABLE BEARINGS, PROVIDING LIGHT OVER-ALL ALLOY WEIGHT. IT CARRIED ¼-IN. CAST BOTTOM.
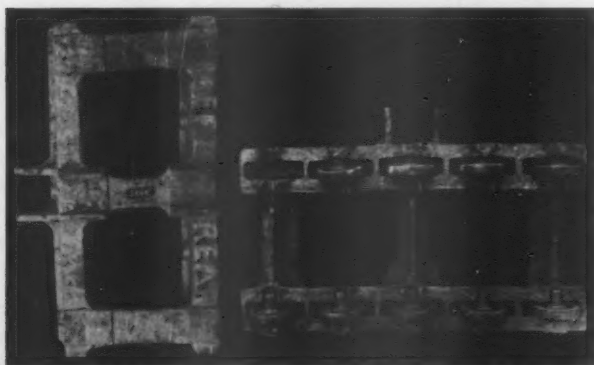
FIG. 8—ALLOY TRANSFER CAR (88 PER CENT NI.—18 PER CENT CR.) USED IN CARBURIZING FURNACES FOR 10 YEARS CONTINUOUSLY AND STILL IN GOOD CONDITION.

fewer rollers on wider centers as wheels on the "tray." Fig. 8 is an illustration of a transfer car of 38 Ni.—18 Cr. alloy, used in constant service for 10 years and still in good condition. This car operated at carburizing temperatures, transferring every carburizing box that passed through a return flow recuperative carburizing furnace.

Pushing Trays Through Furnace

27. So far we have been considering continuous pusher type furnaces. We believe that the use of alloy in compression as a ram-
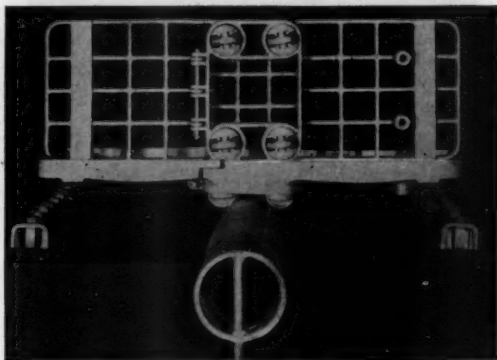
Fig. 9 — Smooth Screw Furnace Conveyor—The Rotation of Tube Moves Trays at any Desired Speed, With Hinged Trays Having Replaceable Rollers and Bearings. Furnace Is Self-emptying and Can Be Operated Continuously or Intermittently. One Tray Shown Tipped Edge-wise on Top of a Tray in Correct Position.

rod to push trays through a furnace is unnecessarily costly. Much of the weight of compression in trays is there solely to take the thrust. The stresses are intermittent and combine oddly with the load stresses to distort trays, shorten their life by "fatigue."

28. There are patents pending on four different methods of moving trays through a furnace without compression. One of these, illustrated in Fig. 9, is now installed in a furnace of a large gear manufacturer at Muncie, Ind. This conveyor can run at any desired speed, start, stop, back up, and can be operated intermittently or continuously. It is self-emptying and can be run at any percentage of production required, one tray or a full load. It is not necessary to push a dummy charge into the furnace to get a live charge out and reverse the process every time the furnace is shut down as is the case with all "pusher" furnaces. The simplest
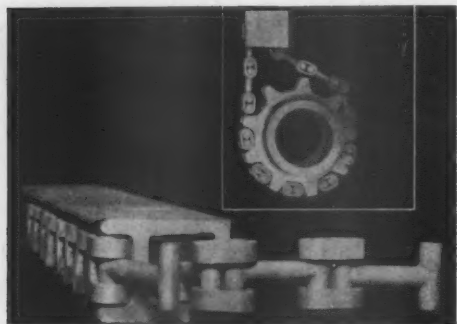
FIG. 10—AN "IN-A-RAIL" CHAIN WHICH RUNS ON TOP OF A HOLLOW RAIL, RETURNING IN
THE SAME RAIL INSIDE THE FURNACE, ELIMINATING HEAT LOSS AND COMPLICATED CONSTRUC-
TION OF UNDER-HEARTH RETURNS.
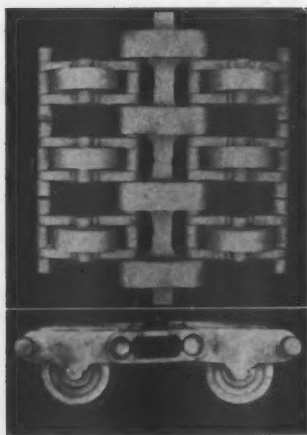
FIG. 11—ROLLER CHAIN CONTINUOUS HEARTH. THE LOAD CAPACITY OF THIS HEARTH IS
250 LB. PER SQ. FT. AT 1700°F.

tension conveyor mechanism is the "in-a-rail" chain of a type of construction which runs on top of a rail and returns inside the same rail eliminating heat losses and costly furnace hearth construction. The inner chain runs over a turn-tooth sprocket of only a 7¼-in. pitch line diameter with 4-in. shaft. This is also made with a roller in the rail suitable for malleable furnace loads. (See Fig. 10.)

29.   There is no reason why traveling-belt-type, moving, alloy hearths cannot be built with roller supports to eliminate trays entirely, carrying any desired load at any desired speed. A section of such roller chain hearth with load capacity of 250 lb. per sq. ft. at 1700°F. is illustrated in Fig. 11 and can be made in any length or width.

BATCH TYPE FURNACE CONSIDERATIONS

30.   Continuous, short-cycle furnaces of pusher, conveyor or smooth-screw type, are too large, represent too great investment, or are lacking in flexibility for many applications where "batch" type furnaces are indicated. Such batch-type furnaces are equipped with the same alloy components, roller-rails, trays and combustion tubes as the continuous furnaces. They represent entirely different alloy problems, however, for the charge is moved in and out of the furnace while relatively cold. This means that the trays sit in one place on the rollers for long periods of time under heat and under static unvarying load which tends to make the trays conform to the support, unless they are made much heavier than in a continuous furnace. The "self-leveling" roller rail, previously described, is one solution but it is only a partial solution as the mass expansion and "growth" of the charge "sideways" must be provided for in either tray or roller design, or both.

31.   The loads in batch short-cycle furnaces are much heavier per square ·foot of hearth area than in continuous furnaces, and the bearing loads are much greater. The fact that the rollers do not turn while under malleablizing temperature, necessitates making them heavier, and it is desirable that they do not come to rest under load repeatedly in the same position. Convenient roller rails are not desirable in batch furnaces as they provide neither load-conforming support or provision for expansion of charge and trays at right angles to the roller travel.

32.   The charge in batch furnaces frequently may cover 100

sq. ft. or more of hearth area. It will expand and grow in all dimensions from the center, frequently a matter of several inches. Failure to provide for this expansion has spread and rolled over roller rails — made some sorry messes.

Dry Skidding

33. The problem of "dry-skidding" (sliding of charge as it grows and expands) tons of charge on grids or trays, and taking the side movement of the trays, part of which is reversed on cooling, is interesting to the designer. The writer has approached this problem from four distinct mechanical conceptions and has achieved four separate satisfactory solutions as follows:

(1) The use of a "hot-buggy" tubular chassis "trays with wheels," in which the roller trunnions have sufficient side movement to provide for expansion of and between trays. The tubular chassis is supplemented by light grids to provide grating support for small castings as desired. The hot-buggies come out of the furnace "cold," so there is no intermittent heating and cooling of alloy. The rails in the furnace are simple channels. The charge being "on wheels" can be moved any distance desired outside the furnace. Bearings are renewable.

(2) The use of a series of hinged trays linked fore and aft to form a semi-flexible hearth unit, supported on self-leveling rails. Provision for side expansion is incorporated in the hinge design.

(3) Triangular trays, supported on a roller train (Fig. 12). The triangular tray is the only unit design which is inherently in register. This is because it has true three point support, and like a three legged stool, cannot shift its bearing. Two triangles make a rectangle, or a rectangular hearth of any size can be made of a series of triangular trays or sections. Each triangular tray can have three rollers, or, as in the design shown, the number of rollers can be reduced by carrying the trays on the "roller-train" shown. This design is full floating, will carry heaviest loads and give longest alloy life.

(4) Sectional one piece trays of solid construction, or hinged as desired, can be carried on hollow alloy balls instead of rollers. The balls are trapped in such a manner that
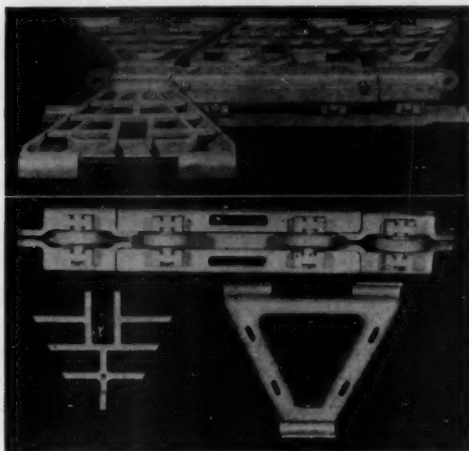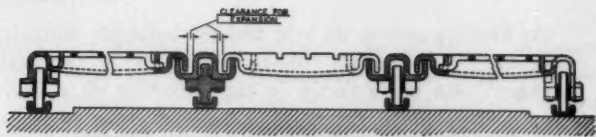
FIG. 12—TRIANGULAR SEGMENTAL HEARTH (CAR TYPE). THIS HEARTH IS COMPOSED OF
TRIANGULAR TRAYS WITH PROVISION FOR EXPANSION OF CHARGE. THREE POINT SUPPORT
INSURES UNIFORM BEARING AND MAXIMUM LOAD CAPACITY.
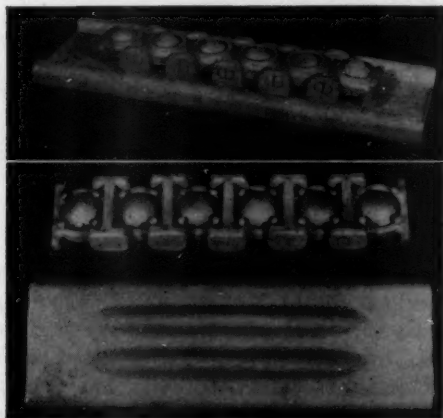
FIG. 13—BALL-IN-CHAIN TYPE OF CONSTRUCTION WHICH CAN BE PUSHED OR PULLED TO
MOVE HEAVY LOADS AND PERMIT MOVEMENT OR EXPANSION IN ANY DIRECTION. WEIGHT OF
LOAD IS CARRIED ON RUNNERS LEAVING BALLS FREE. THE RAIL SECTION, BUILT TO SPAN
PIERS, HAS RENEWABLE HARD ALLOY BEARING SURFACE.

the housing carries no load and the balls can move freely
in the "ball-in-chain" design (Fig. 13). The "ball-in-
chain" can move freely in any direction on a flat rail,
or bearing plate, the movement being actuated by push-
ing or pulling the trays, the load, or the "ball-in-chain."

34. All of the above methods are covered by patents issued
or pending.

CAUSES OF FAILURE IN ALLOY TRAYS

35. Failures in alloy trays invariably occur in commercially
sound units at the following points: edges, intersections and in
varying metal sections. Sections in most common use and most
designs employing them have high temperature differentials in
close proximity which impose great strains in heating and cooling,
set up the fatigue strains which cause ultimate cracking and fail-
ure. A ½-in. x 1½-in. rectangular cross section bar, for example,
or any similarly proportioned form, is a very poor form to resist
heat fatigue. A different form, having the same cross sectional
area will withstand more than twice as many temperature changes
without cracking.

36. Uniform surface to mass ratio is essential to uniform
heating and cooling. It is equally essential to uniform shrinkage
in the mold, which is determined by cooling area of mold per cu. in.
of metal. Suffice to say here that the old checkered grid, corrugated
plate, and other accepted types of trays is definitely on the way out.
The new trays call for more exacting foundry practice, more costly
pattern equipment, and scrupulous inspection, for quality is vital,
but the result will be drastic weight reductions with greatly in-
creased net-to-gross ratio of charge to tray weight, making sharp
improvement in overall furnace efficiences.

THE FUTURE

37. It is the writer's belief that no alloy tray yet used in
malleablizing will equal the life of trays used in continuous anneal-
ing furnaces. Further, that roller-rails so far installed in malle-
ablizing furnaces cannot be expected to approximate the life of
rails in carburizing furnaces. The "easier alloy operating condi-
tions" of malleablizing furnaces is a myth based on sales
exuberance.

38. There is every reason, however, why a much better job

on alloys in malleablizing furnaces can be done. This can come about through the frank recognition that of greatest importance in the furnace is the alloy mechanism, that the best is none-too-good, and by the application of the best alloy engineering and foundry experience available.

39. There is so much ahead in the heat-treatment of iron, such a glorious future which the metallurgist is unfolding to the progressive iron foundryman, and which able furnace engineers are making possible, that it is a privilege and a pleasure to participate in this development. Short-cycle malleablizing is not only a self-evident reality but, to most shops, a vital necessity. The modern furnace has made this possible, the "works" of which are alloy castings.

DISCUSSION

Presiding: L. N. SHANNON, Stockham Pipe Fittings Co:, Birmingham, Alabama.

J. D. CORFIELD[1]: I have been interested in Mr. Harris' paper, particularly because he stresses the efficient utilization of the industrial jewelry we call "heat resisting alloy." I am sure that the various ways in which he has achieved that purpose have been very enlightening to us all. But I think that in a paper of this type, devoted particularly to short cycle continuous malleable furnaces, it would be a mistake to pass by without some mention of an entirely different type of conveyor in which the work is carried directly or in trays on a series of live rolls. In this type of furnace, all the cooperating parts of the conveyor are entirely outside the furnace; therefore, many of the complications Mr. Harris stressed do not exist.

The roller hearth type furnace also makes possible the most efficient use of heat resisting alloys, the rollers being in rotation, and the alloy can be assumed to have somewhat more load carrying capacity per unit of cross sectional area than the stationary hearth furnace parts such as the rails which operate in a stationary hearth. It also gives the furnace designer somewhat more latitude in designing the furnace by elimination of the rail supporting members.

Since many short cycle malleable furnaces of the roller hearth type have been built and are in successful operation, I felt that this type of conveyor should be brought to the attention of this meeting.

MR. HARRIS: Concerning Mr. Corfield's remarks with respect to the roller type — I take it what he means is a tubular roller extending through the furnace and having cold bearings on the outside of the furnace. That type is thoroughly successful and workable. There is

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[1] Michigan Steel Casting Company, Detroit, Mich.

nothing the matter with it. It has been very widely used in the sheet annealing business, carrying relatively light loads. Any loads that have been carried, to my knowledge, are materially less than the loads now contemplated with the long push malleable jobs. Further, I am quite sure from my knowledge of this that Mr. Corfield has some other solution to arrive at, that the total amount of alloy used on that type of hearth is conservatively from 50 to 200 per cent more than the amount of alloy used on any of the straight roller furnaces. I do not think there is anything the matter with the job. You are going outside the furnace on the bearing support and are buying alloy to form a spanner beam, which is unnecessary excepting as it applies to that design, just as in the drive push job, or in the roller rail job pushed from outside the furnace, you are using the roller as a ramrod. It has, however, the advantage of economy on trays because it does not compress the tray. But I should say your overall alloy cost is materially higher. I would be willing to state conservatively that you cannot build that type furnace without from 50 to 150 per cent more alloy to do the same job than some of the types described. As far as mechanics is concerned, it is a perfectly good job.

MR. CORFIELD: As to the amount of alloy required or the overall cost of the roller conveyor, disregarding the mechanics of either I disagree. In some studies with which I am familiar, there is practically little difference. In one case, it costs considerably less in favor of the roller hearth furnace, since trays are not needed. With the average pusher job you have both boxes and trays to consider.

MR. HARRIS: The boxes have some weight. Let us assume you cut the box weight 50 per cent. The way that is done is by means of sides on the trays. You call them box trays but the sides are relatively light. You can run your diameter up and your wall section down. That is a foundry limitation primarily. But in that type of design, to fit a furnace 100 feet long, in which the entire hearth is made of 4 to 6 inch tubes, there will be a very material amount of apparatus as compared to three, four or five roller rails going through at a weight of 10 to 15 pounds per foot per rail. Mechanically, I would say the job is good, but I feel quite sure we can cut the weight 40 per cent with any one of the three sets I have mentioned.

The Effect of Different Electrolytes on the A.F.A. Clay Determination

By Joseph F. Haseman*, Ithaca, N. Y.

1. The work described in this paper was undertaken with the object of determining what effect electrolytes, other than sodium hydroxide, might have in making A.F.A. clay determinations, and furthermore what results different amounts of them might give.

2. The American Foundrymen's Association Committee on Standard Tests originally recommended the use of a one per cent sodium hydroxide solution to disperse the clay in naturally-bonded sands when making the standard fineness test, but later this was changed by recommendation of the Committee to a three per cent solution, as this appeared to give more uniform results. The use of distilled water also was recommended, and it has been pointed out that this should be as nearly neutral as possible.

Previous Investigation

3. Other investigators have tried different electrolytes, but the series was not as extensive as that discussed in the present paper. Galliher[1] ran a series of comparative tests using ammonium hydroxide, sodium oxalate, and sodium citrate. He terms his method "sedimentation by continuous volume," and obtained his data by measuring the cumulative per cent of the total volume of material that settles throughout a given period of time. From the results obtained, he concluded that sodium oxalate and sodium citrate were the most efficient dispersing agents of those tried.

4. It should be pointed out that he found these two electrolytes best adapted to the type of samples with which he experimented, and did not feel justified in concluding that they would

be the best in all cases. Furthermore, his samples all showed the equivalent of about 80 per cent A.F.A. clay. They were therefore to be regarded as clays rather than clay-bonded sands.

5. Krumbein[2], as a result of experiments with fine-grained sediments and calcereous clays, came to the conclusion that sodium oxalate was the best all round electrolyte to use. He found that good results were obtained throughout a wide range of concentration, but that this range of effectiveness varied considerably for different samples. The optimum concentration of the sodium oxalate for his particular samples was N/100, or that which would be obtained by the use of a 1.34 per cent solution according to the A.F.A. procedure. Olmstead, Alexander, and Middleton[3] have recommended N/200 sodium oxalate as best for soils.

6. Krumbein also obtained fair results with sodium carbonate. Here too, the optimum concentration of the electrolyte varied for different samples. He used concentrations ranging from 1.34 to 5.36 per cent (according to the A.F.A. procedure) with good results for certain materials. It appears that his samples ran high in clay content, as the one he cites as an example contained a little over 40 per cent A.F.A. clay.

7. The United States Bureau of Public Roads[4, 5], after extensive tests with various deflocculating agents, decided that sodium silicate is the most efficient deflocculator. The samples they test are mostly clays, but their decision is interesting in view of the results obtained by the writer.

CLAYS USED IN PRESENT INVESTIGATION

8. The materials used for the present investigation were a series of 20 clay-bonded sands whose A.F.A. clay content ranged from 6 to 39 per cent. The companies who kindly supplied them were asked to include samples which had given some trouble in the A.F.A. fineness test. By choosing such samples, it was hoped that any irregularities in behavior would be emphasized.

2 Krumbein, W. C., *"Dispersion of Fine-Grained Sediments,"* JOUR. OF SEDIMENTARY PETROLOGY, vol. 3, no. 3, pp. 121-135, December 1933.

3 Olmstead, L. B., Alexander, L. T., and Middleton, H. E., *"A Pipette Method of Mechanical Analysis of Soils Based on Improved Dispersion Procedure,"* U. S. Dept. Agric., Tech. Bull. 170, 1930.

4 Wintermyer, A. M., Willis, E. A., Thoreen, R. C., *"Procedure for Testing Soils for the Determination of the Subgrade Soil Constant,"* PUBLIC ROADS, vol. 12, p. 65, 1931.

5 Theoreen, R. C., *"Comments on the Hydrometer Method of Mechanical Analysis,"* PUBLIC ROADS, vol. 14, p. 93, 1933.

9. The clay determination was made in accordance with the method described in the AMERICAN FOUNDRYMEN'S ASSOCIATION publication* on the testing and grading of foundry sands.

ELECTROLYTES USED

10. The electrolytes tried were sodium hydroxide, sodium carbonate, sodium oxalate, sodium citrate, sodium pyrophosphate, and sodium silicate. Solutions containing 1, 3, and 5 per cent of the electrolyte were tried in each case. In addition, it might be mentioned that a sample was run with liquid soap solution and another with ammonium hydroxide with good results.

WATER

11. Distilled water, having a uniform pH value of about 6, was used, and the temperature was maintained at 25° $\pm$ 3° C. A duplicate sample of the 3 per cent concentration was run for each electrolyte tried, until it was determined that duplicate results checked closely. If duplicate tests checked within 0.4 per cent of the total sample, it was regarded as sufficiently close, but in most cases the checks were within 0.2 per cent.

RESULTS OBTAINED

12. The results obtained with the different electrolytes, and different concentrations of the same, are shown on the graphs, Figs. 1 to 10. Table 1 shows the average number of siphonings

* "Testing and Grading Foundry Sands, Standards and Tentative Standards," A.F.A. (1931), pp. 86-92.
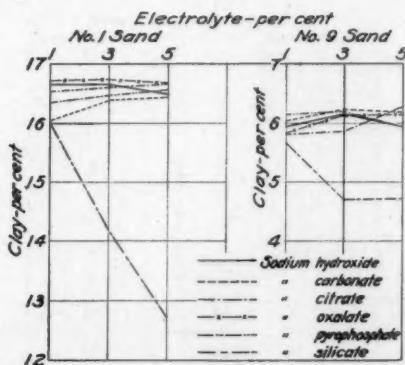
FIG. 1—TEST RESULTS OF VARYING PERCENTAGES OF DIFFERENT ELECTROLYTES WHEN TESTING No. 1 AND No. 9 SANDS.

WHEN USING VARIOUS ELECTROLYTES

Table 1
AVERAGE NUMBER OF SIPHONINGS NECESSARY TO REMOVE CLAY

Sand No.	Sodium-Hydroxide	Sodium-Carbonate	Sodium-Pyrophos-phate	Sodium-Oxalate	Sodium-Citrate	Sodium-Silicate	Per Cent A.F.A. Clay
1	6	6	6	6	7	8	16.70
2	7	8	7	8	8	9	11.90
3	10	9	11	9	9	11	13.35
4	6	6	6	6	6	12	21.30
5	9	11	12	11	10	14	28.30
6	10	10	10	10	15	14	14.50
7	12	13	14	13	14	14	17.50
8	7	7	7	6	6	12	22.38
9	6	7	6	6	6	9	6.28
10	11	11	11	10	11	15	29.00
11	10	10	10	10	10	15	26.30
12	7	7	8	8	7	8	10.30
13	12	12	13	13	15	16	22.30
14	16	14	14	14	12	17	18.60
15	14	14	14	14	14	15	30.70
16	17	16	16	17	17	20	38.60
17	10	10	11	11	11	13	16.60
18	7	7	9	7	7	10	21.24
19	12	13	12	11	13	15	24.74
20	6	6	6	6	6	13	16.30
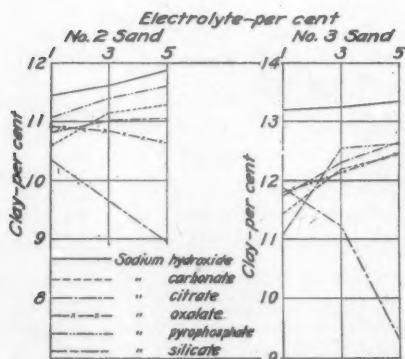
FIG. 2—TEST RESULTS OF VARYING PERCENTAGES OF DIFFERENT ELECTROLYTES WHEN TESTING NO. 2 AND NO. 3 SANDS.
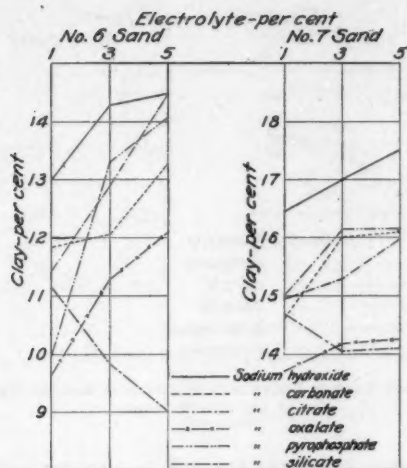
FIG. 3—TEST RESULTS OF VARYING PERCENTAGES OF DIFFERENT ELECTROLYTES WHEN TESTING
No. 6 AND No. 7 SANDS.
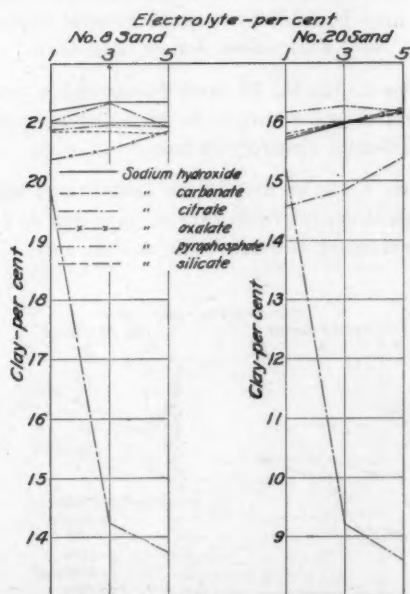
FIG. 4—TEST RESULTS OF VARYING PERCENTAGES OF DIFFERENT ELECTROLYTES WHEN TESTING
No. 8 AND No. 20 SANDS.
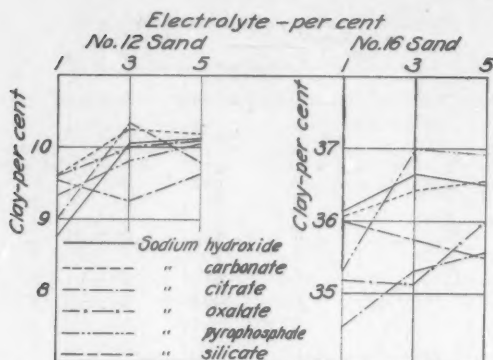
FIG. 5—TEST RESULTS OF VARYING PERCENTAGES OF DIFFERENT ELECTROLYTES WHEN TESTING No. 12 AND No. 16 SANDS.

that were necessary to remove the A.F.A. clay for each electrolyte used and also the maximum clay content determined for each sand.

13.   The numbers at the head of each graph refer to the sand sample, and it may be said, by way of general explanation, that they came from New York, New Jersey, and Ohio.

14.   Samples 5, 10, 11, 15, and 16 showed a relatively high clay content, and therefore might be expected to show the variations with the different electrolytes best.

15.   Samples 7, 13, 14, and 16 are fine-grained sands. In most cases, they required considerable care in handling as a great many washings were necessary to remove the A.F.A. clay.
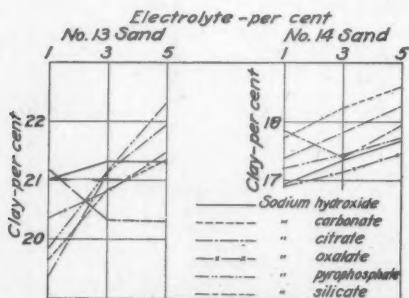
FIG. 6—TEST RESULTS OF VARYING PERCENTAGES OF DIFFERENT ELECTROLYTES WHEN TESTING No. 13 AND No. 14 SANDS.
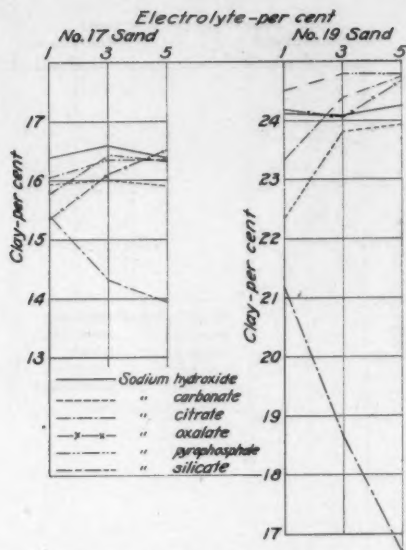
FIG. 7—TEST RESULTS OF VARYING PERCENTAGES OF DIFFERENT ELECTROLYTES WHEN TESTING
NO. 17 AND NO. 19 SANDS.
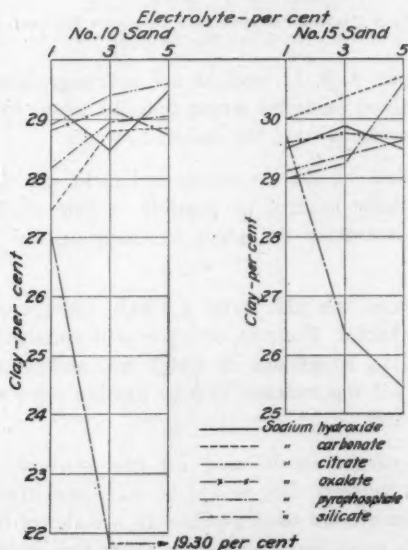
FIG. 8—TEST RESULTS OF VARYING PERCENTAGES OF DIFFERENT ELECTROLYTES WHEN TESTING
NO. 10 AND NO. 15 SANDS.
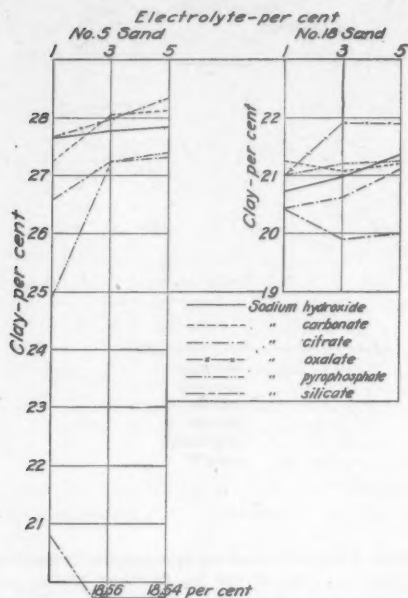
Fig. 9—Test Results of Varying Percentages of Different Electrolytes When Testing No. 5 and No. 18 Sands.

16. Samples 4, 8, 11, and 18 are coarse-grained sands, some of which contained particles larger than 10 mesh, which were first removed. These sands were the easiest to wash.

17. In view of the recommendation to use distilled water which is as nearly neutral as possible, a few experiments were also made to determine the effect of changing the water to acid or alkaline.

18. To raise the pH value a small amount of ammonium hydroxide was added. Four cc. of 5 per cent ammonium hydroxide solution added to 10 gallons of water was sufficient to raise the pH to 8. The pH was reduced to 5 by passing some carbon dioxide through the water.

19. Two of the sands used for this purpose were high in clay, as it was thought they would be more sensitive. The results of these experiments on sand number 16 are shown in Fig. 11. Of the other sands tried, none seemed to show any marked difference when treated in this way. Thus it would seem that the use of
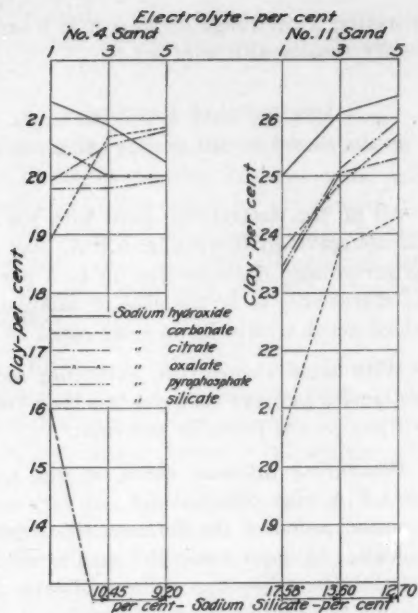
Fig. 10—Test Results of Varying Percentages of Different Electrolytes When Testing No. 4 and No. 11 Sands.
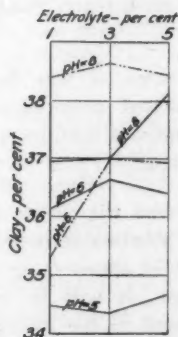
Fig. 11—The Effect of Distilled Water pH on No. 16 Sand When Using Varying Percentages of Different Electrolytes.

distilled water having a pH range of from 5 to 8 can be expected to give satisfactory results with most sands.

DISCUSSION OF RESULTS

20.   The results shown by the graphs bring out several interesting facts.

(1)   All of the electrolytes used, with the exception of sodium silicate, gave an increasing A.F.A. clay content with increasing percentage of electrolyte up to 5 per cent. There were some exceptions, as in the case of samples 4, 9, 10, 12 and 15, which acted erratically in some cases.

(2)   With most electrolytes, excepting sodium silicate, the change from 3 to 5 per cent was less than from 1 to 3 per cent, and 3 per cent is probably sufficient.

(3)   Comparing different sands, it was noted that the amount of A.F.A. clay obtained did not vary much with the optimum concentration of the different electrolytes, excepting sodium silicate. In most cases the results obtained by the several electrolytes, excepting sodium silicate, did not vary by more than 1 per cent of the total sample.

(4)   In 8 sands out of the 20 tested, sodium hydroxide gave the best results, and in 6 of the remaining cases its inferiority of position was almost negligible. In the other cases, slightly higher A.F.A. clay content was obtained with sodium pyrophosphate, sodium carbonate, or sodium oxalate. Sodium citrate gave consistently good results with every sand except number 20.

(5)   In three cases, it was found that one per cent sodium carbonate caused flocculation of the clay. This is noticeable in the graphs of samples 4 and 10, as indicated by their lower clay content.

(6)   In a few cases, the curves showed a lower clay content with 3 per cent solution of the electrolyte than with 1 or 5 per cent. Number 10 seems to be an outstanding example of this. In this case, it will be noticed that the sodium hydroxide curve is bent in the wrong direction. This seemed so exceptional that checks were run, but the results were practically the same.

(7) Sodium silicate behaved in a most unexpected way, considering the results obtained by its use in other lines of work.

21. In all of the samples tested, with four rather doubtful exceptions, the A.F.A. clay content decreased, with an increase in concentration of the electrolyte. In most cases, this decrease was extremely rapid as is illustrated by the graphs of samples 4, 8, 11 and 20. Difficulty was also encountered in washing out the clay (see Table 1), and with any given sand, more siphonings usually were required with sodium silicate than with any of the other electrolytes.

22. Two points may be noted, although we cannot say yet whether they are of special significance. One is that iron-stained sands gave trouble with sodium silicate. The second is that fine-grained sands gave better results with sodium silicate than the coarse-grained ones.

23. Since in all cases one per cent sodium silicate gave better results than 3 or 5 per cent, ½ per cent was tried with one sand, but it was found to give poorer recovery.

24. The peculiar behavior of sodium silicate is interesting in view of the fact that the United States Bureau of Public Roads uses it for dispersing clays. It appears to give better results with such material than it does with clay-bonded sands.

25. It is evident, of course, that in nature we may find every gradation between a clay free of sand or silt, and a clean sand. At what clay content in this gradation sodium silicate ceases to be effective, remains to be determined.

Conclusion

26. In conclusion it would seem safe to say that for the A.F.A. clay determination, sodium hydroxide probably is the best all round electrolyte, and the 3 per cent concentration now recommended by the Committee on Standard Tests is quite satisfactory.

27. It is not felt that the brief results communicated herewith finish the problem. There is a strong possibility that the texture of the A.F.A. clay, its mineral composition, base exchange capacity, and even other factors may affect the results, and

account for some of the divergences of behavior exhibited. The writer hopes to be able to attack these problems later.

ACKNOWLEDGMENT

28. The writer wishes to express his thanks to the Ayers Mineral Company, George F. Pettinos, Whitehead Brothers Company, and the National Bureau of Standards for supplying him with samples. Acknowledgment is also due to Dr. H. Ries, for guidance in the work, and for the many helpful criticisms and suggestions he offered.

DISCUSSION

In the absence of the author, this paper was presented by Prof. H. Ries, technical director, A. F. A. Foundry Sand Research Committee, Head, Geology Department, Cornell, University, Ithaca, N. Y.

Presiding, HAROLD W. JOHNSON, Greenlee Foundry Co., Chicago, Ill.

R. E. GRIM:[1] *(written discussion)*. The author has done an important service in pointing out the variation of analytical results for A. F. A. clay determinations when different preliminary treatments with electrolytes are used. In the Illinois State Geological Survey laboratory many mechanical analyses have been made of clays and clayey sands which emphasize that the factors of constitution are important in evaluating mechanical analyses of molding sands.

Let two molding sands be postulated, each containing 20 per cent A. F. A. clay. The clay in one sample contains 50 per cent quartz in particles 20 to 5 microns in diameter and 50 per cent clay minerals in particles less than 5 microns in diameter. The other clay contains 10 per cent quartz in particles 20 to 5 microns in diameter and 90 per cent clay minerals of the same kind and size as the first sample. All other factors being equal, the latter sample will require more electrolyte for suspension because the clay minerals have higher requirements for electrolytes than quartz.

Let two molding sands be postulated, each containing the same amount of A. F. A. clay and the same percentage of the same clay mineral in the clay grade. In one sand Na^+ is the dominant exchangeable ion, in the other sand, cations other than Na^+ are the dominant exchangeable ions present. Haseman has shown that an excess of electrolyte reduces suspending qualities. Because the first sample already carries Na^+ only a small amount need be added to produce a good suspension. In the second sample because little Na^+ is present originally, more would be required for a suspension.

[1] Illinois State Geological Survey, Urbana, Ill.

It is known that all clays do not contain the same clay minerals and that different clay minerals possess different adsorptive capacities and break down to extremely fine particle size with different degrees of ease. Thus, a sand with a clay grade composed largely of montmorillonite which is a clay mineral with high base exchange capacity and great readiness to break down to a very fine particle size, should show different electrolyte requirements and different ease of removal of the clay grade than a sand with a clay grade composed of kaolinite. Kaolinite is a clay mineral with low adsorptive capacity and low ability to break down in particle size when worked with water.

H. L. DAASCH:[2] *(Written discussion)*. The unexpected behavior of sodium silicate as reported by Mr. Haseman is most interesting and it suggests the need of more fundamental information on the physics and chemistry of our methods of sand and clay testing.

We have been carrying on certain investigations of foundry sands at the Iowa State College and in the course of the work have had occasion to study fineness and clay content of our sands. The Bouyoucos Hydrometer method seemed appropriate. We attempted to coordinate the Standard A. F. A. Tests with the hydrometer test. The thought behind the effort was that hydrometer data might supplement sieve analysis and thus give a size distribution from 1 micron on up.

The total A. F. A. clay did not always check the 20 micron and less material determined by the hydrometer. Since the latter process involved sodium metasilicate as the electrolyte, it seemed possible that the difference in the electrolytes was the cause of failure to check.

The A. F. A. method with the substitution of sodium metasilicate for sodium hydroxide was therefore used. We thus had tests such as those reported upon by Mr. Haseman. Comparative results on three sands are shown in Table 2. It should be noted that the three sands here reported are ones which yielded dissimilar A. F. A. and hydrometer clay. Sieve analyses are also shown for each test.

Table 2

Sand	Electrolyte	6	12	20	30	40	50	70	100	140	200	270	Pan	Clay
A	Sodium Hydroxide	0	0	0.04	0.08	0.33	0.42	1.47	4.88	7.32	10.06	16.96	50.35	7.79
A	Sodium Meta-Silicate	0	0	0.0	0.06	0.33	1.58	2.76	4.97	7.27	11.46	10.35	47.20	13.02
B	Sodium Hydroxide	0	0	0.0	1.48	4.28	8.02	11.96	11.98	4.72	3.12	3.74	32.08	18.44
B	Sodium Meta-Silicate	0	0	0.20	1.28	4.97	9.20	12.77	12.88	4.58	5.57	4.59	32.38	14.00
C	Sodium Hydroxide	0	0	2.18	2.00	3.08	4.78	8.28	18.58	21.58	16.98	7.28	10.88	3.82
C	Sodium Meta-Silicate	0	0.68	2.08	1.88	3.08	4.46	8.16	17.26	20.66	16.48	7.18	11.18	5.96

Sands A and C show much higher clay when sodium metasilicate is used as the electrolyte. This is apparently the opposite from trends shown by Mr. Haseman.

[2] Professor, Mechanical Engineering, Iowa State College, Ames, Iowa.

It will be noticed that when sodium metasilicate gives higher clay content, the greatest discrepancies in sieve analyses are in the finest sieves and pan. The use of metasilicate yielded low percentage of fine sand and pan material with correspondingly higher clay. Sample B with higher clay for sodium hydroxide as the electrolyte shows major difference in 50 and 70 mesh sieve size particles.

It is conceivable that large grains of sand are made larger by adhering clay; that pan (and even finest sieve) material might result from clay particles adhering one to the other; and that the effectiveness of an electrolyte in removing clay layers from a sand grain may be much different than its effectiveness in separating clay from clay.

It is my understanding that those interested in road materials are concerned with size data within the A. F. A. clay and pan ranges. The materials are high in clay. Does their recommendation for the use of sodium metasilicate result from a superiority in separation of the large quantities of fine constituents?

The three tests that I have quoted are entirely inadequate but the above line of reasoning might be borne out by further tests along this avenue of approach.

Whereas Mr. Haseman suggests as many as sixteen syphonings for the clay determination we have found that 25 and even 30 syphonings are necessary with some sands for complete clarity of the water. While this caution is not advocated for commercial tests, it is an item which should not be overlooked in research work.

Finally, it should be remembered that the data presented here have applied to sands which we have found "difficult" to handle. They are the exception rather than the rule. The reason for troubles with these rarer cases is undoubtedly dependent, as Mr. Haseman suggests, on mineral composition and characteristics and we hope Mr. Haseman continues to work along the lines he has indicated.

DR. RIES: I am very much interested in Professor Daasch's results and wish to ask him whether he used three sands only, or whether he simply quoted three here as examples.

MR. DAASCH: Those three sands were the only ones we checked with the sodium hydroxide and the sodium metasilicate because they were the ones that gave the greatest diversions in the hydrometer tests. On the others, we got very good relationships in putting together the data. The A.F.A. clay with sodium metasilicate here checked rather reasonably the clay with the hydrometer using the metasilicate as well. The two systems, using the same electrolyte, seemed to check.

DR. RIES: There are a number of factors which we still have to determine. There is much work still to be done and if more than one investigator is working, it would be highly desirable for the several investigators to try samples of the same sands. It is possible that because of conditions sands in the west might vary from those in the east, although it is not certain; they might be different in one climatic condition from what they would be in another one.

Report of Steel Division Committee on Methods of Producing Steel for Castings

To Members of the Steel Division, American Foundrymen's Association:

A study by your committee of the information available on steel making practices and processes during the year 1937 discloses a few developments of a remarkable nature.

In the open-hearth field, aside from the literature evidencing the continued interest in insulation and slag control, a very good paper[1] on the operation of basic open-hearth furnaces in a steel foundry was presented by John W. Porter to an A.F.A. meeting last October.

A development, which possibly may be credited to 1937, is the utilization of the cupola furnace for the pre-melting of open-hearth scrap. It is understood that at least one producer is now installing this type of equipment. One difficulty of such a practice is the cost and the problem of holding down the sulphur absorption from the cupola fuel.

The most striking development in the electric arc furnace is the use, by the Ludlum Steel Company, of rotating hollow, graphite electrodes inclined slightly from the horizontal toward the arc to permit briquets of chrome ore and coke placed in the electrodes to flow freely into the bath from the arc. This furnace, being of the indirect arc type, is used in the production of stainless steels.

The subject of cupola hot-metal duplexing for electric steel foundries has been carried further by some companies but progress reports have not been available to your Committee.

The revival of interest in the converter process seems to be expanding. It is contended that, with the present day knowledge of desulphurizing practice and the physical chemistry of steel making, it is possible to produce steel by this means capable of meeting exacting chemical and physical specifications. Modern acid

[1] Porter, J. W., "Basic Open Hearth Practice," Iron Age, vol. 140, no. 22, Nov. 25, 1937, pp. 39-42, 92. Also Blast Furnace and Steel Plant, Nov. and Dec., 1937.
Note: Presented before Steel Session, 42nd Annual Convention, Cleveland, O., May 17, 1938.

electric furnace practice, as prescribed by the late George Batty and others, parallels to a large extent the operation carried out automatically during the converter blow. Such test reports of converter steel as have been made available indicate excellent properties.

One of the most interesting developments of high frequency induction melting of steel is the adaptation of such a furnace for slagging operations, as reported by Victor Stobie, in *The Engineer*, December 31, 1937. By dropping the primary coil with respect to the charge, providing sloping walls at the slag line, and putting on a permanent cover, he claims to have made refining and slagging operations both simple and rapid. Carbons as low as 0.02 per cent are readily obtained. Efficient dephosphorization and some desulphurization are claimed. The slag used is essentially a calcium silicate with iron ore added for oxidation. Among other things, it is claimed that the carbon content of the bath can be estimated closely by the appearance of the slag. The author indulges in much speculation and draws what are probably unwarranted conclusions but there seems to be no question of the ability to carry out such operations as are described.

The Siemens & Halske Company report (see abstract *Metal Progress*, June 1937) the installation and successful operation of a 6-ton high frequency induction furnace. Furnaces of 8-ton capacities are under consideration. The motor generator set delivers 1600 KW. at 500 cycles with voltages of 2300 to 3000. The construction of the furnace is simple to make maintenance easy and excellent power efficiency is claimed. An acid lining, prepared by ramming in quartzite, is used.

An article on the "Dephosphorization of Steel in an Induction Furnace by Alkaline Slags" written by H. Siegel, and abstracted by *Metals and Alloys*, July 1937, describes how about 0.05 per cent phosphorus was added to induction furnace melts (in magnesia crucibles) of 0.1 per cent carbon steel and the rate of dephosphorization by various basic slags observed. Adding sodium carbonate to slags high in lime and FeO, made the slags much more fluid and resulted in rapid dephosphorization.

In a report of an original research, K. Minato and S. Kobayashi of Japan Steel Mills Company, Ltd., describe work done on various alloy steels in a 500 Kg. (1100 lb.) acid lined, high frequency furnace. Diminution of inclusions was satisfactorily at-

tained by the following operation: Addition of ferromanganese (0.2-0.4 per cent) and acid slag (1-2 per cent) before melting of the charge, double removal of the FeO enriched slag, replacement of the latter by glass or $CaO-SiO_2$ mixture during the refining, cutting off electric circuit after last deoxidizer was added and again cutting in circuit for a few minutes before tapping. (Abstract *Metals and Alloys*, September 1937.)

Work on the development of the low frequency induction furnace is being carried on but there is no information of a definite nature available. It is hoped that the elapsed time between the preparation of this report and its presentation may produce something of interest, and if so, it shall be verbally reported.

In the report of this Committee for 1935, a number of details of the Sesci type furnace were given. Since then, efforts have been made to keep in close touch with the progress of this method of melting and the last information available was that, although the furnaces actually installed in a steel foundry were operating very well for certain classes of work, producing steel of a type suitable for castings, other sales of this type of furnace were not being pressed until such time as the technique of using a basic lining has been mastered. The reason for this would appear to be sulphur absorption from the fuel during melting. It can be seen quite easily that this factor could mitage against the furnace's success, when making a specification material.

So far as is known, no further use of the Brackelsburg furnace, as a steel producing medium, has occurred in this country during 1937, but it is evidently being used for this purpose in Europe, especially Germany and Sweden. A descriptive article, by A. Geudras, on its use appeared in April 1937. It is his claim that it is superior to the open-hearth, and that the carbon content can be held to within 0.01 per cent of the desired value. A complete process is described in detail for a furnace of 5000 Kg. (5.5 ton) capacity.

The Perrin process, mentioned in the 1936 report and discussed by Jerome Strauss, who gave a very useful outline of the position of this process, is further described in an excellent article entitled, "Perrin's Rapid Steel Refining," written by R. Perrin of Paris, France, which appeared in *Iron Age*, October 14, 1937. It is recommended to anyone really interested in the process, the article being very complete in its description of the aims and intentions

of the originator. It is hoped that by next year, developments here
or abroad will have proceeded to such a degree that some truly
informative comments can be included in the report of this Com-
mittee.

In the field of deoxidizers, one of the important developments
is the use of titanium to control the grain growth of medium man-
ganese steels with the result that yield strength and impact resis-
tance are markedly increased with no loss in ductility. See *Metals
and Alloys,* May 1937.

It is claimed by certain investigators that combined deoxi-
dizers, that is, alloys containing more than one deoxidizing element,
are superior to single element deoxidizers in conferring desirable
physical properties on steel. No details or reasons for the action are
available.

A study of the effect of aluminum on the physical properties
of Grade B steel was reported at a regional foundry conference by
C. E. Sims. The details of this study are being presented at one
of the sessions of this convention. In brief, it was found that if
aluminum is added in just sufficient quantity to completely deoxi-
dize a steel, the absence of iron oxide causes the iron and man-
ganese sulphides to precipitate as part of a eutectic to the great
detriment of ductility. An excess of aluminum, however, forms
some aluminum sulphide (Al_2S_3), which being less soluble than
the iron and manganese sulphides, causes the sulphides to precipitate
earlier and form larger and more widely spaced inclusions which
have decidedly less effect on the ductility.

The following published articles, some of which are referred
to in the text above, are suggested by your Committee as being
worth careful reading.

1.   Geiger, G. I., *"Basic Open Hearth Process,"* STEEL, March
29, April 12 and 26, 1937.

2.   Herty, Jr., C. H., *"Slag Viscosity Control Affords Greater
Uniformity in Steel,"* IRON AGE, June 17, 1937.

3.   Porter, John W., *"Open-Hearth Practices in a Steel Cast-
ings Plant,"* IRON AGE, November 23, 1937.

4.   Herty, Jr., C. H., *"Present Day Open-Hearth Control
Methods Compared to Conditions in 1924,"* A.S.M "THE REVIEW,"
January, 1938.

5.   National Open-Hearth Conference, Birmingham, Ala.,
IRON AGE, April 22, 1937.

6. Sullivan, John D., Witschey, R. A., *"Residual Metals in Open-Hearth Steel,"* METALS AND ALLOYS, April, 1937.

7. *"Hollow-electrode Furnace for Melting and Refining Stainless Steel Castings,"* THE FOUNDRY, February, 1937. Stainless steels made in this furnace are claimed to have freedom from porosity with excellent properties and corrosion resistance, as well as good machinability due to retained chromium oxide inclusions.

8. Wilcox, R. J., *"Processing and Properties of Electric Furnace Steel Castings,"* THE FOUNDRY, vol. 65, no. 6, June, 1937, pp. 90, 95.

9. Weitzer, H., *"Effect of Heat Insulation of Arc Furnaces on Current Consumption,"* STAHL U. EISEN, June 24, 1937. Tests were made on a number of Heroult furnaces of 6.5—15 ton capacity. Insulating the furnace vessels lowered current consumption 3 per cent and melting time 11 per cent.

10. Gregg, A. W., *"New Light on an Old Process,"* THE WHITING FOUNDER, Summer 1937.

11. Walker, T. R., and Dodswell, C. J., *"The Manufacture and Utilization of Electric Steel in the Foundry,"* PROCEEDINGS, Institute of British Foundrymen, 1936. The complete article did not become available until this year. It presents a detailed description of basic electric furnace practice for the manufacture of steel castings.

12. Geudras, A., *"Cast Steel From the Rotating Furnace Heated with Pulverized Coal,"* REV. FONDERIE MODERNE, April, 1937.

13. MacCaughey, W. J., Lee, H. C., *"Hearth Refractories for Steel Making,"* METALS AND ALLOYS, June, 1937.

14. Perrin, R., *"Perrin's Rapid Steel Refining,"* IRON AGE, October 14, 1937.

15. Dudar, A., *"Some Notes on Duplex Process,"* METALLURGY, May, 1937.

Respectfully submitted,

Committee on Methods for Producing Steel for Castings

H. D. PHILLIPS, *Chairman*

F. A. MELMOTH   C. E. SIMS

JOHN HOWE HALL   CHAS. D. DIETRICH

ERNEST LANCASHIRE

DISCUSSION

Presiding: L. E. EVERETT, KEY CO., E. ST. LOUIS, ILL.

H. D. PHILLIPS: Since this report was prepared, Mr. Melmoth, one of the Committee members, has received a letter from a friend in Europe pointing out that the Sesci position in Europe has hardly changed since last year where steel foundries are concerned. The basic lining experiments have not yet produced quite satisfactory results, but work is still being continued. The plant working on steel castings is the Crewe Works of the London, Midland and Scottish Railroad Company in England. The furnace is fitted with an acid lining and doing very well indeed. They have reached a figure of 108 heats per lining, which brings the refractory cost to about $1.00 per ton. They feel the acid lining problem has been solved but in the opinion of Mr. Melmoth's friend, the full value of the furnace will not be realized until the basic lining has reached its full development.

This, of course, is probably due to the sulphur absorption from the fuel.

Another development possibly credited to 1937, reported to me by our Division Chairman, J. H. Hall, is the development of direct pouring from induction furnaces to molds. His statement is that this practice results in an improved yield and in more uniform casting temperature. This practice is being followed in at least one plant.

CHAIRMAN EVERETT: Of particular interest in the district in which I am now located, is the utilization of the cupola in connection with an open hearth furnace. Mr. Gregg can probably give further information on this.

A. W. GREGG[1]: The installation referred to by Mr. Everett is a hot-blast cupola, desulphurizing-ladle, holding furnace combination in a steel plant which has no blast furnaces. It is now operating on an all-cold charge. It is proposed to melt all scrap (cast iron and steel) in the cupolas. The metal, after desulphurizing, will be transferred to the holding furnace from which it will be delivered to the open hearth furnaces as required. This is a very interesting application of the cupola and holds tremendous possibilities for cutting the cost of producing open hearth metal. The installation would be warranted in a steel foundry which has considerable tonnage. The installation referred to will be started soon and we will probably have some very interesting figures to report at the next convention.

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[1] Foundry Engineer, Whiting Corporation, Chicago, Ill.

Report of Steel Division Committee on Heat Treatment of Steel Castings

To Members of the Steel Division, American Foundrymen's Association:

Your committee on heat treatment has kept in close contact with all published information relating in any way to any type of heat treatment as applied to steel castings and presents the following short report covering significant developments during the year.

From the standpoint of furnace equipment, the most important development reported during the year was a description of a five-zone, continuous heat treating furnace for the heat treatment of manganese steel castings. This furnace was described by John Howe Hall in *Iron Age* of July 22, 1937, in an article entitled "Continuous Heat Treating Furnace in a Steel Foundry." Although the use of the continuous furnace in the steel foundry is not a new development, the construction of this furnace is of such a nature that it merits further consideration in this report.

Instead of using the usual alloy steel rail, or walking beam or endless chain construction characteristic of continuous furnaces, this furnace employs regular cars which are mechanically pushed through the furnace. Castings are piled directly on the cars thus eliminating the high initial and maintenance costs of the high alloy trays, etc.

The furnace has an inside length of 30 ft., and contains five heating zones, separated only by arch walls having relatively small clearance between the charge and wall.

Originally the following temperatures were maintained in the respective zones: Zone 1, 300°F.; zone 2, 600°F.; zone 3, 940°F.; and zones 4 and 5, 1940°F. The temperature control system, however, is such as to provide extreme flexibility in each of the zones. For example, zones 1 and 2 have a range of 0-1000°F.; and zone 3

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has a range of 800-1800°F. Zones 1 to 3 hold one car each, zone 4 holds two cars and zone 5 one car.

Since this furnace was designed for the heat treatment of Hadfield manganese steel, it is so constructed that when the car is pushed from the final zone the whole car may be mechanically tipped thus dumping the charge into the quench tank within 40 seconds.

This furnace is oil fired and has effected a fuel saving of 35 per cent over the batch type equipment formerly used, reducing the oil consumption from 35 to 22 gal. per ton. The author further indicates a saving through reduction of the amount of scale and estimates a 1 per cent increase in the shipping weight of the castings.

Although installed for the heat treatment of Hadfield Manganese steel, this furnace also has been used for the heat treatment of carbon and low alloy structural steel castings. In a few instances, castings have been simultaneously heated to three different temperatures.

No significant changes have occurred in the temperatures or procedures involved in the foundry methods for full annealing, normalizing, liquid quenching and drawing other than the degree to which they have been applied.

In those foundries producing medium or small steel castings, there has been a trend toward the greater use of double or triple heat treatments. This is especially true in those plants making considerable quantities of the low alloy steels. This, to some extent, has been brought about by the increased extent to which impact testing is being applied to cast steel products. The beneficial effects of double normalizing and drawing or liquid quenching and drawing, advocated for several years by various foundry metallurgists, are becoming more widely recognized resulting in the trend already mentioned.

R. C. Heaslett, in an article in *The Foundry* for December, 1936 and January, 1937 entitled "Correlating Physical Properties with Service Requirements," states that of the three factors cleanliness of steel, analysis and heat treatment affecting resistance to impact, heat treatment is the most important. He gives the following physical properties for a heat containing carbon 0.29 per

cent, manganese 0.70 per cent, silicon 0.38 per cent, phosphorus
and sulphur 0.03 per cent as shown in Table 1.

Table 1

	Full Annealed	Annealed and Normalized	Liquid Quenched and Drawn
Yield Point (lb. per sq. in)	38,000	40,000	46,000
Tensile Strength (Lb. per sq. in.)	65,000	67,000	70,000
Elongation in 2 in., (per cent)	29.0	26.0	25.0
Reduction of Area, (per cent)	40.0	45.0	52.0
Izod Impact (ft. lbs.)	12.0	22.0	40

In some instances, liquid quenched and drawn carbon steel
castings have replaced single normalized and drawn low alloy cast-
ings with equally good results and at lower cost. Properties of some
double normalized and drawn cast steel are reported by T. N. Arm-
strong in an article entitled "Nickel Steel Castings in Rolling
Stocks" published in *Metal Progress* for February, 1938.

Greater use of liquid quenching, proportionally, has resulted
in improvements in the mechanical handling of castings in various
stages of the process, resulting in a more uniform product at lower
cost.

In the article by R. C. Heaslett, already referred to, he points
out the necessity for the proper selection of chemical composition
if liquid quenching is resorted to in developing the best properties
for the service intended. Two steels, given the same water quench-
ing treatment, on test pieces ranging from 1 to 5-in. in diameter
show the hardness values. Given in Tables 2 and 3.

Flame hardening, mentioned by your committee in its report
for 1936, has made significant advances during the year and at the
present time, there are a considerable number of installations
regularly treating a large variety of cast steel parts. The improve-
ments which have been made during the year may be divided into
two divisions. On the one hand, there has been considerable im-
provement in the mechanical set-ups involved so as to widen the
application of the process and secondly, considerable information
has been accumulated regarding the chemical compositions most

suitable for flame hardening and the service performance of cast-
ings treated in this manner. The process has been found especially
adaptable to the surface hardening of gear teeth of medium or
relatively heavy sections.

Table 2

0.50 PER CENT CARBON STEEL

Diameter, Inches	Brinell Hardness at Surface	Brinell Hardness at Center
1	592	321
2	401	285
3	341	248
4	293	229
5	255	223

Table 3

0.40 PER CENT C, 0.80 PER CENT CR., 0.25 PER CENT MO. STEEL

Diameter, Inches	Brinell Hardness at Surface	Brinell Hardness at Center
1	601	578
2	555	514
3	534	477
4	495	429
5	461	401

Flame softening, although not yet commercially applied to
steel castings, has been utilized in softening the hardened edges
of flame cut low alloy steel plate, as reported in *Machinery* for
March, 1938.

A new process for heat treating, not yet applied to the product
of the foundry, has been reported by E. Davenport in an article
entitled "Heat Treatment of Steel by Direct Transformation of
Austenite—Austempering," published in *Steel* for March 29, 1937.
The process consists of heating to a suitable temperature above the
critical range and then quenching in a quenching bath held at 300
to 800°F., depending upon the results desired. This results in the
direct transformation of the austenite direct to the desired metal-
lurgical constituent. This is in contrast to the usual transformation
to martensite followed by tempering to the desired constituent
which occurs in normal liquid quenching. This process results, at
a definite hardness, in a very pronounced increase in the reduction
of area and shock toughness. One of the disadvantages, at the

present time, is that it requires a very careful selection of steels and can be applied only to thin sections up to ⅜-in.

With the increase in the manufacture of the already known alloy steels and the development of new compositions, the use of preheating before flame cutting of risers, or before welding, is being practiced to a considerable degree by many foundries. However, considerable variation occurs in the temperatures used and the position of the preheating process in the cycle of other foundry operations. However, no published information has come to the attention of your committee in this respect.

Looking into the future, your committee in 1938, proposes to make a survey of the types of heat treatment being used in the steel foundry, the degree to which each is used, and the results which are being obtained.

Although the above report summarizes the most important developments of the year which have come to the attention of your committee the following published articles are recommended by the committee for further reading.

1. Heaslett, R. C., *"Correlating Properties with Service Requirements,"* THE FOUNDRY, vol. 64, no. 12, pp. 26-27-71 (1936), vol. 65, no. 1, pp. 28-29-81-82 (1937).

2. Hall, J. H., *"Continuous Heat Treating Furnace in a Steel Foundry,"* IRON AGE, July 22, 1937.

3. Wilcox, R. J., *"Processing and Properties of Electric Furnace Steel,"* THE FOUNDRY, vol. 65, no. 6, pp. 90, 95, June, 1937.

4. *"Typical Properties of Annealed and Heat Treated Alloy Steel,"* Data Sheet, MACHINERY, May, 1937.

5. Armstrong, T. N., *"Nickel Steel Castings in Railroad Rolling Stocks,"* METAL PROGRESS, Feb., 1938.

6. Nelson, W. L., *"Properties of Metals for Use in Low Temperature Processes,"* OIL AND GAS JOURNAL, Oct. 21, 1937.

7. Davenport, E., *"Heat Treatment of Steel by Direct Transformation of Austenite—Austempering,"* STEEL, March 29, 1937.

8. *"Flame Hardening,"* STEEL, Sept. 27, 1937.

9. Rolf, R. L., *"Selective Hardening with Oxyacetylene Flame,"* STEEL, vol. 101, no. 18, pp. 49-51 (1937).

10. Seemann, A. K., *"Surface Hardening with Oxyacetylene Flame,"* METAL PROGRESS, Sept., 1937.

11. Sykes, W. E., *"Torch Hardening Methods for Gears,"* STEEL, Sept. 27, 1937.

12.   Jones, J. A., "*Present Trend in Engineering Alloys*," IRON AGE, May 20, 1937.

13.   Juraschek, F., "*Portfolio of Heat Treatment Handling Methods*," IRON AGE, Oct. 28, 1937.

14.   "*Heat Treating Developments in 1937*," STEEL, Jan. 3 1938.

15.   "*Flame Softening*," MACHINERY, March, 1938.

16.   Whitely, J. H., "*Precise Control of Atmospheres Required to Prevent Surface Reactions*," METAL PROGRESS, vol. 32, no. 6, pp. 783 (1937).

Respectfully submitted,

Committee on Heat Treating of Steel Castings

D. C. ZUEGE, *Chairman*   H. H. BLOSJO
WERNER FINSTEN   E. R. YOUNG

DISCUSSION

Presiding: L. E. EVERETT, Key Co., East St. Louis, Ill.

E. C. TROY[1] *(Written Discussion):* The writer would like to call Mr. Zuege's attention to his (Mr. Zuege's) remarks concerning the quenching characteristics of low carbon (non-alloy) steels.

There is some question in the writer's mind as to advisability of trying to water quench the steels under discussion to meet specification properties.

It would seem self evident that the cooling rate and not the cooling medium is the variable which introduces various physical properties into the steel.

Mr. Zuege's arguments that the final microstructure would be the same after a high temper regardless of the original degree of undercooling, are, in all likelihood, very near correct. However, there will be a difference which may be easily noticed in the data in Tables 4 and 5. This difference would be even more exaggerated had the tests been made of castings of ever greater difference in volume.

It would be next to impossible to obtain representative test specimens for a group of different size castings. The differential shown between small and large test specimens from the same melt, normalized, would be even greater were we to include quenching of these low carbon steels as a commercial practice, assuming, of course, that the necessary degree of control to quench to a uniform microstructure or hardness is lacking in the foundry.

FRED GROTTS[2]: There ought to be something more said in regard to quenching steel castings. Practically all of our literature emphasizes the fact that the quenching of castings is very dangerous and should be restricted. That line of reasoning is obsolete and ought to be qualified and eliminated as much as possible.

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[1] Metallurgist, Dodge Steel Co., Philadelphia, Pa.
[2] Vice-President, Lebanon Steel Foundry. Lebanon, Pa.

Table 4

ANALYSIS OF MELT

Carbon, per cent.........................0.25
Manganese, per cent......................0.77
Silicon, per cent.........................0.35
Sulphur, per cent.........................0.030
Phosphorus, per cent.....................0.026

HEAT TREATENT

Spec. No. 1—Heated 1650°F.—held 2 hr.—furnace cooled.
Spec. No. 2—Heated 1650°F.—held 2 hr.—air cooled.
Spec. No. 3 and 4 (legs from a double leg test bar)—Both
    heated to 1650°F.—held 2 hr.—quenched in
    water.
    No. 3 quenched in (water at 65°F.)
    No. 4 quenched in (water at 53°F.)
    Both No. 3 and No. 4 tempered together for
    2 hr. at 1225°F. and air cooled.
    Specimen No. 3 after quenching 227 Brinell.
    Specimen No. 4 after quenching 340 Brinell.

Table 5

PHYSICAL PROPERTIES

Specimen No.	No. 1	No. 2	No. 3	No. 4
Tensile Strength, lb. per sq. in...	68,900	75,250	84,250	88,450
Yield Point, lb. per sq. in.......	37,400	43,700	57,300	64,850
Elongation in 2-in., per cent....	31.0	30.5	27.0	27.5
Reduction of Area, per cent...	45.7	50.6	54.6	59.3

There are many foundries in the United States that are water and oil quenching a big percentage of their castings. Our technical meetings have hesitated to bring out this point of quenching. This particular item in Mr. Zuege's paper should be further emphasized.

Distinctly better physical properties are secured by quenching. The idea that quenching steel castings develops cracks and strains has been unnecessarily stressed and should be corrected, as forgings and rolled sections will crack and strain in the same manner if improperly handled. It is my experience that intricate forms can be quenched and drawn by using judgment and that, in some cases, the strains are less than when air cooled. The fact that a liquid medium gives a uniform result, while air quenching results in patches of the casting being dark while other patches may be red hot.

I appreciate the importance of the section of this paper given to quenching as I have observed the results obtained for the past 10 years by a certain foundry that treats large intricate forms and this foundry has always had a very superior product.

JOHN HOWE HALL[3]: As you probably all know, I was the first to quench steel castings as a regular practice, beginning in 1909, and I agree with most of what Mr. Grotts speaks about, but there is one thing that has not been pointed out. If steel castings are to be quenched, they have to be quenched with good judgment.

I think the reason quenching of steel castings has hung back as much as it has, is because people know that in a great many foundries they will not use good judgment and in addition, there are many consumers who will use even less sense than the foundries. If you allow some people to go as far as they like, they will be demanding that everything be quenched. If you are going to quench a large casting you need proper equipment so that it will not be quenched until it is entirely cold, but will be taken out while still very warm and placed in another furnace to be "drawn". Of course, there are not many foundries as yet that are equipped to do such work properly.

In 1913, a friend of mine who worked at what was then the Pennsylvania Steel Co., water-quenched and drew two heavy Mayari steel locomotive frames for Pennsylvania Railroad engines, by following the above described practice.

MR. ZUEGE. Mr. Hall is right in that judgment must be used in quenching castings. However, looking at it from the other viewpoint, there are a good many more castings that can be satisfactorily liquid-quenched than there are castings that cannot be satisfactorily liquid-quenched, providing reasonable care is taken in quenching methods.

Of course, in the foundry that produces the smaller steel castings, the equipment involved is not as costly as it is where big castings are handled.

Liquid quenching does have a very definite field (speaking from past experience at our plant) in the casting foundry producing intermediate or small steel castings.

CHAIRMAN EVERETT: Mr. Grotts, what would be your suggestion as to any proper steps that we might request our Committee to take in connection with these specifications that prohibit quenching?

MR. GROTTS: There should be some action taken but I hardly know how it should be handled. There should be more discussing of such articles as this one by Mr. Zuege and more stressing of the value of the physical properties obtained. Another reason for the pushing of this article is for the purpose of enlarging the field for steel castings. I feel if we can change the attitude of distrust as regards variables and breakage, the field for steel castings will be enlarged.

[3] Consultant on Steel Castings, Germantown, Philadelphia, Pa.

In regard to physical properties, I am sure that impact values will be increased. Where we anneal or normalize and draw, in some cases, we get about 20 ft.-lb. Izod. This can be increased readily to 40 to 60 ft.-lb. Izod by quenching and drawing.

Impact committees exist in practically all technical organizations. They recognize that impact value is one of the most important items in connection with the success or failure of the steel casting. By quenching, impact value is improved as well as yield point and reduction of area.

Quenching of castings will bring out special physical properties, which are very important as selling points and new developments where we have to meet the competition of welded structural designs and general welding propaganda. Steels that will take special treatments are hard to duplicate in welded designs.

E. C. TROY[4]: I assume the steels under question are of that class known as low carbon non-alloy.

It seems necessary to control the quenching rate of this type of steel to a very close margin if uniformity is to be obtained.

The results obtained under careful control in the laboratory are hardly indicative of the results which would be obtained under shop conditions. The water temperature, scale formation, and size of section are all variables of great importance. It would be difficult to obtain representative test specimens of assorted castings after a quenching treatment.

MR. HALL: I do not mean to imply that I do not agree with Mr. Grotts, but the first step is to get engineers to ease up in their specifications on this prohibition of quenching. We have made a good deal of progress already. It is not very long ago that Committee A-4 of the A.S.T.M., of which I was then vice-chairman, drew up under my direction a recommended practice for the heat treatment of steel castings. When the full Committee considered my draft, they first cut out quenching altogether; next they cut out normalizing and drawing. They did allow plain normalizing, after a fashion; that is, they thought it would be permissable to open the doors of the furnace and accelerate the cooling a little. A motion was made that the title of the draft be changed from "Recommended Practice for the Heat Treatment of Steel Castings" to the "Recommended Practice for the Annealing of Steel Castings." When I inquired the reason for that motion, the assembled engineers and consumers said they did not like to have the idea that castings could be heat-treated; they ought to be annealed, and nothing more.

That was not many years ago, and now we have A.S.T.M. specifications which allow normalizing and even quenching of castings of certain types. We have gone a long way in this direction, and now a few more hard shoves will get us where we want to be.

MR. ZUEGE: In quenching for hardness, where you are trying to get a very definite Brinell hardness range, it is very important to control all of the conditions under which the castings are quenched because otherwise you do not stay within your hardness range.

When you are quenching for toughness, (and I take it that right now we are speaking more about quenching for toughness than we are for hardness), a good many of these variables iron themselves out because you give the castings a high draw. After the material is quenched, regardless of whether it is martensitic, troostitic or sorbitic, after the high draw, it will be changed to sorbitic anyway.

So these variations, that might seem to be critical, are not as critical as Mr. Troy would have us believe, especially when quenching for toughness. If quenching for hardness, that is a different proposition. Then each casting is heat-treated individually under very carefully controlled conditions.

C. E. SIMS[5]: A little clarification is needed here. I agree with Mr. Grotts that there should be no prohibition against the quenching of castings, for the reason that, in many cases, one can get the properties desired better that way than in any other way. At the same time, this matter of whether to quench or not to quench has to be decided by the particular conditions. There are certain compositions that it is almost impossible to injure with quenching. There are others that are very susceptible to cracking. The composition has to be designed for the purpose in mind.

I am talking about design of composition. Some steels are designed to have optimum properties in the normalized condition. A good example is a fine-grained, intermediate manganese steel. These steels attain their optimum combination of properties in the normalized state, and serve a very large field. There are other compositions which have very poor properties in the normalized state but have excellent properties in a quenched condition. Such compositions should be quenched. There is no more justification, however, in contending that all castings are improved by quenching, than there is to say that no castings should be quenched.

MR. GROTTS: What I want to do is to correct the inherent fear, held by many people, of variables in steel castings. What has been discussed applies to forgings as well as rolled sections. Again referring to quenching steel castings, you should use your best judgment and, of course, analyze the section in regard to heating and cooling, but the same applies to all materials.

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[5] Supv. Metallurgist, Battelle Memorial Institute, Columbus, Ohio.

Analysis of Non-Ferrous Foundry Defects

Foreword

In the fall of 1934, the Non-Ferrous Division, through its Advisory Committee, appointed a committee to lay out a schematic procedure for the analysis of defects commonly found in sand castings of brass and bronze. This committee first listed and defined what it called the symptoms or those imperfections, usually visible, which call attention to the fact that a casting is defective. This report was published in A. F. A. Transactions, Vol. 43 (1935), pp. 247-250. The committee then continued its work and herewith publishes its second report classifying defects and listing probable causes.

The personnel of this committee is as follows:

> Chairman, H. M. ST. JOHN, Detroit Lubricator, Co., Detroit

> L. H. FAWCETT, U. S. Naval Gun Factory, Washington, D. C.

> HAROLD J. ROAST, Canadian Bronze Co., Ltd., Montreal, Canada

> WM. ROMANOFF, H. Kramer Co., Chicago, Illinois

Second Progress of Non-Ferrous Division Committee on Analysis of Defects

Members of Non-Ferrous Division, American Foundrymen's Association: In this, the second progress report of your committee, we present a classification of casting defects, listed according to their appearance, evident to the eye, which marks castings as defective. This classification is shown below in the left hand column. After each classification is given a reference number which indicates the probable cause of defect. These probable causes are listed in the right hand column below under their respective reference numbers.

In spite of the considerable amount of study spent upon this

outline, your committee feels that it is still imperfect and incomplete. Criticism and suggestions will be welcomed to the end that a more perfect and useful analysis of casting defects may be formulated.

Respectfully submitted,

Committee on Analysis of Casting Defects

H. M. St. John, *Chairman*

Note: Explanation of Classification Listings of Defects and Causes.

In the following pages the classification of defects is given in one column with the probable cause in the column opposite. The numbers appearing before the *Causes* in the right hand column refer to the *Defect* numbers in the left hand column.

Classification of Defects

1. **Shift.** A casting in which the cope and drag portions do not exactly match at the parting line.
Cause reference: A-1.

2. **Misrun.** A casting which lacks completeness due to the fact that the mold cavity has not been wholly filled with metal. There may be a smoothly rounded hole through the wall of the casting, or one or more extremities may be only partly filled out.
Cause references: B-2, C2, E-2, F-2, G-2, K2.

3. **Cold Shut.** The casting appears to be cracked, but on closer examination, it is found that the metal has failed to join or coalesce along the line of the apparent crack. In some cases, partial coalescence leaves a line of weakness which later does crack.
Cause references: B-3, C-3, D-3, E-3, F-3, G-3, K-3.

4. **Crush.** A casting showing deformation, apparently due to displacement of sand when the mold was closed.
Cause references: A-4, B-4, D-4.

4a. **Swell.** A casting showing deformation on cope side due to displacement of sand by pressure of metal.
Cause reference: B-4.

5. **Variation in Wall Thickness.** A casting which at one or more points shows more or less metal than is called for by specification. It may be uniformly too heavy, uniformly too light, or may be too thick on one side of a cored cavity, too thin on the other.
Cause references: A-5, D-5.

Classification of Probable Causes of Defects

A—Condition of Pattern, Flask, Core Box, etc.

(1) Loose dowel pins and inaccurate fitting of cope to drag will result in shifts.

(4) If the core print is too small for the core, a core crush will result.

(5) Worn core boxes will cause the core to be too large, resulting in too thin a casting wall. If the pattern makes too large a core print, the core may shift or float, resulting in a variation in metal thickness.

(6) In the case of a core crush, a sand wash is also likely to develop.

B—Position of Mold

(2) The mold must be so placed that the molten metal in the sprue will be substantially above the level of the highest casting, in order that the hydrostatic pressure will be sufficient to prevent misruns.

(3) A cold shut may result from failure to take precaution B-2.

(4) The weights used must be heavy enough to prevent swells but not so heavy as to cause a crush.

C—Gating

(2) The gates leading to the individual castings in the mold must be large enough to permit free flow of the metal. Otherwise, misruns may result.

(3) Restricted flow of metal may also cause cold shuts.

Defects

6. Sand Wash. The casting may have rough lumps of metal at some points on its surface or may exhibit rounded corners which should be sharply defined. At other points there will be roughly granular depressions or holes.

Cause references: A-4, D-6, E-6, H-6.

6a. Scab. A rough spot, usually on a thin-walled portion of the casting, the wall being slightly thicker than normal at this point, porous and containing sand grains.

Cause reference: D-6a.

7. Sand Blow. The casting shows an unnaturally smooth depression at one or more points on its outer surface.

Cause references: D-7, E-7.

8. Core Blow. The casting shows a smooth depression, sometimes black, sometimes golden, on an inner surface where there is a cored cavity or, often, a gas pocket in some heavy portion of the casting above the level of the cored cavity.

Cause reference: H-8.

9. Burning Into Sand. Certain outer parts of the casting have a rough, sandy appearance as if the metal had penetrated freely between the sand grains, some of which are completely surrounded and enclosed in the outer wall of the casting.

Cause references: E-9, F-9 & 10.

Causes

due to inadequate gating.

(12a) "Wormy" surface near the gate is sometimes due to inadequate gating.

(14) Shrinkage cracks may frequently be due to improper gating or to the lack of risers where needed.

D—Molding Practice

(3) A slight crush at the parting line may force a fin of sand into the metal, producing an appearance very similar to that of a cold shut. A true cold shut may also be due to insufficient venting.

(4) Crushes are usually due to carelessness on the part of the molder.

(5) Variation in wall thickness may be caused by uneven ramming.

(6) Too light ramming may result in sand wash.

(6a) Too heavy ramming may result in scabs because the mold face is so dense as not to have room for expansion when exposed to heat of metal.

(7) A sand blow may be caused by insufficient venting.

E—Quality and Condition of Molding Sand

(2) If the sand is too wet, it may chill the metal excessively and cause misruns.

(3) Wet sand, by chilling, may make the metal sluggish and result in cold shuts.

(6) If the sand has insufficient bond, or is too dry, the mold may be weak, tending to permit sand wash.

Defects

10. **Burning Into Cores.** Sometimes a rough, sandy inner surface, similar to the penetration of metal into green sand, but more commonly in the form of metal fins penetrating into the core and containing trapped grains of core sand.

Cause references: F-9 & 10, H-10.

11. **Sand Sticking in Cored Cavities.** Even when there is no apparent "burning in," castings having intricate cored passages, or cavities of small dimension and relatively inaccessible, sometimes show a tightly adherent coating of sand.

Cause reference: H-11.

12. **Surface Imperfections.** Castings otherwise of apparently perfect quality sometimes have surface defects which may be of importance on the ground of appearance only. In other cases, these apparently superficial defects are indicative of more deep seated ailments. The following items, 12a, b, c, d are listed as typical.

12a. **Wormy Surface.** The surface of the casting, usually in the vicinity of the gate, shows irregular depressions, shallow but elongated, similar in appearance to worm tracks. These depressions are often filled with a deposit of zinc oxide, and are sometimes accompanied by a very poor fracture.

Cause references: C-12a, F-12a.

12b. **Surface Stain.** The surface of the casting shows a black discoloration of varying size and shape.

Cause reference: H-12b.

Causes

(7) Wet sand, or sand with too low permeability, is likely to result in sand blows.

(9) Sand which is too coarse, too dry, or has too high a permeability, may permit the metal to burn-in.

(12d) Coarse sand or dirty sand which contains fragments of cores or metal, will cause the surface of the casting to be rough or pitted.

F—Quality of Molten Metal

(2) Sluggish metal may cause misruns as a result of contamination by impurities. These are most likely to be oxygen or sulphur, present in the metal as charged into the furnace, or by contamination from fluxes or furnace fuels.

(3) Cold shuts are also influenced by impurities in the metal, as stated in F-1.

(9 & 10) Excessively fluid metal resulting from the use of too much phosphorus, or from the presence of such impurities as aluminum and silicon, is likely to burn into molding sand and cores.

(12a) In leaded alloys, very small percentages of silicon (0.03 per cent) or aluminum are likely to produce a "wormy" surface.

(12c) Contamination by silicon also tends to produce lead sweat. Silicon may have been present in scrap used in the furnace charge, may be picked up from silicon carbide in a furnace lining or, in the electric

Defects

12c. **Tin or Lead Sweat.** The surface of the casting is more or less covered with a thin layer of white metal. In the case of lead, the sweat often occurs in spots or lumps, sometimes of considerable thickness.

Cause reference: F-12c.

12d. **Rough or Pitted Surface.** Although the casting does not show evidence of sand washing or scabbing, the surface is rough or exhibits an occasional angular pit, sometimes so deep as to leave an objectionable scar on a finish-machined surface.

Cause references: E-12d, F-12, K-12.

13. **Solid Inclusions.** With a fracture which otherwise appears to be good, the walls of the casting contain particles or small chunks of non-metallic substance, or separate pieces of metal not coalesced with the body of the casting.

Cause references: F-13, K-13.

14. **Shrinkage Cracks and Cavities.** With a fracture apparently good, the casting shows at one or more points a crack or cavity where the metal has pulled apart while it was still in a plastic condition. The walls of the shrinkage cavity are usually tarnished to a color which varies from orange to dark brown.

Cause references: C-14, G-14, L-14.

Causes

furnace, may be produced by the action of oil, electrode smoke or other carbonaceous material.

Lead sweat is most likely to result from too low a proportion of tin in the alloy. With lead as high as 10 percent, the percentage of zinc also should be maintained at 6 percent or above.

It is possible that the presence of an excessive amount of gas (as hydrogen, for example) dissolved in the molten metal, may force lead to the surface of the casting.

(12) Very fluid fluxes cannot be skimmed readily from the surface of the metal and are likely to be carried into the mold, resulting in surface imperfections on the casting.

(13) Solid inclusions in the walls of the casting, apparent in the fracture, are also often due to "soupy" fluxes and slags.

(15) A coarse, dendritic, discolored fracture is most commonly due to the presence of silicon or aluminum in the alloy.

(16) Gas porosity frequently results from the presence of hydrogen or sulphur dioxide, absorbed by the molten metal from the furnace atmosphere. Hydrogen is often present as a result of water or oil in the furnace charge.

Defects

15. Weak or Discontinuous Structure. The fracture of the casting is bad at practically all points. Commonly the structure is dendritic with minute fissures between the large crystals, which are tarnished to an orange or brown color. Sometimes, with crystals of more normal size, the fracture is of a loosely granular appearance rather than fibrous. Other varieties of abnormal fracture may be encountered.

Cause references: F-15, G-15.

16. Gas Porosity. This condition, sometimes called sponginess, is marked by small but clearly discernible gas bubbles, usually segregated near the surface, but underneath the skin, of heavy sections of the casting. The cavities are approximately round and bright, free from tarnish.

Cause references: F-16, G-16, L-16.

Causes

G—Pouring Temperature

(2) Metal which is sluggish, because it is too cold, will often cause misruns.

(3) Cold, sluggish metal is commonly responsible for cold shuts.

(14) If other conditions are correct, shrinkage cracks and cavities may safely be charged to too low a pouring temperature.

(15) Too high a pouring temperature will result in a coarse, weak crystalline structure.

(16) Gas porosity is often due to too high a pouring temperature.

H—Quality and Condition of Cores, Parting Compound and the Like

(6) An overbaked core will be weak on the surface and may result in core-sand wash.

(8) Insufficient baking may cause a core blow.

(10) If the cores are baked too rapidly, or if the amount of binder used is not properly proportioned, they may crack and permit metal fins to form. Overbaking may result in burning-in.

(11) Sticking of core sand in inaccessible cored passages may be prevented by the use of a rubber core wash.

(12b) Surface stains are sometimes due to the excessive use of flour or molasses.

Causes

K—Skimming and Pouring Practice

(2) If pouring is interrupted or slackened before complete, a misrun may result.

(3) A fragment of oxidized metal from the lip of the pouring ladle will sometimes lodge in the wall of the casting, producing an appearance similar to cold shut.

(12) Faulty skimming or careless pouring may result in fragments of slag or oxidized metal entering the mold and lodging on the surface of the casting.

(13) Solid inclusions in the body of the casting may be caused as indicated in K-12.

L—Design of Casting

(14) The casting design should be such as to avoid heavy sections and light sections in close proximity. If this is not possible, a fillet of ample radius will tend to prevent shrinkage cracks at corners.

(16) Without such fillets there will be a tendency to pour at such high temperatures as to result in gas porosity in heavy sections.

Report of Steel Division Committee on Test Coupons

To Steel Division, American Foundrymen's Association:

During the past few years your committee has attempted to collect data upon the mechanical properties obtained on specimens procured from the casting as compared to the properties of a standard test coupon. These attempts have not been successful as the individual foundries solicited have been reluctant to furnish information of this type. In view of the fact that the mechanical properties obtained on samples trepanned from the heavier sections of castings do not equal those of the standard coupon poured from the same heat the committee feels this reluctance has been justified in that such data might injure the particular foundry submitting the data.

Data of this type, however, are becoming more and more important as engineers are now insisting upon trepanned samples in many instances. Unfortunately, many of these engineers have had no foundry or metallurgical experience and expect these trepanned samples to possess the properties ordinarily given in a handbook for the particular analysis in question. This has, in many cases, caused considerable difficulty to foundrymen producing castings to specifications of this type. Therefore, your committee prepared and circulated a questionnaire in an endeavor to learn whether or not any data pertaining to this subject was available and, if so, whether it would be submitted to the committee for study. The committee offered to furnish standard forms and accept data sent in anonymously in order that the results would indicate the position of the industry rather than that of a few individual foundries.

At the present time thirty-one organizations have returned their questionnaires. Thirteen, or 45 per cent, of these organizations have indicated they possess information of the type desired

which they will submit to the committee. All of these organizations indicated a preference for the standard form. Eleven organizations wished to be named as contributors of data, two organizations preferred to remain anonymous, and none desired credit for any specific data. Several organizations have already submitted data for study.

Returned questionnaires usually contain additional notes or qualifying statements which are sometimes quite important if they are properly analyzed. The note most frequently encountered on this questionnaire was to the effect that the contributor had some data but not sufficient to warrant its submission to the committee. Your committee would like to point out that in a study of this type each scrap of data is important. Data on one or two heats which, as far as one foundry is concerned are "unimportant" are very useful when added to several additional "unimportant" scraps of information.

It has also been disclosed in conversations with members of the committee that in some cases the questionnaire did not reach the proper people in the organization and as a result information that is available for use has not been reported.

Therefore, your committee desires at this time to appeal to all members of the A.F.A. who have any data upon this subject to submit it to the committee for study in order that this information can be made available to engineers and considered in the preparation of specifications.

<div align="center">

Respectfully submitted,

Steel Division Committee on Coupon Tests

</div>

J. M. SAMPSON, *Chairman*	W. C. HARTMAN
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FRED GROTTS	H. D. PHILLIPS
V. T. MALCOLM	

Physical Properties of Cast Iron In Heavy Section

By E. R. Young[1], V. A. Crosby[2], and A. J. Herzig[2], Detroit, Mich.

Abstract

The authors have gathered data on 68 different plain and alloyed irons from companies in various parts of the United States to study the physical properties of cast iron in 2-in. sections. The results were obtained from data secured on A.S.T.M., type "C", test bars 27-in. long and 2-in. diameter broken on 24-in. centers to secure transverse test data. Tensile tests were made on bars machined from the broken sections of the transverse bars. Brinell tests also were made. From these data, modulus of resilience and modulus of rupture were calculated. Impact tests on 1.125-in. diameter round, ground bars machined from the broken half of the 2-in. diameter arbitration test bars, also were made. On the basis of results obtained, the authors conclude that 70,000 lb. per sq. in. tensile strength unheat-treated cast iron of 2 in. section is available today commercially.

1. The physical properties of cast iron are so related to the size of the section in which the iron is cast that one should not discuss cast iron properties without explicit knowledge of the section under consideration. This close relation long has been recognized to apply to common cast iron by foundrymen, by metallurgists and, to some extent, by design engineers.

2. The published work on the development of high-strength cast iron, which development has been in progress for the past few years, has been confined largely to section sizes represented by the standard A.S.T.M. 1.2-in. diameter arbitration test bar. As a result, the opinion is prevalent that the high strength of the newly developed irons is available only in comparatively small sections. It is the purpose of this paper to survey briefly the field of heavy section, high strength iron and, by means of a summary showing

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Table 1

CHEMICAL AND PHYSICAL PROPERTIES OF 68 DIFFERENT IRONS

Melting No.Unit*	T.C. %	C.C. %	G.C. %	Si. %	Mn. %	S. %	P. %	Cr. %	Ni. %	Mo. %	V. %	Ti. %	Cu. %	Treated	Transverse St'gth. lb.	Deflection, in.	BHN center	BHN edge	Tensile St'gth. lb. per sq. in.	BHN	Resilience, in.-lb.	Mod. of Rupture	Impact, ft.-lb.
1 C	2.70	0.72	1.98	2.33	1.08	0.09	0.07	0.23	1.10	0.63					14,750	0.368	269	286	60,400	273	3117	------	42.8
2 C	2.76	0.75	2.01	2.42	1.02	0.09	0.08	0.20	1.10	0.55					14,520	0.370	265	269	64,000	269	3086	------	53.5
3 C	2.87	0.67	2.00	2.54	1.10	0.10	0.08	0.20	0.72	0.43					14,200	0.307	262	269	65,400	269	2277	------	55.0
4 C	2.90	0.79	2.20	2.30	0.73	0.09	0.08	0.18	0.71	---					10,900	0.337	228	241	43,235	228	2140	------	
5 C	2.85	0.75	2.10	2.15	0.72	0.18	0.07	0.18	0.16	0.62					12,350	0.339	241	241	61,800	241	2449	------	
6 C	2.92	0.77	2.15	1.88	0.86	0.11	0.08	0.18	0.75	---					9,350	0.226	228	228	52,910	228	1157	------	
7 C	3.07	0.66	2.41	2.35	0.99	0.09	0.08	0.24	0.35	0.28					11,880	0.380	228	217	47,020	217	2710	------	
8 C	2.86	0.66	2.20	2.41	1.04	0.10	0.07	0.18	0.25	0.25					12,500	0.354	241	241	57,230	241	2663	------	
9 C	2.89	0.64	2.25	2.34	1.02	0.10	0.08	0.20	0.70	---					12,180	0.355	241	217	44,090	217	1463	------	
10 C	3.16	1.02	2.14	1.86	0.55	0.08	0.06	0.41	1.08	0.73					11,340	0.252	286	277	60,500	277		------	
11 C	3.12	0.80	2.32	1.83	0.89	0.10	0.05	0.86	1.24	---					10,950	0.257	265	241	49,530	------	1728	------	
12 C	3.04	1.10	1.94	1.55	0.75	0.11	0.05	0.93	1.18	---					11,000	0.320	286	286	51,400	------	1424	------	
13 C	3.03	1.04	1.99	2.04	0.84	0.08	0.07	1.00	1.08	0.55					11,440	0.315	255	286	51,000	------	1440	------	
14 C	3.10	1.07	2.03	2.03	0.84	0.08	0.07	0.99	1.07	0.55					10,500	0.318	228	286	50,400	------	1289	------	
15 C	3.20	0.75	2.45	1.17	0.81	0.09	0.09	---	1.99	0.19					9,680	0.204	228	235	38,200	228	1699	------	41.0
16 C	3.16	0.68	2.48	1.17	0.81	0.11	0.09	---	2.03	0.20					9,460	0.200	228	228	40,800	228	1626	------	41.5
17 C	3.35	0.70	2.65	1.23	0.85	0.11	0.09	---	2.06	0.19					10,340	0.261	228	228	42,000	228	2243	------	44.5
18 C	2.87	0.73	2.14	1.22	0.70	0.10	0.10	---	2.23	0.18					10,560	0.275	255	255	47,650	255	1625	------	35.0
19 C	2.13	0.78	2.35	1.09	0.85	0.11	0.11	---	1.95	0.20					8,625	0.242	228	228	38,180	228	1175	------	41.5
20 E	2.75	0.59	2.16	3.21	1.04	0.08	0.03	0.13	1.09	1.03					14,850	0.364	255	302	69,000	286	2991	------	104.0
21 E	2.75	0.62	2.13	2.63	1.10	0.08	0.03	0.13	1.13	1.18					16,200	0.412	290	302	72,430	293	3581	------	104.8
22 E	2.76	0.74	2.02	2.34	1.08	0.08	0.02	0.13	1.10	1.03					14,050	0.347	277	307	73,000	293	2546	------	113.0
23 E	2.62	0.71	1.91	2.40	0.95	0.04	0.01	0.13	1.16	1.10					15,550	0.362	302	321	76,360	302	2968	------	97.9

*Melting units—C = Cupola; E = Electric Furnace; B = Brackelsberg Furnace.

Table 1 (Continued)
CHEMICAL AND PHYSICAL PROPERTIES OF 68 DIFFERENT IRONS

Melt. No.	Ing Unit*	T.C. %	C.C. %	G.C. %	Si. %	Mn. %	S. %	P. %	Cr. %	Ni. %	Mo. %	V. %	Ti. %	Cu. %	Treated	Transverse St'gth. lb.	Deflection in.	BHN center	BHN edge	Tensile St'gth. lb. per sq. in.	BHN	Resilience in.-lb.	Mod. of Rupture	Impact ft-lb.
24	E	2.57	0.75	1.82	2.30	0.95	0.03	0.02	0.12	1.35	1.10	---	---	---	---	14,750	0.380	311	326	79,740	311	2723	---	104.4
25	E	2.65	0.60	1.96	2.35	1.01	0.08	0.02	0.12	1.12	1.06	---	---	---	---	17,200	0.442	302	321	75,880	302	4192	---	91.4
26	E	2.58	0.73	1.85	2.51	0.95	0.03	0.02	0.12	1.12	1.02	---	---	---	---	16,340	0.408	302	321	76,780	302	3689	---	116.8
27	C	3.20	0.73	2.47	1.24	0.88	0.11	0.31	---	---	0.03	---	---	---	---	---	---	179	202	33,500	---	---	---	---
28	C	3.41	1.09	2.32	1.11	0.84	0.11	0.25	0.41	0.50	0.32	---	---	---	---	---	---	207	241	40,190	---	---	---	---
29	C	3.51	---	---	2.04	1.04	0.11	0.07	0.76	1.70	---	---	---	---	---	9,400	0.298	226	235	28,550	228	---	---	---
30	C	3.36	---	---	2.39	0.60	0.07	0.07	0.46	1.85	1.06	---	---	---	---	13,000	0.445	262	269	48,000	262	---	---	---
31	E	3.18	---	---	2.00	0.76	0.02	0.08	---	1.70	0.46	---	---	---	---	7,000	0.380	255	285	43,600	255	---	---	---
32	E	3.16	---	---	2.05	0.87	0.01	0.08	0.48	0.40	0.73	---	---	---	---	10,600	0.360	241	255	49,600	241	---	---	---
33	E	3.57	---	---	2.43	0.93	0.01	0.09	---	---	1.44	---	---	---	---	11,000	0.409	262	277	46,800	262	---	---	---
34	E	3.68	---	---	1.83	0.94	0.02	0.10	---	---	0.98	---	---	---	---	9,000	0.397	248	255	40,000	248	---	---	---
35	C	2.52	---	---	2.06	0.74	0.10	0.10	---	0.91	---	---	---	---	---	10,980	0.300	---	---	53,450	---	---	---	top 31.0
36	C	2.72	---	---	2.26	1.03	0.12	---	---	0.76	---	---	---	---	---	10,580	---	---	---	53,000	---	---	---	top 43.8
37	C	2.53	---	---	2.26	0.93	0.11	---	---	1.13	---	---	---	---	---	10,620	---	---	---	53,000	---	---	---	---
38	C	3.07	0.80	2.27	1.91	1.09	0.06	0.09	0.11	---	0.18	---	---	1.39	---	8,800	0.250	---	255	40,900	---	---	---	---
39	C	3.03	0.80	2.23	1.99	1.10	0.06	0.09	0.09	---	0.60	---	---	---	No	10,690	0.315	---	269	49,750	---	---	---	---
40	C	3.79	---	---	1.40	0.45	0.12	0.16	---	---	---	---	---	---	No	---	---	125	168	16,600	---	---	40,400	---
41	C	3.40	---	---	1.91	0.50	0.08	0.66	---	---	---	---	---	0.18	No	---	---	164	193	22,800	---	---	45,400	---
42	C	3.56	---	---	1.58	0.47	0.10	0.60	---	---	---	---	0.09	0.06	No	---	---	155	199	26,300	---	---	54,000	---
43	C	3.30	---	---	1.20	0.72	0.10	0.44	---	---	---	---	---	---	No	---	---	192	206	32,300	---	---	70,500	---
44	C	2.71	---	---	2.88	0.45	0.66	0.04	---	---	---	---	---	---	No	---	---	196	217	36,000	---	---	56,100	---
45	B	2.75	---	---	1.88	0.58	0.06	0.08	---	---	---	---	---	---	No	---	---	205	187	41,000	---	---	60,000	---
46	E	2.89	---	---	1.33	0.27	0.08	0.58	---	---	---	---	---	---	No	---	---	212	255	33,300	---	---	66,600	---

*Melting units—C = Cupola; E = Electric Furnace; B = Brackelsberg Furnace.

Table 1 (Continued)

CHEMICAL AND PHYSICAL PROPERTIES OF 68 DIFFERENT IRONS

No.	Melting Unit*	T.C. %	C.C. %	G.C. %	Si. %	Mn. %	S. %	P. %	Cr. %	Ni. %	Mo. %	V. %	Ti. %	Cu. %	Treated	Transverse Str'gth. lb.	Deflection, in.	BHN center edge	BHN	BHN	Tensile Str'gth. lb. per sq. in.	Resilience, in.-lb.	Mod. of Rupture	Impact, ft-lb.
47	C	3.03			0.00	0.49	0.12	0.42							No				221	245	39,900		70,300	----
48	C	3.80			2.05	0.75	0.06	0.05							1%ZrMnSi				207	207	47,300		86,300	----
49	C	1.97			2.30	1.05	0.10	0.21					0.11		1% "								84,900	----
50	C	2.43			1.81	1.30	0.09	0.17					0.11		¾% CaSi								66,000	----
51	C	3.65			1.34	0.58	0.10	0.11							¾% "				183	183	29,500		60,600	----
52	C	3.06			1.35	0.75	0.08	0.19						0.42	¾% "				207	215	47,700		88,300	----
53	E	3.45			2.44	0.56	0.05	0.20	0.61	2.45													45,700	----
54	C	3.11			1.68	0.53	0.05	0.26	0.56	1.97									207	228	33,700		65,300	----
55	C	3.27			1.33	0.43	0.08	0.59			0.54			0.36							37,800		66,000	----
56	E	3.52			1.73	0.96	0.03	0.04		1.10									255	262	43,100		73,200	----
57	E	3.18			1.12	0.84	0.07	0.58				0.44							192	280	38,500		67,500	----
58	E	3.50	0.61	1.89	2.73	0.98				1.02	1.03				{0.20Fe3O4 / 0.10FeSi}	16,200	0.439			302	76,550			120+
59	E	2.50	0.60	1.90	2.82	0.98				1.02	1.01					12,790	0.333			255	54,080			44.4
60	E	2.46	0.64	1.82	2.59	1.06				1.01	1.04			1.06	0.10CaSi	16,200	0.392			302	67,400			120+
61	E	2.50	0.65	1.85	2.80	0.99				1.04	1.06				0.10CaSi	17,100	0.435			286	74,800			112.1
62	E	2.44	0.65	1.70	2.84	1.03	0.01	0.01	0.02	1.06	0.99					13,200	0.412			302	55,880	302		77.3
63	E	2.36	0.51	2.35	2.79	1.03	0.01	0.01	0.04	1.07	0.12					9,460	0.360		302	207	48,780	207		41.0
64	E	2.63	0.48	2.15	3.10	1.07	0.02	0.02	0.03	1.13	1.09					13,850	0.450		285	302	59,400	203		88.3
65	E	2.62	0.60	2.02	2.50	1.11	0.01	0.01	0.21	1.05	1.00					12,000	0.327			293	59,450			46.9
66	E	2.57	0.64	1.93	2.44	1.10	0.01	0.01	0.21	1.03	1.02					13,130	0.342			331	60,100			59.2
67	E	2.59	0.69	1.90	2.45	1.07	0.01	0.01		1.03	1.02	0.11				18,425	0.248			311	69,550			70.8
68	E	2.90	0.67	2.23	2.52	1.58	0.01	0.01		1.51	1.02	0.06				13,700	0.291			302	72,600			71.1

*Melting units—C = Cupola; E = Electric Furnace; B = Brackelsberg Furnace.

data for a number of irons, to acquaint the design engineer and the user of gray cast iron with the mechanical properties now available in heavier section, insofar as these properties are represented by the A.S.T.M., type "C", 2-in. diameter arbitration test bar.

3. It is interesting to note that only 4 years ago, 60,000 lb. per sq. in. tensile strength cast iron in 2-in. section was considered impractical by cast iron specification committees. As will be seen, commercial irons are available today which give tensile strength in excess of 70,000 lb. per sq. in. in 2-in. section.

Plain and Alloy Irons Tested

4. In a survey of the field of high strength cast iron in heavy section, it is logical to discuss the subject from the standpoint of (1) plain carbon iron and (2) alloy iron. Quite often there is the possibility of using a low-carbon iron with higher base strength to compensate for the drop in unit strength resulting from increased section. Within well-recognized limitations, a low-carbon base iron of superior strength often will solve the strength problem introduced by increasing section size. The extent to which low-carbon, plain irons can be applied to heavy sections is limited by such factors as increased chilling tendencies, more rapid solidification and increased shrinkage, necessitating larger gates and risers. There is also the ever present difficulty of producing low-carbon iron consistently, since much better synchronization and control of operations, as well as other special knowledge and skill, are required.

5. To develop increased strength in larger section, and thus to widen the range of useful strength of cast iron, foundrymen also have introduced alloys, applying them in accordance with the principles established through experience with irons for light and medium sections.

Presents Data

6. The data presented below represent 68 different heats of iron produced by ten different companies, including heats produced both commercially and experimentally.

7. Table 1 lists the chemical analysis and physical properties of each iron. Complete information was not available for all irons but it can be said that the highest strength cupola irons were made from an all-steel charge plus ferro-alloys. The electric furnace irons were either "cold melt" or "duplexed". The type of melting unit
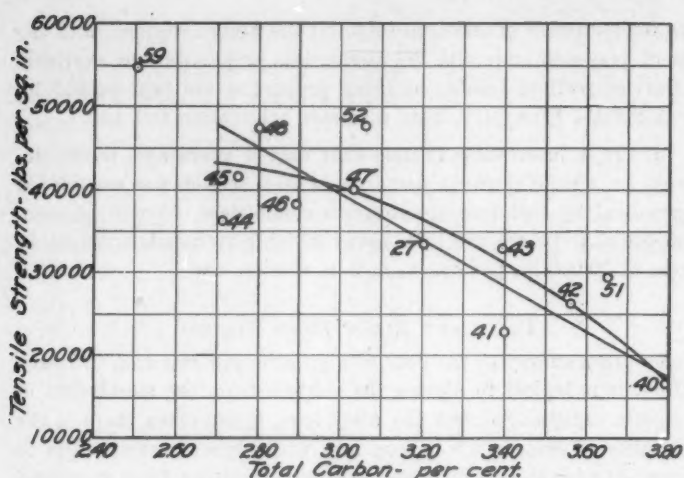
FIG. 1—TENSILE STRENGTH OF PLAIN CAST IRONS PLOTTED AGAINST PER CENT TOTAL CARBON.

is indicated for each heat. All data were obtained from 2-in. diameter transverse test bars. Transverse tests were made on bars 27-in. long and 2-in. diameter broken on 24-in. centers. Tensile tests were taken on bars machined to 1.25-in. diameter on centers (A.S.T.M. Type "C"). Brinell hardness was taken on the center and near the edge of the transverse and tensile bars which were
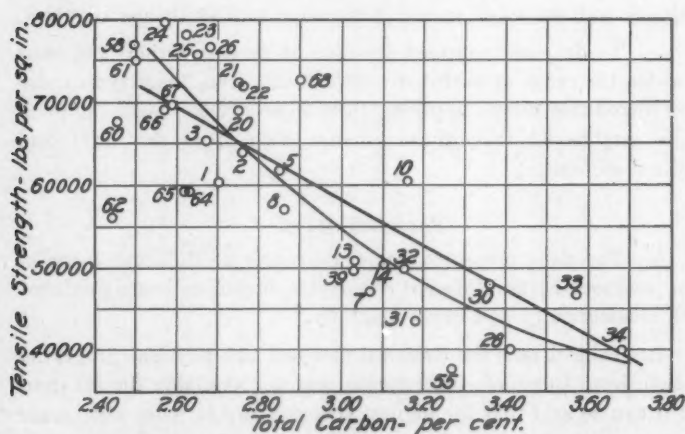
FIG. 2—TENSILE STRENGTH OF ALLOY CAST IRONS CONTAINING MORE THAN 0.20 PER CENT MOLYBDENUM PLOTTED AGAINST PER CENT TOTAL CARBON.

also used to determine modulus of resilience, modulus of rupture and impact strength. It should be noted that impact strength was determined on a 1.125-in. round, ground bar, machined from the broken half of the 2-in. arbitration bar, which was then broken on 6-in. supports on a Charpy impact machine of 120 ft.-lb. capacity.

8.   Figs. 1, 2 and 3 show tensile strength plotted against total carbon content for each of the 68 heats examined. To simplify the interpretation of results, these irons have been divided into three groups: Fig. 1, plain carbon irons; Fig. 2, alloy irons containing more than 0.20 molybdenum; and Fig. 3, alloy irons containing less than 0.20 molybdenum. This separation is purely an arbitrary one but is intended specifically to segregate those irons in which molybdenum has been intentionally added as a strengthener from
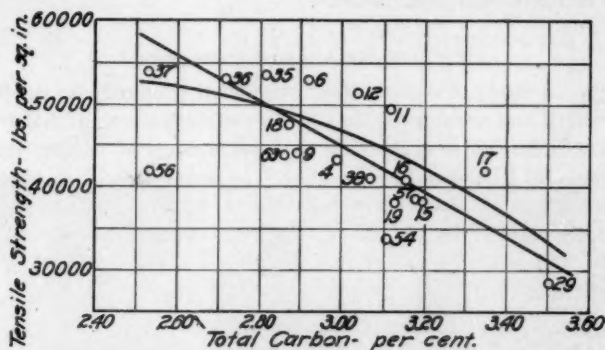
FIG. 3—Tensile Strength of Alloy Cast Irons Containing Less Than 0.20 Per Cent Molybdenum Plotted Against Per Cent Total Carbon.

those irons in which the presence of molybdenum was incidental. The serial number of each iron is shown on the charts to enable the reader to identify each iron listed in Table 1.

9.   The points on the charts are scattered. The scattering is a result of the differences in composition within each class and the fact that irons produced by various processes are grouped in the same class. If the mean curves are drawn intentionally to the same slope on each chart (on the assumption that the relation of tensile strength to total carbon content is constant regardless of the type of iron), these curves give a reasonable summation of the plotted points.

10.   On this basis, the following conclusions seem to be warranted:

a. Alloy irons containing less than 0.20 molybdenum show an increase in tensile strength over plain carbon irons of the order of 5,000 lb. per sq. in.

b. Alloy irons containing more than 0.20 molybdenum show an increase in tensile strength of the order of 14,000 lb. per sq. in.

11.   These figures are in agreement with the generally accepted alloy iron data based on the experience of the industry from 1.2-in. standard arbitration bars. This would indicate that the graphs for 2-in. bars cover a sufficient number of irons to be reasonably representative of the strength of irons in this section size.

12.   Table 2 shows comparative tensile values of identical irons in 1.2-in. and 2-in. sections.

Explains Balanced Composition

13.   A study of composition, considered primarily on the basis of strength and secondarily on the basis of deflection and hardness, offers information in connection with the subject of balanced composition. By "balanced composition" is meant that proportionate concentration of elements which results in the formation of a eutectoid matrix during the cooling cycle characteristic of the section. Structurally, the matrix should not contain free ferrite or free carbide.

Table 2
Comparative Tensile Values in Lb. per Sq. In.

Heat No.	2.0-in. Dia.	1.20-in. Dia.
1	60,400	83,900
2	64,000	82,700
3	65,400	75,550
31	43,600	45,600
32	49,600	55,500
33	46,800	48,700
34	40,000	40,275
35	53,450	58,900
58	76,550	82,280
61	74,800	74,100
62	55,880	71,280
63	43,780	48,400
64	59,400	72,600
65	59,450	52,450
67	69,550	67,000
68	72,600	74,750
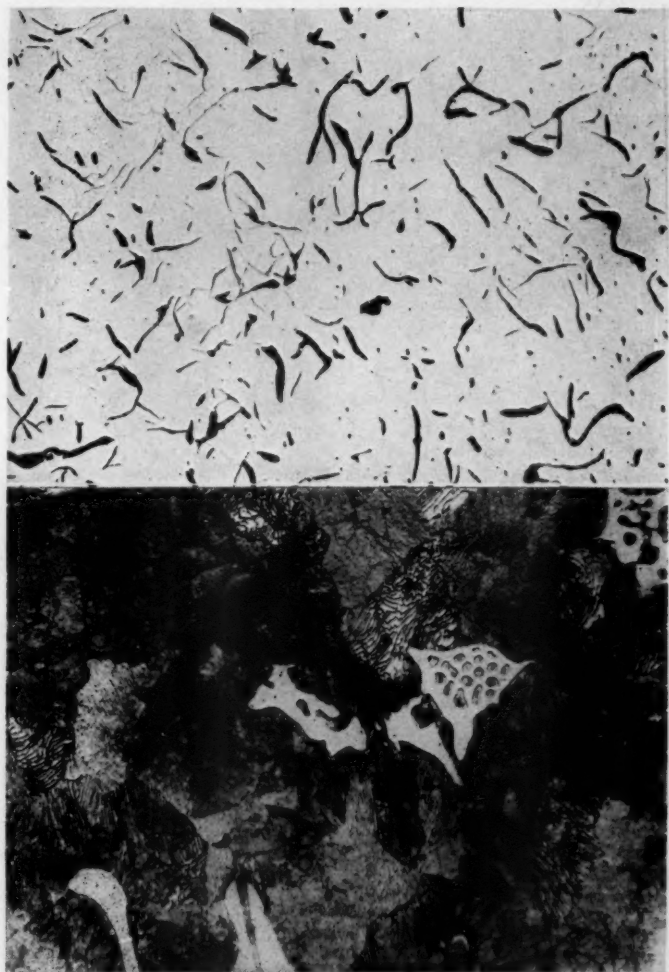
Fig. 4—Top—Iron No. 12, Unetched 100x. Bottom—Same Iron Etched with 2 Per Cent Nital, 500x.
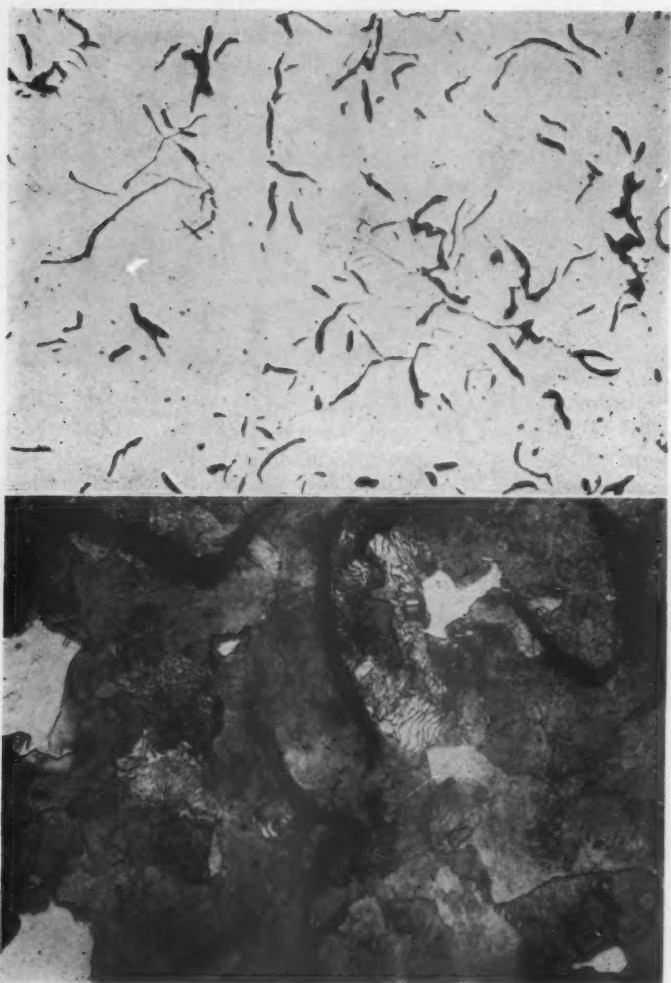
FIG. 5—TOP—IRON No. 14, UNETCHED, 100X. BOTTOM—SAME IRON ETCHED WITH 2 PER CENT NITAL, 500X.
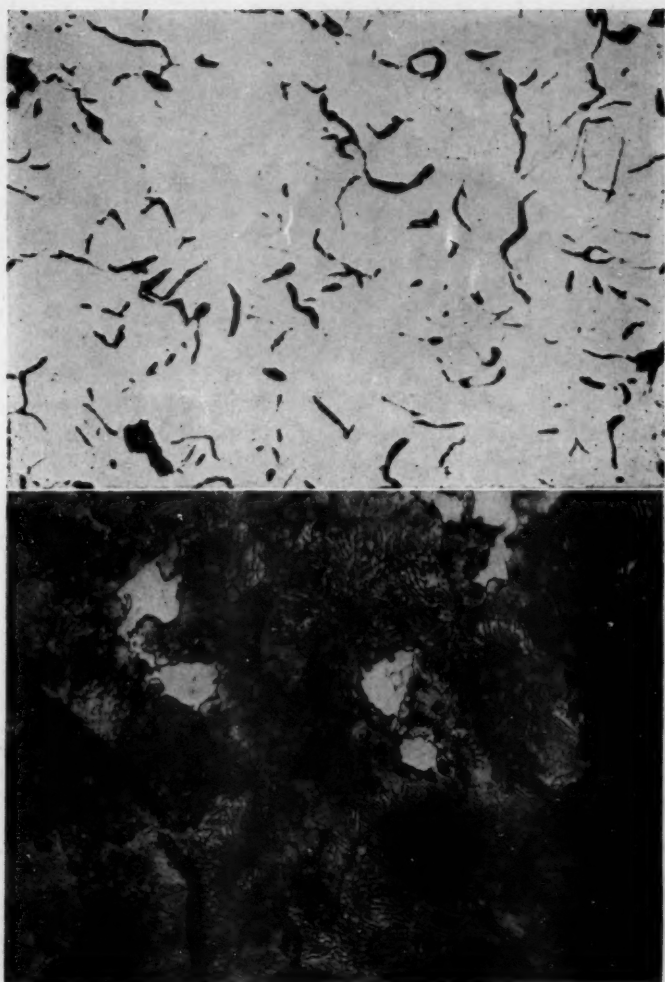
Fig. 6—Top—Iron No. 10, Unetched, 100x. Bottom—Same Iron Etched with 2 Per Cent Nital, 500x.
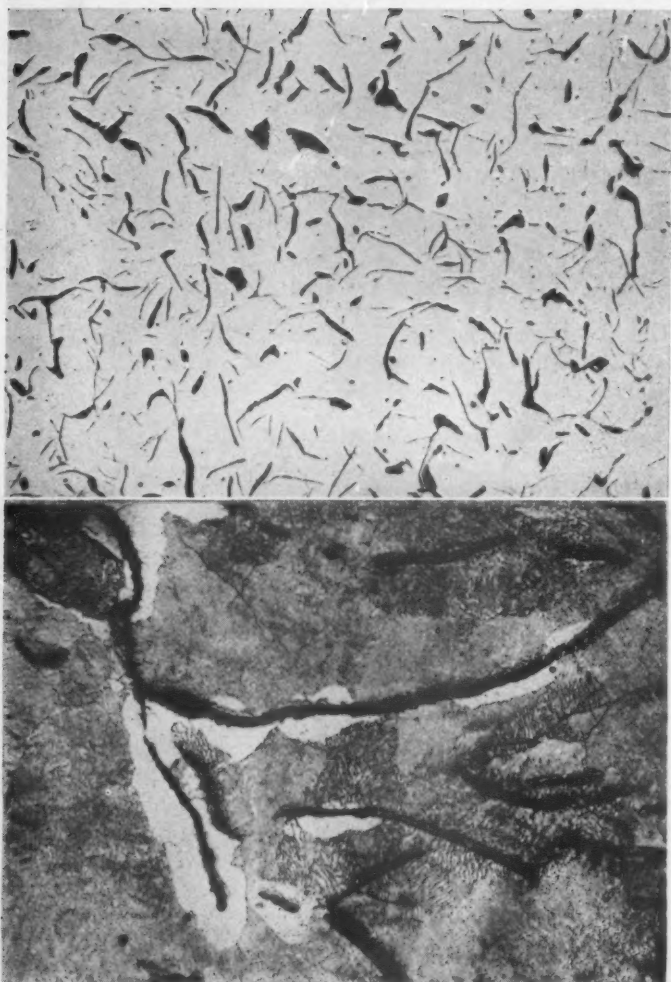
FIG. 7—TOP—IRON NO. 63, UNETCHED, 100X. BOTTOM—SAME IRON ETCHED WITH 2 PER CENT NITAL, 500X.

14. A balanced composition is prerequisite to optimum physical properties in any composition. For example, heats 13 and 14 (Fig. 5) are low in strength in view of their total carbon and molybdenum content when compared with heat 12 (Fig. 4). The base iron represented by heat 12, already containing an excessive amount of combined carbon, could not be improved by the addition of a potential carbide-forming element such as molybdenum. It should be noted that the increase in silicon content of approximately 0.50 has not brought the structure to desired balance. When the potential carbide-forming elements are decreased (as in heat 10), a marked improvement in strength results despite the fact that iron 10 (Fig. 6) is not to be regarded as an example of a perfectly balanced composition.

COMPARES RESULTS ON VARIOUS IRONS

15. Heat 63 (Fig. 7) is another example of poor structural balance for high strength in 2-in. section. Whereas heat 12 was out of balance on the hyper-eutectoid side, heat 63 is out of balance on the hypo-eutectoid side. The addition of carbide-forming elements would improve materially the chemical balance for higher physical properties. Numerous examples illustrating this can be found in Table 1. (Heats 5, 20-26 inclusive, 68).

16. Irons 15-19 inclusive, (See Fig. 8) show disappointing strength in view of their high alloy content. These are examples of poor composition balance. While the strength of iron 18 is better than the others of this group, the strength is low considering the total carbon content and alloy cost. Each of these irons could be improved by an increase in silicon content and the addition of more carbide-forming elements. The criticism of this group is not that a eutectoid matrix has been sacrificed, but that the strengthening characteristics of silicon have been sacrificed to produce that matrix. (Compare iron 18 with iron 5, Fig. 9.)

17. From the foregoing, it is apparent that, although balanced composition tends to improve the properties of cast iron, it does not in itself insure high strength, even when incorporated with alloy additions. A cupola iron which may be regarded as truly high strength material is represented by heat 5. It especially is signficant that this iron exceeds 60,000 lb. per sq. in. tensile strength at a Brinell hardness of only 241. The composition of this iron forecasts the direction in which the elements silicon and carbon should be adjusted to produce higher strength in heavy section.
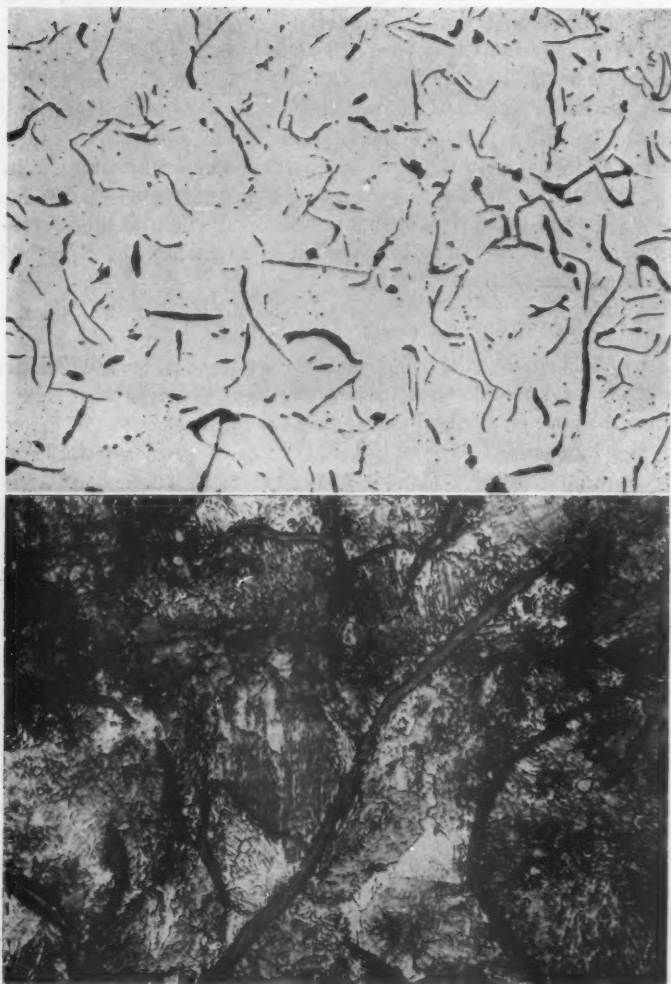
FIG. 8—TOP—IRON NO. 18, UNETCHED, 100X. BOTTOM—SAME IRON ETCHED WITH 2 PER
CENT NITAL, 500X.
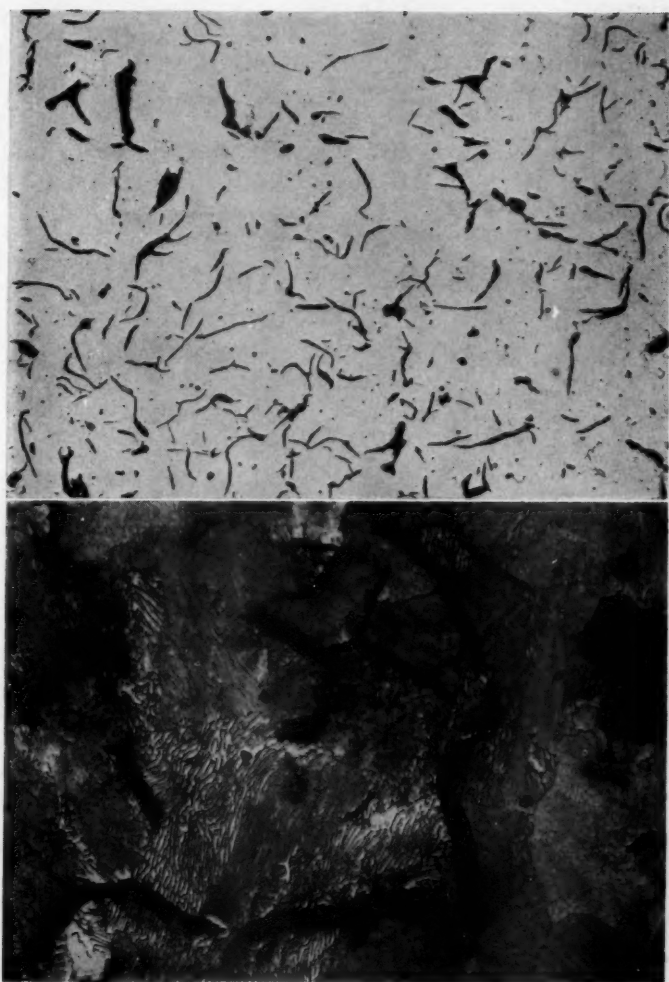
FIG. 9—TOP—IRON NO. 5, UNETCHED, 100X. BOTTOM—SAME IRON ETCHED WITH 2 PER CENT NITAL, 500X.
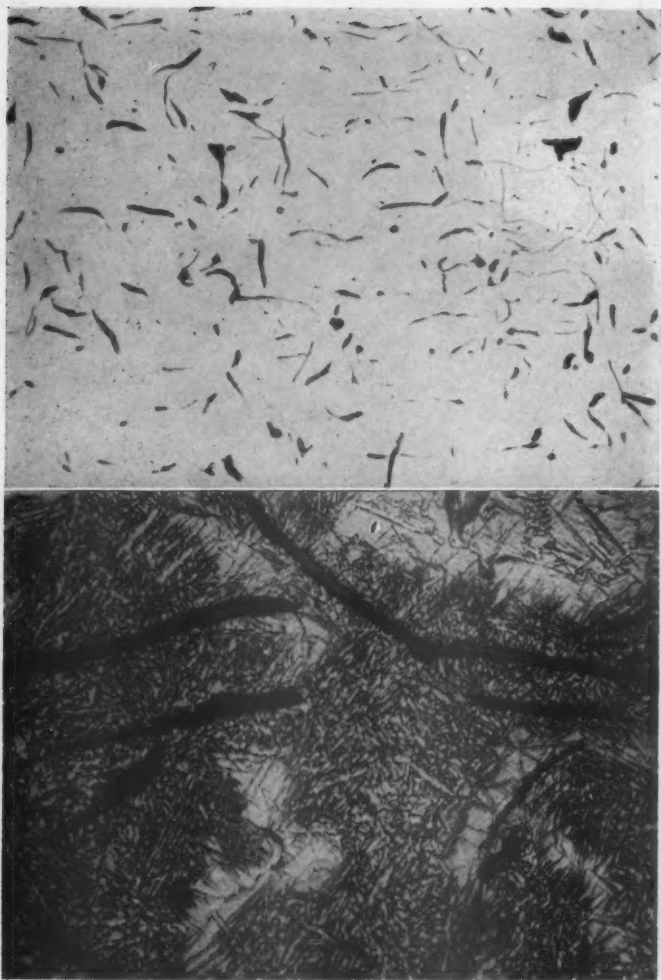
Fig. 10—Top—Iron No. 21, Unetched, 100x. Bottom—Same Iron Etched with 2 Per Cent Nital, 500x.
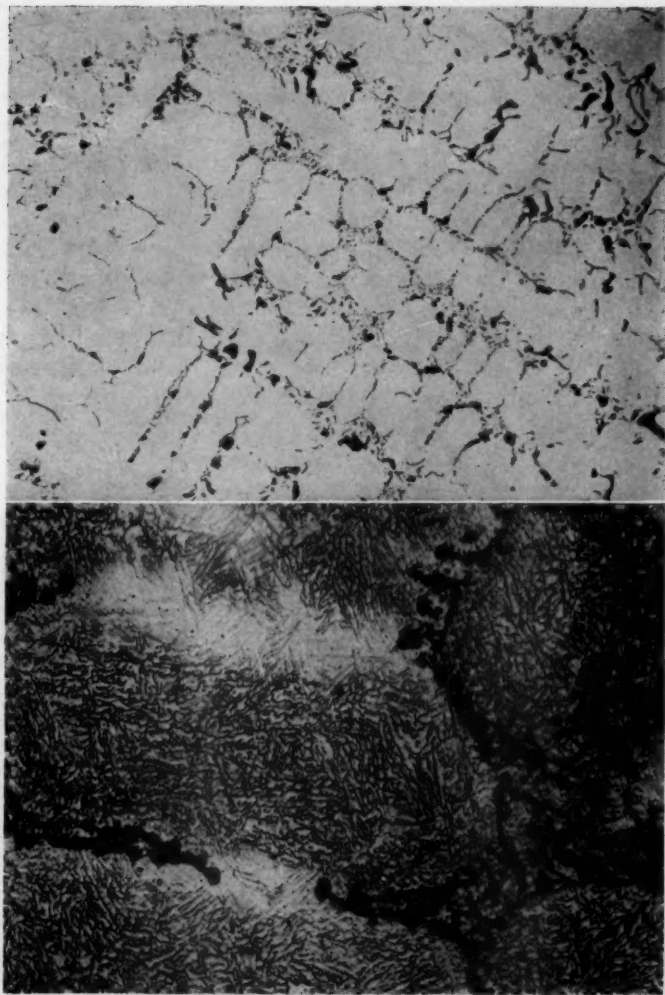
Fig. 11—Top—Iron No. 65, Unetched, 100x. Bottom—Same Iron Etched With 2 Per Cent Nital, 500x.
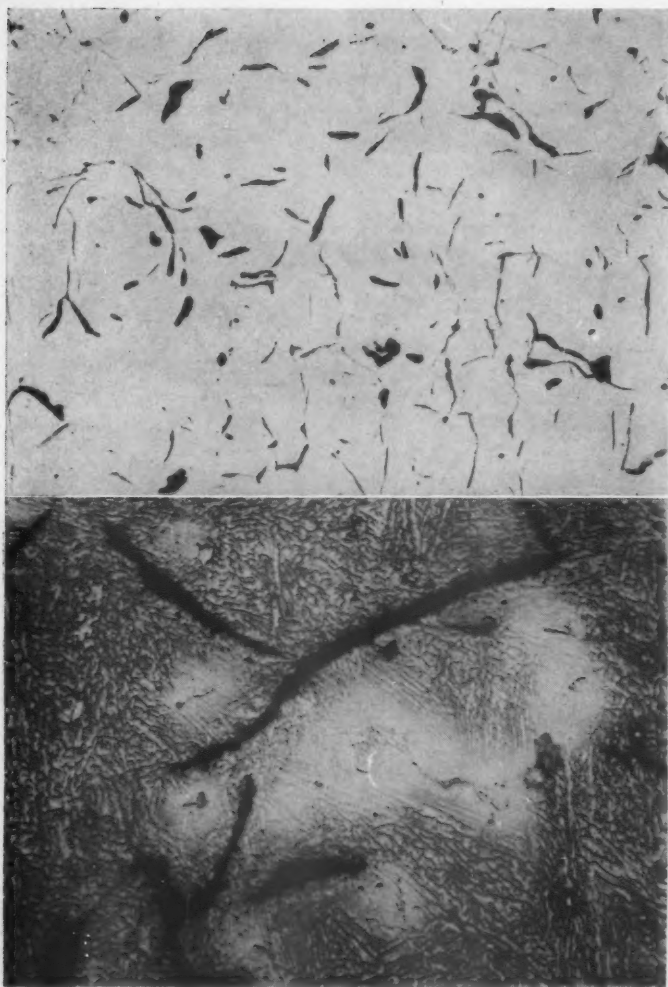
FIG. 12—TOP—IRON NO. 58, UNETCHED, 100X. BOTTOM—SAME IRON ETCHED WITH 2 PER
CENT NITAL, 500X.

18. Heat 3 illustrates the application of this adjustment. Carbon contents as low as that given in heat 3 scarcely can be regarded as regular cupola production. However, means are available for producing this composition more effectively and with greater consistency, as in the electric furnace. That commercial irons so produced possess the excellent strength characteristics forecast for them, is illustrated in heats 21-26 inclusive (Fig. 10).

19. Naturally, there are many occasions when all elemental prerequisites to high strength seemingly are present but the anticipated properties are not realized. Such a state prevails in heat 65 (Fig. 11). An examination of the microstructure reveals a desirable structure in the matrix comparable in all respects to the structure of iron 58 (Fig. 12). The failure to realize high physical properties is accounted for by the abnormal type of graphite resulting from overheating.

Conclusions

20. On the basis of these results, it can be said that 70,000 lb. per sq. in. tensile strength cast iron in 2-in. section unheat-treated is available commercially today, and that irons containing approximately 2.50 per cent carbon and 2.50 per cent silicon, when properly alloyed to obtain a balanced structure with approximately 300 Brinell and processed for normal graphite, produce maximum physical properties in 2-in. section.

DISCUSSION

Presiding: FRED J. WALLS, International Nickel Co., Detroit, Mich.

R. S. MACPHERRAN[1] *(Written discussion):* This paper contains much interesting information. It brings out very well the increase in strength, due to the additions of molybdenum. It would be very interesting some time to get a comparison with possible increases in strength, due to chromium, nickel, or vanadium.

While we are in agreement with the authors' findings on alloys, I cannot follow them in their conclusion that the adoption of a specification calling for 70,000 lb. per sq. in. tensile strength in a 2 in. section bar is at present justifiable. An analysis of their report shows that forty-one bars were cast from cupola iron. Of these forty-one bars, only five broke over 60,000 lb. per sq. in. and only two of these were over 62,000 lb. per sq. in. in tensile strength. Not a cupola made bar reached 70,000 lb. per sq. in.

The authors report twenty-six bars cast from the electric furnace. Of these, only nine passed 70,000 lb. per sq. in. and four more were between 60,000 and 70,000 lb. per sq. in. An electric furnace naturally lends itself to the making of small amounts of iron of a desired alloy composition. It certainly would not seem to us, however, that because isolated test bars have here and there been made to exceed a certain load, that, therefore, this load should be accepted as a specification for a minimum breaking strength. It would be most interesting to note which of these ten companies sending in results, would be willing to accept an order for a ten-ton casting, based on the minimum 70,000 lb. per sq. in. tensile strength in the 2 in. bar. The authors may not be aware that the various specifications making sub-committees of Committee A-3, A. S. T. M., met in March of this year, and decided not to favor the adoption of a 60,000 lb. or a 70,000 lb. per sq. in. tensile strength classification for cast iron.

In our own experience, I note that of the last twenty-six 2 in. bars cast of our high test iron, only four exceed 60,000 lb., and none go to 70,000 lb. per sq. in. While this strength could undoubtedly be increased by raising the alloys, we certainly would not like to manufacture under a specification involving a guarantee of a 70,000 lb. per sq. in. minimum.

There is nothing in the present setup to prevent a private agreement between buyer and seller, calling for an agreed minimum load and any foundry may agree to make iron of any desired alloy content. Both of the above are often done.

In view of the interest shown in higher test iron, it is an open question whether Committee A-3 should not be approached to appoint a new sub-committee to develop specifications for "electric furnace alloy iron," with recommended alloy compositions and specified characteristics.

W. H. ROTHER[2] *(Written discussion):* It is interesting to note the statement made by the authors of "Physical Properties of Cast Iron in

[1] Chief Metallurgist, Allis-Chalmers Mfg. Co., Milwaukee, Wis.
[2] Metallurgist, Buffalo Foundry & Machine Co., Buffalo, N. Y.

Heavy Section" to the effect that commercial irons are available today which give tensile strength in excess of 70,000 lb. per sq. in. in a 2 in. section. While this undoubtedly is a true statement, it is also a statement that could be misinterpreted very easily.

All plain cast irons in Table 1, regardless of melting medium are below the 60,000 lb. class, and no cupola cast iron listed in Table 1 is in the 70,000 lb. class. Only those irons melted in the electric furnace show tensile strengths of 70,000 lb. per sq. in. and over. However, these irons are loaded with alloys, which makes them prohibitive for competitive business.

Seventy-five per sent or more of cast iron today is cast from cupolas and due to limited tonnage obtainable in electric furnace melting, large castings with heavy cross section could not be cast from this melting medium.

Therefore, it is our contention, that in view of the large tonnage that is cast from cupolas in comparison to that of the electric furnace, cast irons having strengths of 70,000 lb. per sq. in. in 2 in. test bars cannot be produced in the cupola at the present time.

MR. YOUNG: We are gratified to have Mr. MacPherran's comments on high-strength iron, because he is a pioneer in that field of foundry work. We are certainly pleased that he checks with our findings on molybdenum. With regard to alloys other than molybdenum, there are considerable data in the paper but these are not sufficiently comprehensive to permit definite conclusions to be drawn. We did not try to generalize on them.

There was one very interesting point brought up on which I am not entirely clear, and that is as to whether A. S. T. M. specifications generally are considered as minimum rejection limits or an example of the typical properties of the iron. I think everybody will agree with Mr. MacPherran that if we were going to produce castings subject to a strength specification of 60,000 lb. per sq. in., we would want a minimum strength rejection limit on test bars no higher than 55,000 lb. per sq. in., even if the test bars were apparently sound. We do not contend that every casting or every test bar of a certain grade of iron will always meet the average strength accepted for that grade.

Mr. MacPherran questions our conclusions on the basis of the limited number of irons tested. Our thought in presenting these data was to survey the whole field from the available information on 2 in. sections, and there are many compositions included on which no one would expect to reach high strengths. However, the 60 and 70 thousand pound irons have been consistently produced when desired. Our conclusions are based only on those compositions which can reasonably be expected to reach high strengths.

I am quite interested in his last comment, and several of us on this A. S. T. M. Committee are in touch with recent developments. We understand that a survey to say whether or not the range of strength should

be extended is under consideration. If this is the case, we have accomplished one of the things we wanted to do with the paper — namely, draw attention to the fact that irons are being produced which are not now recognized by the present strength classifications.

G. P. PHILLIPS[3]: One thing that seemed to me to be of quite great interest in these micrographs shown was in connection with one in which the dendritic structure was obtained due to superheating, or overheating, of the correct structure to secure acicular pearlite and high-strength and high impact value, and that the only remedy is apparently avoiding the high temperature. Are there any steps one can take in some cases where he might want highly superheated iron for some practical handling reason. Is there anything one can add to the iron to overcome the bad effects of the superheating to avoid the dendritic structure?

MR. CROSBY: I believe that the late addition of silicon will go a long way towards eliminating this condition. Of course, the more you superheat, the larger the late addition of silicon necessary to obtain a normal graphite. If you have only a small amount of dendritic pattern due to a very slight overheating, 15 or 20 points of silicon as ferrosilicon added not more than two minutes before pouring should eliminate that condition. If the iron is superheated quite a bit, a large amount of silicon would be necessary to deoxidize the iron properly. We think the dendritic pattern is due to dissolved oxides and/or occluded oxygen. Therefore, the more pronounced the dendritic pattern, the more ferrosilicon would be required to produce a normal graphitic structure.

MEMBER: What do you call superheat?

MR. CROSBY: Any heat too high to get a normal graphitic structure. I will explain further by saying that if one is melting iron containing a large amount of graphite, a much higher superheating temperature can be tolerated. If an all-steel mixture is being melted, a low superheating temperature is necessary. In electric furnaces a temperature of 2700°F. is probably the maximum for an all-steel mixture, while 2900°F. can be easily tolerated with 3.75 carbon iron. The important consideration is how high a temperature can you use and still not have a dendritic pattern upon solidification. And that is something you have to determine for your own shop conditions.

F. J. DOST[4]: Why did you say to add the ferrosilicon not more than two minutes before you pour?

MR. CROSBY: We have tried 15 minutes and one minute before pouring and have found if the silicon is added too far in advance of the pouring, its effectiveness is lost — probably not entirely, but to a large extent. If added in the furnace nine or ten minutes before pouring, it has very little beneficial effect in eliminating the dendritic pattern, while when added immediately before pouring, as in ladle additions, the maximum effect is obtained.

MR. DOST: Mr. Crosby, does the amount of late addition have any bearing on the total carbon content of the metal coming out of the spout? If you have 2.50 or 1.75 total carbon coming out of the spout, will the

addition of silicon vary it in that case? In answering the question before, you said the amount of addition depended upon the degree of dendritism you got in the structure. Would that not depend somewhat upon the total carbon content. Would that not vary depending upon the total carbon?

MR. CROSBY: A dendritic structure is more likely to occur in low carbon than in high carbon iron. A 2.50 carbon iron is very susceptible, while with 3.50 carbon a dendritic structure is difficult to produce.

MR. DOST: Have you done any further study on graphitic additions to stop that? I wondered if you had found whether graphite would tend to eliminate it as much as silicon?

MR. CROSBY: We have not done any experimental work along that line. The reason we confined ourselves largely to the use of silicon for this particular purpose was because we were definitely of the opinion that when we added silicon, we were adding an element that also gave us some solid solution strengthening effect, at the same time eliminating the dendritic pattern.

MR. DOST: Since I talked with you some time ago, we have made the addition of both ferrosilicon and graphite and have had very good success in eliminating the dendritic structure. It takes a combination of both to do that.

CHAIRMAN WALLS: Have the authors made any comparative tests so far as Brinell hardness is concerned on the 1.2 in. and 2 in. diameter bars?

MR. YOUNG: There are some additional data which might be put in the final publication. The hardness values in Table 1 refer to center and edge of 2 in. transverse bars and also in some cases to the tensile bars machined from the transverse bars. Table 2, comparative strengths in 2 in. and 1.2 in. diameter bars, should be amplified with Brinells on both sizes for those few irons on which these figures are available. The hardness of the same iron in 1.2 and 2 in. section depends, as you know, on the type of iron. On the higher strength irons you do not get a wide spread between the two sizes.

CHAIRMAN WALLS: Is that same thing true between the top and bottom section of the broken test bar?

MR. YOUNG: Not definitely; there is very little difference.

CHAIRMAN WALL: In the tensile strength, is there any difference between the top and bottom half in the 2 in. bar? Can we expect to find the same difference as we found in the 1.2 in. bar?

MR. CROSBY: The range in Brinell hardness and tensile strength is not so wide in the 2 in. bars as in the 1.2 in. bars, from top to bottom, probably because of the fact that in the larger bar the metal has a better chance to heat the mold and become equalized in temperature.

MEMBER: A question relative to this dendritic structure and superheating. Does it make any difference as to what type of furnace you

superheat in? Does it make any difference whether it is a cupola, electric or air furnace? Is the dendritic structure about the same on the given temperature?

MR. CROSBY: If the composition remains the same, I do not think there would be any difference. I have seen pronounced dendritic structures taken from iron produced in a cupola as well as from an electric furnace. It is not so generally encountered in cupola iron as in electric furnace iron, because cupola irons have higher carbon. The higher the carbon, the less tendency towards dendritic graphite.

MR. DOST: Have you found any difference in the success of the silicon additions as to the size of the silicon you used? Did you notice any difference in the results, as compared with using a fine grade of ferrosilicon?

MR. CROSBY: If it is finer than 8 mesh, you seem to get a spontaneous oxidation. We do not think you get maximum efficiency from ferrosilicon if it flashes that way. So far as the maximum size is concerned, I would think somewhere around 8 mesh would be most efficient.

MR. DOST: I am referring to probably three or four tons of metal in one ladle.

MR. CROSBY: I would think in that case a quarter of an inch would be perfectly all right.

E. R. YOUNG, V. A. CROSBY and A. J. HERZIG (Authors written closure): Reference is made in the discussion to some additional available data, which could be added to Table 2, giving comparative Brinell hardnesses, in a few cases, on 1.2 in. and 2.0 in. transverse bars of the same iron. As revised by inclusion of these hardness values, where available, Table 2 is presented in Table 2A.

Table 2-A

Heat No.	2.0 in. Diameter Tensile Strength lb. per sq. in.	Hardness B. H. N. Center	Hardness B. H. N. Edge	1.2 in. Diameter Tensile Strength lb. per sq. in.	Hardness B. H. N.
1	60,400	269	286	83,900	302
2	64,000	265	269	82,700	302
3	65,400	265	269	75,550	....
31	43,600	255	285	45,600	....
32	49,600	241	255	55,500	....
33	46,800	262	277	48,700	....
34	40,000	248	255	40,275	....
35	53,450	......	......	58,900	....
58	76,550	......	......	82,280	321
61	74,800	286*	......	74,100	302
62	55,880	302	311	71,280	341
63	43,780	202	207	48,400	217
64	59,400	285	302	72,600	311
65	59,450	......	293	52,450	325
67	69,550	......	311	67,000	335
68	72,600	......	302	74,750	341

* On tensile bar machined from transverse bar.

Both Messrs. MacPherran and Rother have raised the question whether these data indicate that 70,000 lb. per sq. in is widely available in large castings. The authors have not so concluded and do not intend that these data should be extended to castings other than those actually represented by a 2 in. arbitration bar.

The authors intended only to gather and present data for 2 in. sections of iron now regularly produced. It must be borne in mind that many of the compositions tested and reported are not designed to give high strength in 2 in. section.

The matter of competition must be judged on the basis of high strength alloy electric furnace or cupola cast iron versus steel castings and not alloy electric furnace iron versus alloy cupola iron. On this basis the commercial use of alloy cupola iron exceeding 60,000 lb. per sq. in. in 2 in. section and alloy electric furnace iron exceeding 70,000 lb. per sq. in. section is already established.

The authors admit that 70,000 lb. sq. in. in 2 in. section is not promiscuously produceable but they know that iron of this grade, though perhaps not so specified, is today commercially available since it is regularly and consistently produced. However, the selection of bars for this investigation was such that the conclusion reached cannot be attacked from the standpoint that the high strengths reported are the results on isolated freak cases.

Modern Automotive Foundry Sand Practice

By A. S. Nichols*, Chicago, Ill.

The details of sand practice in seven foundries operated by automobile manufacturers were collected by the author. The foundries represented have a daily capacity of over 6000 tons of iron and the information as presented is intended to make available, to the foundry industry in general, the modern methods in use in the production and control of sand for the most important types of castings made. Detailed characteristics of synthetic sands bonded with different clays were determined and are included to provide information concerning some sand properties not regularly included in control tests.

1.   The production and control of foundry sands consisting of clay bonding of base sands, with seacoal added and reclaimed and excess sand utilized, has been pioneered and developed to a high state of perfection by automotive foundries. A study of the methods followed and standards established on sands used for cylinder blocks, cylinder heads and flywheels provides a cross section of the general practice.

2.   Lake sands, bank sands, silica sands and molding sands are used and their characteristics are outlined in Table No. 1. These sands are primarily used to provide a base grain of suitable size, and the A.F.A. fineness number shown for each sand provides a quick means of comparison.

3.   Tables No. 2, 3 and 4 show details of practice on individual sand handling units in 7 different foundries producing cylinder blocks, cylinder heads and flywheels respectively. Table No. 2, covering cylinder blocks, is discussed in the following paragraphs. Reference notes for Tables No. 2, 3 and 4, and 5 will be found at the conclusion of the paper.

* Vice President, Illinois Clay Products Company.

Table 1

CHARACTERISTIC—NEW SANDS USED

On Screen	Ref.	6	12	20	30	40	50	60	70	80	100	140	200	250	270	300	Pan	AFA Fineness No.	Perm.	Base Clay
Foundry Ref.*																				
1. Lake Sands																				
A	LA	0.0	0.1	0.1	0.6	6.6	32.5	32.3	15.4	8.2	2.7	1.4	0.1		0.0	0.0	0.0	50.0	170	
B	LB	0.0	0.0	0.0	0.0	12.08	36.54	14.0	12.58	14.7	8.53	0.0	0.0		0.0		1.17	55.0	275	
C	LC	0.0	0.0	0.0	0.3		50.7		41.3		7.0	0.2	0.1	0.0	0.0	0.0	0.0	47.0	280	
D	LD				1.4	10.7	34.4		42.4	10.3					0.3			46.0		
E	LE	0.0	0.0	0.0	0.0	11.5			77.8	9.8	0.2	0.3			0.3		0.1	49.5		
G	LG	0.0	0.0	0.2	3.1	17.0	43.5		31.3	0.2	4.7	0.0	0.0					43.0		
2. Bank Sands																				
B	BB	0.0	0.0	0.0	0.0	0.0	1.4	0.6	0.95	22.0	40.9	31.6			2.4		0.15	107	63	
D	BD		0.0	0.0	0.0	0.7	2.6		8.0	26.6	35.7	21.0			4.4		0.2	99		
E	BE	0.0	0.0	0.0	0.0	0.3			7.0	37.3	32.6	20.9			1.3		0.3	97		
3. Silica Sands																				
D	OCD			0.2	4.4	55.7	47.0		5.5								0.6	34		
E	OCE			1.5		77.2			19.9		1.2	0.2						36.4		
F	OCF			0.1		58.8		30.7		8.9	0.35	0.75	0.16	0.07		0.03	0.03	37.6	690	
F	OFF			0.0		1.47		2.05		42.6	13.5	31.1	6.6	0.8		1.05	0.6	94.3	121	
G	OG	0.0	0.6	0.3	10.5	36.4	28.9		11.5	1.0	3.4	1.1			0.3			41.3		
4. Molding Sands																				
D	MCD	0.7	3.9	6.3	3.3	8.9	19.2		26.9		19.5	4.0	3.0		1.0		3.3	59.1		
G	MCG	0.0	0.7	4.7	19.1	22.7	22.8		12.3	0.6	5.9	2.6	5.6		1.8		1.4	49.5		22%
G	MFG	0.0	0.3	0.5	0.6	1.6	2.5		3.7	0.2	6.9	2.9	44.0		26.1	10.5		158		10.3%

* In reference designation, last letter refers to foundry and preceding letter or letters to sand type where L = Lake, B = Bank, O = Silica, M = Molding Sand, C = Coarse and F = Fine.

CYLINDER BLOCKS

4.   Three foundries use lake sand, alone, as a base. The average A.F.A. fineness of these lake sands is about 50. Three others use silica sand with an average A.F.A. fineness less than 40. Another uses lake sand, silica sand and molding sand, the average fineness of the combination being about 46. The average quantity of new sand used per ton of iron is about 140 lb.

5.   The maintenance of permeability standards, in every case, represents a careful balance of the variables peculiar to each individual foundry. The average unit sand permeability is about 96, and individual permeabilities range between 68 and 128. Not one is troubled with blows or other defects attributed to low permeability, and even those using the coarsest sands, produce cylinder blocks which meet the high standards of surface finish and smoothness maintained in the industry.

6.   The grain fineness of the sands used, and the quantity added, apparently have a much greater effect in maintaining desired permeability standards than does the percentage of bond and sea-coal added to the sand.

7.   The development of multiple casting of cylinder blocks, involving the pouring of nearly 800 pounds of iron per mold, has in no way been hindered by any lack of suitable molding materials. Extremely coarse silica sands are used for this type of production, one foundry bonding with bentonite alone, and the other using 1/3 bentonite and 2/3 fire clay, each obtaining excellent results.

8.   In addition to the two foundries pouring two cylinder blocks per mold, five other foundries are included. One of these uses fire clay alone, two use bentonite alone, one uses bentonite and fire clay and another fire clay, bentonite and a special clay.

9.   The average bond consumption is about 38 lb. per ton of iron ranging from about 10 lb. to nearly 70 lb. The latter figure, from foundry C, may be considered inaccurate as representing true bond consumption on cylinder blocks only, as it is a shop average figure, and in other instances where both detailed and shop average figures are available, the consumption on cylinder blocks alone is lower than the entire shop average. The average strength of unit sands is about 2.6 lb. green shear, and slightly higher for facing where facing is used.

Table 2
CYLINDER BLOCKS*
General

	A	B	C	D	E	F	G	Avg.
Foundry								
Molding Method	J-S	S-S	J-S-S-R	J-S	S-S	S-S	S-S	
No. castings poured—8 hrs.[1]	3120	960	864	342-340-114[7]	880-520[11]	320	476-142[11]	
Pouring weight—lbs. each	194	288	267	315- 65- 47	215-237	391	286-341	
Tons iron poured—8 hrs.[1]	302.6	188.2	115.4	67.8	156.2	62.6	92.3	
Lb. green sand in mold	320	790	1278	814-197-158	830-940[4]	750[4]	778-970	
Lb. green sand per lb. iron[3]	1.65	2.74	4.77	2.58[12]	1.05-1.97	1.93	2.62	2.49
Lb. cores in mold	169	291	212	255-0-1.5	300-340	315	224-282	
Lb. cores per lb. iron[4]	0.87	1.01	0.70	0.71-0.14-0.08	0.69-0.72	0.81	0.70-0.83	0.80
Tons sand in unit	50	100	120	60	200	50	150	
Tons green sand molded	499.2	379.2	550.8	181.7	304.8	120.0	240.9	
Est. times sand used—8 hrs.	9.99	3.79	4.58	3.03	1.52	1.5	1.64	3.91
Tons new material add.—8 hrs.[2]	22.68	10.44	9.93[3]	0.86	23.0	4.81	8.4	
Lb. new material per ton iron	149.5	151.0	172.2	202.3	294.3	153.8	182.0	199.3
Method of adding	F & C.M.	F & D	C.M.	F & D	D	D	C.M.	
Mixing equipment in unit	PM	PM	PM & R	MX	MU	MU	MU	

New Sand-Permeability Factors

	A	B	C	D	E	F	G	Avg.
Tons new sand (Added—8 hrs. See table No. 1) { L, OC, MC }	16.85	9.0	5.76[2]	2.90 / 1.82 / 3.01	18.0	2.9	6.0	
Total tons new sand added	16.85	9.0	5.76	8.82	18.0	2.0	6.0	
Lb. new sand per ton iron	111.4	130.2	88.5[3]	201.7	280.5	62.6	130.0	139.9
Avg. permeability unit sand	68	113	86	75	120	138	85	96
Avg. permeability—facing	68	118	88	88				90
Lb. green sand per lb. cores[5]	1.0	2.7	6.0	4.4	2.8	2.4	3.2	3.12
Lb. bond and seacoal per ton sand molded[6]	23.4	7.5	18.5	33.7	32.8	31.3	19.5	23.5
A.F.A. fineness new sand	50.0	55.0	47.0	46.0	30.4	37.6	41.5	44.0

Table 2 (Continued)

Bond-Strength Factors

Foundry	A	B	C	D	E	F	G	Avg.
Total lbs. bond per ton iron	22.0	10.5	67.8[3]	49.0	48.0	31.9	39.0	38.3
Lb. fire clay per ton iron	22.0	----	55.5	17.8	32.0	31.9	----	31.8
Lb. bentonite per ton iron	----	10.8	12.8	26.7	16.0	----	32.5	21.7
Lb. special clay per ton iron	----	----	----	4.5	----	----	----	4.5
Lb. cereal per ton iron	2.8	----	----	----	----	----	6.5	6.5
Avg. green shear unit sand	2.8	2.8[10]	2.75	3.0[10]	2.5[10]	2.6	2.0	2.62
Avg. green shear facing	2.35	2.8[10]		3.2[10]				2.78
Avg. pouring temp. of	2558	2640	2700	2675	2675	2680	2675	2660

Seacoal-Moisture

Foundry	A	B	C	D	E	F	G	Avg.
Lb. seacoal per ton iron	16.1	10.0[13]	20.0[13]	41.7	16.0	29.1	13.0	20.9
% seacoal unit sand	10.5[38]		9.5[38]	5.5[37]	5.8[37]	5.0[38]	5.5[37]	
% seacoal facing	12.5[38]			7.0[37]				
% moisture—unit sand	5.2	3.45	4.0	4.25	2.5	3.0	2.7	3.56
% moisture—facing	5.2	3.25		3.75				4.05

Hardness

	Drag	Cope
Avg. mold hardness	70	70

81

* For reference notes and abbreviations see page 926.

Table 3
CYLINDER HEADS*
General

Foundry	A	B	C	D	E	F	G	Avg.
Molding Method	J-S	S-S	J-S-SF	J-S	[13]	S-S	J-S	
No. castings poured—8 hrs.[1]	3860	3000	960-1040[13]	340-342-114[7]		640-240[17]	310-750[15]	
Pouring weight lb, each.[2]	93.0	85.0	112-44	65-315-47[9]		98-137	78-588	
Tons iron poured—8 hrs.[3]	156.24	52.5	87.0	67.0		31.36-18.84	34.04	
Lb. green sand in mold.	107	107	320-228	197-814-158		400-225	199-104	
Lb. green sand per lb. iron[5]	1.8	2.1	2.8-3.5	8.0[5]		2.0-1.4	2.5[15]	2.6[15]
Lb. cores in mold.	61	85	75.6-8.0	9.0-255-1.5		10-200	7.5	
Lb. cores per lb. iron[6]	0.66	1.0	0.68-0.13	0.14-0.71-0.03		1.0-1.27	0.09[15]	
Tons sand in unit.	40	25	120	60		60-30	75	
Tons green sand molded.	285.6	160.5	272.1	181.7		52.7-40.0	69.8	
Est. times and used—8 lbs.	7.12	4.23	2.27	3.03		1.05-1.38	0.92	2.85
Tons new material add.—8 hrs.[2]	10.26	6.78		0.88		3.03-0.68[4]	8.48	
Lb. new material per ton iron.	181.8	257.6		292.3		193.0-71.6	498.5	240.7
Method of adding.	C.M.	D.	C.M.	F. & D.		F. & D.-D.	F. & D.	
Mixing equipment in unit.	P.M.	MU.	PM. & R	MX		MU-PM	MU	

New Sand-Permeability Factors

	A	B	C	D	E	F	G	Avg.
Tons new sand (Added 8 hrs. See Table No. 1) L	4.14	3.0		2.99			1.5	
B		3.0						
OC				1.82		1.0-0.325		
OF						1.15-		
MC							6.0[15]	
MF				2.01			0.5	
Total tons new sand added.	4.14	6.0		0.92		2.15-0.325	8.0	187.6
Lb. new sand per ton iron.	58.0	228.5		201.0		186.9-34.6	470.6	78
Avg. permeability—unit	71	65	85	75		90-90	70	71
Avg. permeability—facing	....	....		88		78....	Cope Drag 48 50[15]	
Lb. green sand per lb. cores[8]	2.7	3.0	4.2-28.5	22.0[15]		40.0-1.1	26.5[15]	16.0
Lb. bond & seacoal per ton sand molded[9]	42.8	9.5		33.7		22.4-25.9	18.6[15]	24.7
AFA fineness new sands	56.0	81.0	47.0	46.6		66.0-37.6	55.5	54.8

Table 3 (Continued)
Bond-Strength Factors

Foundry	A	B	C	D	E	F	G	Avg.
Total lb. bond per ton iron	47.2	19.0		49.0		18.4-14.8	12	39.7
Lb. fire clay per ton iron	47.2			17.8				32.6
Lb. bentonite per ton iron		9.5		26.7		18.1-14.8		17.8
Lb. special clay per ton iron		9.5		4.5				7.0
Lb. cereal per ton iron								
Avg. green shear—unit sand	2.2	2.2[10]	2.8	2.0[19]		2.4-2.4	1.75	2.4
Avg. green shear—facing				3.2[19]		3.0-----	Cope 3.2 Drag 2.8	73
Avg. pouring temp. °F	2555	2575	2650	2675		2625	2600	2613

Seacoal-Moisture

	A	B	C	D	E	F	G	Avg.
Lb. seacoal per ton iron	31.1	10.0		41.7		37.6-33.3	28.0	28.5
% Seacoal unit sand	10.8[36]		0.5[38]	5.5[37]		6.8-6.8[36]	6.5[37]	
% Seacoal facing				7.0[37]		7.0-7.0[36]		
% Moisture unit sand	5.2	3.3	4.0	4.25		8.0-8.0	Cope 4.5	3.9
% Moisture facing				3.75		8.0-----	Cope Drag[13] 4.5 4.0	3.8

Hardness

	A	B	C	D	E	F	G	Avg.
Avg. Mold Hardness	78					81-81		

* For reference notes and abbreviations see page 920.

Table 4
FLYWHEELS*
General

Foundry	A	B	C	D	E	F	G	Avg.
Molding method	J-S	S.S	J-S-S	J-S	J-S	S-S	J-S	
No. castings poured—8 hrs.[1]	4080	5700	3360	425-Misc.[10]	3600-2240-3360[11]	640-640[12]	750-810[13]	
Pouring weight lb. each	44.0	51.0	46.7	43.25	25-35-42	62-42	58.8-78	
Tons iron poured—8 hrs.[1]	89.26	145.4	78.5	27.85	153.6	33.8	84.04	
Lb. green sand in mold	91	186	145	107	250-430-330[14]	150-200[15]	104-199	
Lb. green sand per lb. iron[5]	2.07	2.68	3.10	2.40[22]	5.0[23]	2.42[22]	1.7[24]	2.77
Tons sand in unit	60	85	120	60	200	50	75	
Tons green sand molded	185.6	387.0	348.6	66.1	639.8	26.0	69.8	
Est. times sand used—8 hrs.	3.09	11.07	2.01	1.14	3.15	0.56	0.92	3.13
Tons new material add.—8 hrs.[2]	14.01	13.75	*	2.84	16.88	4.78	8.48	
Lb. new material per ton iron	313.6	190.8	*	207.8	220.0	287.0	498.5	286.8
Method of adding	C.M.	F.&D.	F.&C.M.	F.&D.	D.	F.&D.	F.&D.	
Mixing equipment in unit	P.M.	P.M.	P.M.&R.	MX	MU	MU	MU	

New Sand-Permeability Factors

	A	B	C	D	E	F	G	Avg.
Tons new sand (added 8 hrs. See Table No. 1) L	5.94	9.0	*	1.08	11.0	3.88	1.5	
OC				.48			6.0[13]	
MC				.55			0.5	
MF					.68			
B								
Total tons new sand added	5.94	9.0	*	2.11	11.68	3.88	8.0	
Lb. new sand per ton iron	133.0	123.8	*	154.3	151.4	283.0	470.6	211.0
Avg. permeability unit sand	73	95	85	85	88	118	70	57
Avg. permeability facing		95	70	88		95	60	82
Lb. bond and seacoal per ton sand molded	28.9	24.4		21.8	16.6	27.1	12.6[15]	22.1
AFA fineness new sands	50.0	55.0	47.0	46.5	52.1	57.6	55.5	49.1

Table 4 (Continued)
Bond-Strength Factors

Foundry	A	B	C	D	E	F	G	Avg.
Total lb. bond per ton iron	133.0	45.5		17.2	35.8	31.7		52.7
Lb. fire clay per ton iron	132.0							132.0
Lb. bentonite per ton iron		10.7			32.8	31.7		23.1
Lb. special clay per ton iron		34.8		17.2				26.0
Lb. cereal per ton iron				3.3	3.3			3.3
Avg. green shear unit sand	2.45	2.4^{10}	2.8	3.0^{10}	2.2^{10}	2.1	1.75	2.53
Avg. green shear facing		2.7^{10}	2.8	3.2^{10}		3.0	2.4	3.10
Avg. pouring temp. °F	2580	2575	2625	2675	2600	2625	2600	2611

Seacoal-Moisture

	A	B	C	D	E	F	G	Avg.
Lb. seacoal per ton iron	47.6	21.5		36.8	32.5	22.2	28.0	31.4
% Seacoal unit sand	10.75^{35}		9.5^{30}	6.0^{37}	8.7^{37}	6.8^{35}	6.5^{37}	
% Seacoal facing			9.8^{30}	7.0^{37}		7.0^{35}		
% Moisture unit sand	5.4	3.25	4.0	3.8	3.8	3.0	4.5	4.0
% Moisture facing		3.25	4.0	3.8		3.8	4.5	3.8

Hardness

	A	B	C	D	E	F	G	Avg.
Avg. Mold Hardness	78					78		

* For reference notes and abbreviations see page 926.

10. The average pouring temperature does not seem to have appreciable effect on the amount of bond used. The strength maintained seems to be the most important factor in this respect. Those molding by jolt-squeeze methods maintain substantially the same strength as those using sand-slingers.

11. The amount of seacoal used varies from 10 lb. to nearly 42 lb. per ton of iron. The average being about 21 lb. It has been suggested by the foundry reporting the lowest figure that where skin-drying of molds is practiced, there may be important possibilities for reduction in the amounts of seacoal used.

12. Moisture is carefully and frequently checked and controlled to standards of temper established by individual conditions and experience.

13. Mold hardness control, as a daily routine, is strongly advocated by the foundries using it, but is not yet generally followed in the other foundries consulted.

14. Most routine sand testing merely covers green strength, permeability, moisture and seacoal content, but one foundry finds it advantageous to run dry strength, deformation and flowability as additional measures of routine testing.

15. Facing is used by three foundries, and in these instances the bulk of new material additions are added through the facing.

Table 5
FACING SAND MIXTURES

Foundry Casting	A	B	D	D	F	G	G	B	F	G
	⌐Cylinder Blocks⌐		⌐Cylinder Heads⌐					⌐Flywheels⌐		
% Unit Sand	84.2	96.6	67.3	68.1[34]	45.4	26.7	58.9	97.0	48.9[*]	56.4
% Seacoal	2.8	0.4	3.9	3.9	6.7	1.2	2.6	1.0	8.7	1.4
% Fire Clay	0.4	..	..	..	..	..	..	..	..	..
% Bentonite	..	0.3	..	..	2.5	0.6	1.6	1.0	3.4	..
% Special Clay	..	..	0.7	0.5	..	..	..	1.0	..	..
% Cereal	..	..	..	..	..	..	..	..	..	..
% New Sand-L	12.6	2.7	11.1	14.1	..	..	14.8	..	..	21.2
% New Sand-B	..	..	..	..	..	..	..	..	..	..
% New Sand-OC	..	..	8.2	6.3	..	..	..	..	39.0	..
% New Sand-OF	..	..	..	..	45.4	..	..	..	..	..
% New Sand-MC	..	..	8.8	7.1	..	4.5	..	..	..	14.0
% New Sand-MF	..	..	..	..	..	66.8	22.7	..	..	7.0
Green Shear-Avg.	2.4	2.6[10]	3.2[10]	3.2[19]	3.0	3.3	2.8	2.7[10]	3.0	2.4
Permeability-Avg.	68	113	88	88	78	53	50	95	95	60
Moisture-Avg.	5.2	3.3	3.8	3.8	3.3	4.5	4.0	3.3	3.3	4.3
Seacoal %-Avg.	12.5[35]	..	7.0[27]	7.0[37]	7.0[36]	..	..	..	7.0[35]	..

NOTES FOR TABLES 2, 3, 4 AND 5

[1] Figures are given for one sand handling unit only, except see note 17.

[2] Includes new sand, bond and seacoal only; does not include cores or reclaim sand.

[3] All figures on additions based on shop total; detailed figures not available.

[4] Two castings in each mold.

[5] In each mold.

[6] Pounds added per day per ton of estimated total green sand molded per day.

[7] 342 Cylinder Blocks, 340 Cylinder Heads, 114 Bearing Blocks poured on same unit.

[8] Four castings in each mold.

[9] Pouring weight of 3 Bearing Blocks in one mold.

[10] Figures reported in green compression converted to green shear for comparison.

[11] Two sizes this type casting produced on same unit.

[12] Cylinder Blocks only.

[13] Shop average, all castings; detailed figures not available.

[14] Do not produce Cylinder Heads this foundry.

[15] 750 Flywheels and 310 Cylinder Heads poured on same unit.

[16] Cylinder Heads only.

[17] Two types cylinder Heads produced, different units.

[18] Molding sand MCG used as bond; included in new sand additions only.

[19] Several miscellaneous castings poured same unit. Total tonnage and total additions used in all calculations.

[20] 640 Transmissions, pouring weight 42 lbs. poured same unit.

[21] 2240 Transmissions and 3360 Clutch Housings poured same unit.

[22] Flywheels only.

[23] 50% Pitch.

[24] Also used for Flywheels.

[25] Also used for Transmissions.

[26] Estimated actual seacoal.

[27] Loss on ignition after drying.

ABBREVIATIONS USED IN TABLES 2, 3 AND 4

P.M.	Pug mill
R.	Revivifier
MX.	Mixer
MU.	Muller
F.	Facing
CM.	Concentrated Mixture
D.	Direct

One of these three also makes additions to unit sand through a concentrated mixture, while the other two make the balance of their additions direct to the unit sand. Two foundries use concentrated mixtures alone. Two others make all of their additions direct to the unit sand. In the latter two instances there are mullers in the unit, and in all the other cases, the greater part of materials added are mulled before being placed in the unit.

16. Composition of facings and concentrated mixtures are given in Tables No. 5 and 6.

Table 6
CONCENTRATED MIXTURES
PREMIXED UNIT SAND ADDITIONS

Foundry	A	C	G	A	C	A
Casting	Cylinder Blocks			Cylinder Heads		Flywheels
% Unit Sand	....	....	....	....	64.1	....
Reclaimed Sand	40.35	....	....	28.80	...	30.00
Seacoal	2.20	12.00	7.50	17.00	....	10.00
Fire Clay	16.10	28.00	....	25.40	28.4	30.00
Bentonite	....	....	18.00	....	0.4	....
Special Clay						
Cereal	....	....	3.50	....	...	....
New Sand	....	....	....	....	...	....
L	40.35	60.00	....	28.80	7.1	30.00
B	....	....	....	....	...	....
OC	....	....	71.40	....	...	....
OF	....	....	....	....	...	....
MC	....	....	....	....	...	....
MF	....	....	....	....	...	....

CLAY BONDS

17. Fire clays and bentonites are most widely used. Two foundries use cereal binders as auxiliaries. Natural bonded molding sand is used, also, in two foundries.

18. Clay bonds are usually compared and evaluated on the basis of amounts required to give a desired green strength, when mixed with sharp sand and brought to temper. The relationship between those most commonly used are generally well known, in this respect. Much additional information, of value, can be gained through a study of other properties transmitted to the sand by different clay bonds.

19. To provide data of this kind, two fire clays, widely used in the industry, one bentonite, and two special clays were mixed with lake sand (A.F.A. fineness 43.5) and with six percent

seacoal. Sufficient bond was used in each case to make a sand of approximately 10 lb. green compression strength, to simulate, as closely as possible the type of sands most widely used in automotive foundries. After being brought to temper the sands were tested as shown in Table No. 7.

20. The standard A.F.A. test procedure provides only for three blows of the rammer, in preparing the test specimen. Since the foundry will use the sand under varying conditions of ramming, it is believed that data on sands bonded with different clays, rammed to varying densities, would also be of interest. The samples tested as shown on Table No. 7 were therefore also tested for certain characteristics at 2, 3 and 5 rammer blows. These data are shown on Table No. 8.

Table 7

CHARACTERISTICS OF 10 LB. GREEN COMPRESSION, 6 PER CENT SEACOAL SYNTHETIC SANDS BONDED WITH DIFFERENT CLAY BONDS—

	Illinois Fire Clay	Ohio Fire Clay	Wyoming Bentonite	Northern Illinois Clay	Southern Illinois Clay
Percent Moisture	3.5	3.5	2.3	2.4	3.3
Green Permeability	119	110	163	186	171
Dry Permeability	135	124	167	187	200
Flowability (%)	78	83	85	84	79
Deformation	.0165	.0155	.013	.011	.008
Resilience	160	153	130	108	81
Green Shear (Lb.)	2.3	2.35	2.2	2.3	2.55
Grn. Tensile (Oz.)	11.8	15.5	16.0	14.6	13.0
Dry Comp. (Lb.)	59.0	56.0	23	38.0	59.0
Dry Shear (Lb).	12.3	13.5	5.3	8.2	11.2
Dry Tensile (Oz.)	86.0	74.4	20.4	17.8	58.5
Sintering Temp. (°F.)	2683	2565	2506	2592	2642
Expansion (in.)	0.0125	0.0192	0.0095	0.0117	0.014
Contraction (in.)	0.0105	0.023	0.0105	0.006	0.0145
Appearance after expansion-contraction test	Very small expansion cracks	Long expansion cracks	Pit Holes	Short expansion cracks	No cracks

Table 8

MOLD HARDNESS VARIATION

(Comparison of 10 lb. green compression, 6 per cent seacoal synthetic sands with different clay bonds, at 2, 3 and 5 rams)

Material	Illinois Fire Clay			Ohio Fire Clay			Wyoming Bentonite			Northern Illinois Clay			Southern Illinois Clay		
No. of Rams	2	3	5	2	3	5	2	3	5	2	3	5	2	3	5
Mold Hardness	84.5	89	91.5	85	88.5	91.5	84.5	90	92	87.5	89.0	91.5	88.5	88.5	91
Permeability Green	148	119	100	125	110	87	186	163	149	203	186	157	208	171	140
Compression Green	8.6	0.7	14.3	8.95	9.0	12.65	8.7	10.0	10.9	8.7	0.8	11.7	8.25	10.1	12.55
Deformation	0.015	0.0165	0.008	0.020	0.0155	0.021	0.005	0.013	0.016	0.013	0.011	0.0165	0.0135	0.008	0.0065

21. All tests shown in Tables No. 7 and 8 were made by standard or recommended A.F.A. procedure, in the Detroit laboratories of H. W. Dietert Company.

22. The foundries consulted in the preparation of this paper have been exceedingly generous in making available the data herein presented. The author wishes to make grateful acknowledgment to them for their cooperation.

DISCUSSION

Presiding: R. F. HARRINGTON, Hunt-Spiller Mfg. Co., South Boston, Mass.

H. A. DEANE[1] *(Written discussion):* Mr. Nichols' paper will be closely studied by many other gray iron foundries in addition to the automotive foundries. The data in this paper is very valuable because it was obtained from foundries that are highly competitive and are run on a quality as well as economical basis.

It is interesting to note that various sands and various bonds are used satisfactorily. This means that there is no one sand or bond that will give the best results. Therefore, the data in this paper can be used as a reference in checking one's own foundry practice and as a guide on different jobs.

The wide variation in permeability (68-128) shows that the molding practice can be adjusted to the sand being used. This also applied to the bond consumption.

We are surprised that the bond consumption on cylinder blocks is lower than the average for the shop because some burnt core sand usually goes into the heap. However, we presume that most of the cylinders are made on units where the castings are dumped soon after pouring and the sand is prepared by mulling. This procedure would require less bond than heap sand molding.

Mold hardness testing can yield much valuable information, especially if it is used as a check on doubtful jobs. We believe that this test will come into wider use within a short time.

The figures in Table 7 give some interesting data. Further work should be done on this to correlate it with actual molding.

M. B. CHAMARA[2] *(Written discussion):* This paper has been prepared very thoroughly and with utmost attention to details, but I would like to add some information on sea coal which may be of interest.

Our test practice for sea coal is quite old, but satisfactory. I have used this practice as far back as 1930 while at the American Radiator Co., and it was satisfactory for these light castings which are cleaned by merely brushing them with wire brushes. It has proved satisfactory in automotive foundry.

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[1] Deere & Co., Moline, Ill.
[2] Nash Motors Co., Kenosha, Wis.

The method consists of determination of total carbon in combustion furnace (20 minutes per determination). This consists of sea coal, sea coal's coke and carbonadeneous matter other than sea coal, no attempt is made to separate these constituents, the sea coal is being added continuously so much per each mold made, hence, the three constituents are to a limit in fairly constant proportion. In order to give comparative figures, the total carbon content is divided by 0.70 and the result is considered as active sea coal content or rather active carbonadeneous matter content.

Analysis is made on samples used for permeability, etc.; average analysis consisting of samples collected during several days, and variance analysis made from sample taken from one mold, in case of floor sand heap—the average consists of samples taken from all over the pile and variance from two individual samples.

The above method is simple and answers the purpose, and, no doubt, is used by many other foundries.

W. G. REICHERT[3]: Have these figures on expansion and contraction been taken from a one-inch specimen or a standard A.F.A. specimen.

MR. NICHOLS: A one-inch specimen was used.

MEMBER: May we have an expression from Mr. Reichert as to how that affects the properties of the test, as between the two different test bars or specimens?

MR. REICHERT: As long as we have a definite specimen, the comparative results will be the same, but if we compare a one-inch specimen with a two-inch specimen, I think we might get a different result.

CHAIRMAN HARRINGTON: Mr. Nichols, I believe you made a statement in regard to a concentrated mix and its method of make-up and mixing. Personally, I have not been one who believed that it was necessarily advisable to mix very effectively, as by mulling, a concentrated bond mass made up of new sand and clay. I have in mind that mixing it in that manner, while it would make an intimate mixture, it certainly would be a difficult mixture to later on distribute through the heap or through a sand system.

MR. NICHOLS: Those concentrated mixtures are usually mixed dry and the intention is to provide the maximum dispersion and the maximum possible development of the strength. One foundry, giving a list of four different concentrated mixtures, added a note stating, "The percentages of raw materials and the bond additions are varied in such manner as to effect desired changes in the system sand. In this way, control of the sand is maintained with regard to permeability, percentage of moisture, percentage of sea coal and system volume."

CHAIRMAN HARRINGTON: I am a believer in effective mixing, in muller mixing. In many shops, however,—for example, where the new material make-up may consist of a small portion of new sand as a

carrier and clay bond and its quota of sea coal, and particularly where the new sand may not be dry—any very great concentrated mixing at that point would make a mass that would be difficult later on to disperse through the heap. That was the point I was making, and I gained the impression that the automotive industry attemps to mix very thoroughly, possibly by mulling that concentrated make-up.

MR. NICHOLS: In the case of three foundries, yes.

MEMBER: In connection with this last point, I notice that some of the foundries, in adding this make-up, mix it dry, and that make-up is then added to the molds before they go to the shakeout in order to get an even distribution into the system. That does not apply, necessarily then, to what kind of method is used for mixing or mulling new sand; it applies in both cases. It gives a fairly even distribution. Is that right?

MR. NICHOLS: Yes. Two of these foundries add direct to the mold and the third adds direct to the system.

MEMBER: Do the automobile foundries show much concern about preventing the return of core butts, or burnt cores, to the system?

MR. NICHOLS: I believe the answer to that would be, because these synthetic sands permit very easy control regardless of the percentage of sharp sand that comes in from any course. The permeability is easily maintained and the only problem is one of molding volume. The data with regard to cores were included because it was felt that they might have a bearing on the permeability maintained and the amounts of bond material needed. It is difficult however—in fact, impossible—to make definite comparisons without more data than could be obtained.

CHAIRMAN HARRINGTON: I might say that in some foundries other than in the automotive industry, very often the core sand used is a core sand of a particular type, so it will contribute certain physical characteristics to the heap. In foundries that I have been in, they have in some instances possibly sacrificed so far as appearance of the core is concerned and used a coarser grained sand for the purpose of contributing a particular physical characteristic in the heap. That was a case where the core involved was not of particular importance and where they could change the grain of the core in order to let the butts that get into the heap give the heap a particular characteristic.

MEMBER: Relative to these percentages of moisture, do they refer to the moisture in the facing only, or do they refer to the whole heap?

MR. NICHOLS: The percentage of moisture shown for facing is that standard set for the batch of facing as it comes from a mill. The percentage of moisture shown for the unit is the standard at which the unit sand, that is, the sand actually handled, is maintained.

MEMBER: I notice those moistures in no case run much more than 5 per cent. Is that the moisture that they use in their heap sand?

MR. NICHOLS: That figure is shown for the purpose of condensing as much as possible. The particular foundry that has that moisture reports that their moisture standards are between 5.1 and 5.4 per cent.

MEMBER: Those moistures are considerably under the percentages that I have had anything to do with and I wonder if these foundries are using a lot more clay in order to keep the moisture down.

MR. NICHOLS: No; their clay runs about 22 lb. per ton of iron in that particular foundry, which, though it is difficult to draw comparisons, I believe is low. As a matter of fact, that moisture was the highest of any.

MEMBER: Are not these moisture contents for the heap sand lower than found in an average gray iron foundry?

MR. NICHOLS: I believe they are lower. If there are any other foundrymen present who could offer any statements as to the standards that they use in their moistures, it would be helpful, because I think you are absolutely right; there is a wide variation, anywhere from 2.50 up to 7 per cent. I think I can honestly say from what I have seen that this practice represents the best that I know of.

CHAIRMAN HARRINGTON: To answer your question, there are many foundries that are operating on different types of sand in contrast to the types of sand Mr. Nichols has described, in which the moisture contents are of the order of 7 and 8 per cent. It makes a great deal of difference as to the type of sands, the practices and the type of mixing that may be involved.

H. RAYNER[4]: In answering the question the gentleman asked about the core sand, I represent a large automotive foundry. Our experience has been that the burned core sand is usable but the cores that are shaken out unburned have a tendency to lump up, even though an attempt is made to grind them, and the only ones that are satisfactory are the burned cores. We are working on a process now whereby we will burn all our returned core sand at a temperature of about 1200°F., and we believe we can reclaim 100 per cent of the core sand by that method.

CHAIRMAN HARRINGTON: Mr. Rayner has brought out a very important point in this idea of actually taking the unburned butts and bringing them up to such a temperature, around 1200°F., to make them more useful.

MR. RAYNER: That is right. I was looking back through the literature and found the A.F.A. Sand Committee back in 1923 reported that the idea was tried, but those trying it were afraid of the results. I am not sure whether it was said that the grain structure was broken up at those temperatures or not, but that was the hazard anticipated. We have, however, found as a result of a number of experiments in heating up to temperatures high enough to destroy the clay bond and get rid of the combustible material, due to the core binders, that the core sand is in better shape after the burning than it is when we get it in the natural state from the lake. We can get rid of the fines that way and get any grain size that we want from the core sand by the

use of a classifier. Moreover, a temperature of close to 1,000 to 1,200°F. does not break up the grain structure of the ordinary lake sand that we get in our territory. We are quite enthused over the method. What it requires is a large kiln, similar to that used in the cement industry. The equipment is quite costly but very effective, and we figure we can save many thousands of dollars by using it.

We are handicapped for room. That is the reason why that system was not put in some ten years ago. We had an experimental system whereby we used a rotary heat-treating furnace and burned our sand, and we feel it is 100 per cent feasible.

CHAIRMAN HARRINGTON: Those of us who in the old days attempted to recover core sands, found real difficulty, and I believe the real difficulty was the fact that the carbonaceous matter was still there. Mr. Rayner is getting rid of it by this method.

MEMBER: Mr. Rayner, what effect has that heating on the sand reuse for core mixture?

MR. RAYNER: The sand for core mixtures is just as useful as it is for molding sand. In fact, it is more suitable.

MEMBER: Do you wash that sand after you burn it, Mr. Rayner?

MR. RAYNER: No, just use it as it is. In burning the sand, you have to control the atmosphere, because if you do not have the proper atmosphere in your reclaiming furnace, the sand will come out black with a carbonaceous coating. If you raise the heat too high and get a deoxidizing atmosphere, the sand will come out pasty.

Back in 1926 in our research we had quite a bit of difficulty with this rotary heat-treat furnace. When we raised the temperature up to about 1800°, instead of the sand flowing, it actually poured something similar to glass. But we overcame that trouble by cutting down the amount of flame, which was an economy too and produced an oxidizing atmosphere. But there was the problem of treating the sand after burning. That can be taken care of by adding a long extension on to the furnace and the sand pouring into it as it rotates. The incoming air from the original furnace passes through that extension and not only cools the sand, but heats the air, and that heated air, in turn, helps to reclaim the sand. The sand is then put through a classifier, the dust taken out and the grain size required retained.

CHAIRMAN HARRINGTON: We would be interested to know whether your core oil consumption or oil usage was increased by virtue of the fact that you used this reclaimed core sand.

MR. RAYNER: The grains of sand were exceptionally clean, under the microscope resembling small pearls. The carbonaceous material was burned away and the clay bond remained as a fine dust which was separated in the classifier. The resultant reclaimed sand did not use as much oil as the lake sand in its original state. It was dry clean and of the required grain size.

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